Carderock Division
Naval Surface Warfare Center
Bethesda, Md. 20084-5000

CARDIVNSWC-SSM-61—93/09 March 1993
Survivability, Structures, and Materials Directorate
Technical Report

Current Welding Consumables Research in the U.S. Navy

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ABSTRACT

One of the thrusts of U.S. Navy research is directed toward providing new and advanced material systems with improved properties, and developing methods and materials to construct current and future generation naval vessels more economically. It focuses on materials of construction, fabrication methods and consumables, and methods to ensure and enhance structural integrity. One specific area in this thrust is that of welding the Navy's high strength steels. Toward this end, work is being conducted on a number of topics pertaining to welding Navy steels with yield strengths exceeding 100 ksi.

Four tasks are discussed in this presentation. HSLA-100 Welding Consumables Development addresses evaluation of experimental compositions, data analysis, and identification of optimum compositions. Low-Carbon Bainitic Weld Metals discusses the effects of alloying on weld metal strength and cooling rate sensitivity, and the effect of titanium-bearing inclusions on weld toughness. Weldability Methodology addresses transformation expansion, weldability tests, diffusible hydrogen, and cracking models. Welding Fluxes discusses determination of flux composition and correlations with weld performance. Future research for each task is also described.

ADMINISTRATIVE INFORMATION

The work on HSLA-100 welding consumables development, weldability methodology, and welding fluxes reported herein was conducted under the sponsorship of the Ship and Submarine Materials Block Program (ND2B), Program Element 62234N. The program sponsor is Mr. I.L. Caplan, Carderock Division, Naval Surface Warfare Center (CDNSWC), Code 0115. The effort was supervised by Mr. R. DeNale, CDNSWC, Annapolis Detachment, Code 615. The work on low-carbon bainitic weld metals was conducted under the sponsorship of ASN (RDA) DRPM (SSN21), PMS 350, under Task Area 130-90.4, Program Element 64561N. The effort was supervised by Mr. T.W. Montemarano, CDNSWC, Annapolis Detachment, Code 614. This report was prepared under the high strength steel welding consumables task of Block Program ND2B.

BACKGROUND

One of the thrusts of U.S. Navy research is directed toward providing new and advanced material systems with improved properties, and developing methods and materials to construct current and future generation naval vessels more economically. It focuses on materials of construction, fabrication methods and consumables, and methods to ensure and enhance structural integrity. One specific area in this thrust is that of welding the Navy's high strength steels. Toward this end, work is being conducted on a number of topics pertaining to welding Navy steels with yield strengths exceeding 100 ksi. These tasks include welding consumables for high strength steels, low-carbon bainitic weld metals, weldability methodology, and welding fluxes.

The underlying goals of the research performed to develop welding consumables for 100 ksi yield strength steels and establish their weldability are four-fold:

(1) To produce a weld metal which can be deposited free of cracking over a wide range of conditions without preheat/interpass temperature controls while attaining the required yield strength and impact toughness;

(2) To produce a weld metal which can be deposited over a wide range of cooling rates without significant change in the weld metal yield or tensile strength and impact toughness;

(3) To rigorously understand the metallurgical aspects of the weld metal in order to take advantage of the relatively rapid cooling rates associated with fusion welding and the effects of reheating associated with multi-pass welding; and

(4) To fundamentally understand the interaction between chemistry, microstructure, diffusible hydrogen, welding conditions, and the cracking tendency of these high strength steel weld metals.

HSLA-100 WELDING CONSUMABLES DEVELOPMENT

INTRODUCTION

HSLA-100 is a low-carbon, copper-precipitation-strengthened steel developed to replace HY-100 steel for ship structures [Czyryca, 1990]. HSLA-100 has excellent resistance to heat affected zone (HAZ) hydrogen cracking and therefore does not require the preheat controls specified for welding HY-100. The primary benefit in using HSLA-100 as a replacement for HY-100 is the potential cost savings resulting from reduction or elimination of preheat.

No welding consumables have been developed specifically for use with HSLA-100. The current practice is to use MIL-120S [MIL-E-23765/2D, 1987] and MIL-12018 [MIL-E-22200/10B, 1989] consumables developed for use with HY-100. Unfortunately, these consumables require preheating and, in certain applications, postweld soaking, to prevent hydrogen cracking in the weld metal. Consequently, the primary advantage of using HSLA-100 as a replacement for HY-100, namely the ability to weld without preheat, can not be fully realized due to limitations of the available welding consumables.

The objective of the present study is to address this need by developing a wire for gas metal arc welding (GMAW) of HSLA-100 steel. The desire is to design a wire that meets the mechanical property requirements of MIL-120S consumables, namely:

(1) Yield strength from 102 ksi to 122 ksi, and

(2) Minimum CVN impact toughness of 45 ft-lb at -60°F, and 60 ft-lb at 0°F,

and that can be safely welded without preheat or postweld soaks.

The alloy selected for evaluation was a low carbon, Ni-Mo-Ti system. As mentioned earlier, a primary objective is to develop a system that has better resistance to weld metal hydrogen cracking than MIL-120S consumables. The experimental alloy system has a lower C content than the MIL-120S wire and uses no chromium. These modifications are intended to improve resistance to hydrogen cracking. To compensate for the anticipated strength loss from these reductions,

the Ni-Mo-Ti design uses more molybdenum and nickel than the MIL-120S system. Molybdenum is reportedly a more potent strengthener than chromium [Enis and Telford, 1968], but has a less detrimental effect on hydrogen cracking resistance [Hart, 1986] and toughness [Enis and Telford, 1968; Oldland, 1988]. Nickel has been shown to be a moderate strengthener and improves impact toughness by reducing the transition temperature [Gross, 1968]. Of the major alloying elements, nickel also has the least detrimental effect on weld metal hydrogen cracking [Hart, 1986]. Additionally, it was decided that an increase in titanium content could be beneficial. Several authors have found that small additions of titanium can increase both strength and toughness [Gross, 1968; Dorschu and Lesnewich, 1964; Evans, 1992]. Moreover, there is no published evidence that titanium reduces hydrogen cracking resistance.

This program is divided into four phases. The first phase examined 16 variants of the low carbon, Ni-Mo-Ti system. The second stage used results from the first phase to select and evaluate nine additional compositions. In both Phase 1 and Phase 2, flux-cored consumables were used to reduce the cost of producing such a large number of weld metal compositions. Phase 3 involved analysis of data from all 25 experimental formulations and selection of the optimum compositions. The fourth phase, termed the Solid Wire Verification Phase, evaluates the optimum compositions in solid wire form. This paper reviews the results from the first 3 phases of the program. The Solid Wire Verification Phase is ongoing; therefore, only the plans for this phase will be discussed.

PHASE 1: INITIAL EVALUATION OF EXPERIMENTAL COMPOSITIONS

The objective of the initial phase of the investigation was twofold. The first objective was to determine whether adequate weld metal mechanical properties could be achieved in a low carbon, Ni-Mo-Ti system. The second objective was to evaluate the influence of carbon, nickel, molybdenum, and titanium on weld metal mechanical

properties.

Sixteen compositions were evaluated using a 2^4 full-factorial design [Ott, 1988]. The target weld metal contents for each element are presented in Table 1.

Sixteen weldments were fabricated, one using each of the formulations. The weldments were produced in 1-in.-thick HSLA-100 plate using GMAW with pulsed transfer. The joint design was a single-V-groove, with a 60° included angle, and 1/2-inch root opening. The backing bar was 1/2-in.-thick HSLA-100 plate. All welding was performed in the flat position. The shielding gas was 95% argon - 5% carbon dioxide. The heat input range was approximately 51 kJ/in to 53 kJ/in.

The preheat and interpass temperatures were maintained at 225°F to 275°F. Originally, all weldments were to be fabricated without preheat. This approach was rejected in favor of the higher temperatures for three reasons. First, recall that the goal of this program is to develop a solid wire for GMAW. Flux-bearing consumables are typically more susceptible to hydrogen damage than solid wires. Thus, it would be difficult to determine whether hydrogen effects, such as cracking or reduced tensile ductility, were related to the alloy design or the type of consumable used. Secondly, the primary objective of this phase was to evaluate the strength and toughness potential of the alloy system. Hydrogen damage could obscure the influence of the chemical composition on mechanical properties. Finally, a critical test of this low carbon system is its ability to meet minimum yield strength requirement. The higher preheat and interpass temperature, and thus slower cooling rates, provide a more severe test of the yield strength of this system.

Duplicate all-weld-metal tensile tests were conducted on each weldment. Charpy V-notch (CVN) performance at -60°F and 0°F was evaluated for each weldment. Five CVN specimens were tested at each temperature. Both the tensile and CVN tests were conducted in accordance with ANSI/AWS B4.0 [1985]. Weld metal chemical analyses were performed on remnants of tested tensile specimens.

Table 2 presents the results of weld metal chemical analyses and mechanical

property tests. A reasonable amount of deviation from the target levels occurred for carbon, nickel, and molybdenum. Though the target levels were not achieved, the high and low levels were fairly consistent for all of these elements. Weld metal titanium contents were consistently below the target levels. In fact, the high target, 0.03 wt%, was never achieved. Manganese varied from 1.36 wt% to 1.57 wt%, despite attempts to hold it constant at 1.5 wt%. In addition to these elements, the oxygen contents also varied significantly.

The tensile and CVN data reveal that four of the formulations met the yield strength and CVN energy goals. This confirms that the low carbon, Ni-Mo-Ti system is capable of meeting the mechanical property requirements for current MIL-120S wires.

Wong and Hayes [1990] performed stepwise regression analyses on the data presented in Table 2. The analyses were conducted to quantify the effect of carbon, molybdenum, nickel, and titanium on mechanical properties. Table 3 presents the resulting regression equations. The regression coefficient (R^2) is presented for each equation.

Equation (1) of Table 3 presents the relationship between the test elements and yield strength. The high R^2 value indicates that the equation accurately describes the variation in yield strength. This equation illustrates that carbon is the most potent strengthener, followed by molybdenum. Nickel provided mild strengthening; whereas titanium decreased yield strength. Additionally, eq. (1) shows that there was an interaction between titanium and nickel, and between titanium and molybdenum. The terms indicate that the strengthening effect of nickel and molybdenum increased with titanium.

Equations (2) and (3) of Table 3 present the relationship between the test elements and CVN energy. Titanium was found to be statistically significant only when expressed as titanium-to-nitrogen ratio (Ti:N). Thus, unlike yield strength, these equations use titanium to nitrogen ratio rather than the titanium content.

The R^2 values indicate that the equations described the CVN energy variation reasonably well.

According to eqs. (2) and (3), carbon had the most deleterious effect on CVN energy. Molybdenum also decreases CVN energy, though to a much lesser extent than carbon. Though not obvious from inspection of the equations, both nickel and Ti:N improve CVN energy. The beneficial effect of nickel was particularly strong at -60°F. The interaction terms indicate that the detrimental effect of carbon and molybdenum was lessened by increases of nickel and Ti:N.

As discussed in the following section, these results were used to determine the matrix of formulations evaluated in Phase 2.

PHASE 2: EVALUATION OF ADDITIONAL COMPOSITIONS

This phase of the program examined nine additional experimental formulations. The base metal, welding conditions, and test procedures used in this phase were the same as those employed in Phase 1. Based on the beneficial effects of nickel and Ti:N revealed in Phase 1, it was decided that Phase 2 should evaluate higher levels of these elements. Additionally, because the data in Table 2 suggests that manganese influenced yield strength, this element was varied systematically.

Table 4 shows the nine target formulations. The compositions were selected to evaluate the effect of manganese in the range of 1.25 wt% to 1.75 wt% and the effect of nickel at levels beyond those examined in Phase 1. In addition, formulations 20 through 24 were designed to evaluate the influence of specific Ti:N ratios between 2.0 and 3.6. The remaining formulations did not target specific Ti:N ratios, but instead, were designed to reproduce the range of ratios seen in Phase 1 (approximately 1 to 3). Finally, formulations 17, 18, 19, 24, and 25 targeted carbon levels that were slightly different than those evaluated in Phase 1.

Table 5 presents the weld metal chemical analysis and mechanical property results. The weld metal carbon, manganese, and nickel contents were generally in

good agreement with the target values. As was the case with the Phase 1 compositions, silicon and titanium contents were consistently below the target levels. In the case of titanium, this caused some of the Ti:N ratios to be below the target levels.

The mechanical property data reveal that only formulation 19 met the mechanical property goals for strength and toughness. The majority of the formulations met the minimum yield strength requirement. Conversely, with the exception of formulation 19, the CVN energies were unacceptable. Examination of the effects of the individual elements revealed that, when nickel was held between 2.5 wt% and 3.3 wt%, CVN energy at -60°F increased as manganese varied from 1.27 wt% to 1.7 wt%. For the same nominal manganese content, varying nickel from 2.5 wt% to 3.3 wt% had little effect on CVN energy. Increasing nickel beyond 3.3 wt% caused a decrease in CVN energy. The Ti:N variations had no discernable influence on CVN energy.

Using the chemical compositions in Table 5, mechanical properties were calculated from the regression equations presented in Table 3. Table 6 compares the actual and calculated values. This comparison was performed to help explain the disappointing mechanical property results and the observed effect of the individual elements.

In general, the calculated and measured yield strength values showed reasonable agreement. Formulations 19 and 25 showed the largest discrepancies. Equation (1) of Table 3 significantly underestimated the yield strength of formulation 19. This may have resulted because the equation was unable to account for the strengthening effect of manganese (raised to 1.7 wt% for formulation 19). Conversely, the equation grossly overestimated the yield strength of formulation 25. The difference probably occurred because the nickel content of formulation 25 was well beyond the range used to develop the equation. The overestimation suggests that the strengthening effect of nickel diminished at a level between approximately 4.0 wt% (formulation 20) and 6.9 wt%.

Examination of the CVN data in Table 6 revealed that the calculated values typically overestimated the actual absorbed energies. The one exception was formulation 19, which performed slightly better than predicted. This was attributed to the inability of eqs. (2) and (3) to account for the beneficial effect of manganese.

The measured absorbed energies from formulations, 18, 20 through 23, and 25 were significantly below the calculated values. There are several potential reasons for the large discrepancy:

(1) Some of the elements were evaluated at levels beyond those used to develop the equations.

(2) All of the compositional variables affecting impact toughness were not included in the equations. Specifically, elements such as manganese, silicon, and oxygen, that were intended to be held constant, showed some variation.

(3) There may be complex interactions and non-linear relationships that cannot be addressed with a simple linear model.

The unexpected performance of the Phase 2 formulations demonstrated that the influence of the test elements on the impact toughness of the low carbon, Ni-Mo-Ti system was not well understood. Therefore, additional data analysis, incorporating all 25 formulations, was performed to develop a better understanding. The following section briefly describes the analysis and resulting identification of the optimum weld metal compositions.

PHASE 3: DATA ANALYSIS AND IDENTIFICATION OF OPTIMUM COMPOSITIONS

A multiple regression analysis was conducted on the yield strength data from all 25 formulations. Carbon, manganese, nickel, molybdenum, silicon, and titanium were selected as the independent variables. The analysis revealed that carbon, manganese, nickel, and silicon affect yield strength. The influence of titanium was not statistically significant. The resulting regression equation is expressed as a

carbon equivalent-type expression (CE(YS)) in Fig. 1. Note the excellent agreement between the calculated and measured values. It should be noted that, because yield strength is strongly influenced by cooling rate, this expression is only valid for the nominal weld metal cooling rates experienced by 25 formulations.

The CE(YS) expression indicates that carbon is the most potent strengthener. Additionally, the equation predicts that molybdenum provides slightly more strengthening than manganese, and that nickel provides mild strengthening. These trends are consistent with previous investigations [Enis and Telford, 1968; Oldland, 1988], suggesting that this part of the analysis is valid.

The analysis appeared to overestimate the silicon effect, indicating that it provides significantly more strengthening than either manganese or molybdenum. The relatively small variation in silicon, when compared to the variations in the other elements, may have magnified its effect in the regression analysis.

Unlike the yield strength analysis, a multiple regression analysis that included all 25 formulations failed to explain the influence of alloying on CVN energy. Possible reasons for the ineffectiveness of this approach were discussed earlier. One of the primary obstacles was undoubtedly the unintentional and random variation of manganese, silicon, and oxygen. A second obstacle was that the wide variation in yield strengths certainly influenced the CVN energies.

Following the regression analysis, a systematic, non-statistical evaluation of each element was conducted. Re-examination of the Phase 1 data (Table 2) revealed that CVN energy was dependent on the combination of carbon and nickel. In general, nickel was beneficial and carbon detrimental to CVN energy. At the -60°F test temperature, higher nickel levels resulted in higher absorbed energies. For a given nickel level, increasing carbon resulted in lower absorbed energies. The factor Ni+Ni/C was derived to reflect these trends, where Ni and C refer to the wt% of nickel and carbon. Figure 2 uses the factor to illustrate the effect of nickel and carbon on CVN energy at -60°F. When Ni+Ni/C was less than 50, low absorbed energies

were obtained. At Ni+Ni/C greater than 50, fairly good absorbed energies were achieved. The data also illustrate that increasing oxygen lowered the CVN energies.

At the 0°F test temperature, lower carbon resulted in higher CVN energies. At a given carbon level, increasing nickel improved impact toughness. The factor Ni/C-100C was derived to reflect this trend. Since nickel is present in much larger levels than carbon, the multiplication factor of 100 was included to ensure that the importance of carbon content was accurately reflected. This factor is used in Fig. 3. The graph shows that CVN energy steadily increases as Ni/C-100C increases. Acceptable absorbed energies were achieved when Ni/C-100C was greater than approximately 60. Two data points are significantly above the primary band of data. These points appear to follow the same trend, but are offset from the rest of the data. The reason for this offset is unknown.

The factors shown in Figs. 2 and 3 were not able to clearly explain the CVN energies of the Phase 2 formulations. That is, for Phase 2 formulations, Ni+Ni/C greater than 50 and Ni/C-100C greater than 60 did not ensure acceptable absorbed energies. This result indicated that there were other elements that were influencing CVN energy.

After additional evaluation, Mn:Si ratio was identified as the other factor that influenced the CVN energy. Figures 4 and 5 illustrate the effect of Mn:Si ratio on CVN energy at -60°F and 0°F. At both temperatures, optimum impact toughness was achieved at Mn:Si ratios of 8 to 9. Data reported by Evans [Evans, 1986], when re-evaluated in terms of Mn:Si ratio, shows that the highest CVN energies were achieved at a ratio of approximately 9. Court and Pollard [1987] concluded that the notch toughness of shielded metal arc weldments in C-Mn steel depended on the balance of Mn and Si. The metallurgical rationale for this trend is unclear. Widgery [1974, 1975] reported that Mn:Si ratio affects mechanical properties through its influence on inclusion content.

Figure 6 shows the combined influence of the Ni+Ni/C factor and Mn:Si ratio on

CVN energy at -60°F. Figure 7 shows the combined effect of the Ni/C-100C factor and Mn:Si ratio on CVN energy at 0°F. Regions of acceptable impact toughness are highlighted. These figures suggest that, over a wide range of yield strengths and oxygen contents, the impact toughness variation of the low C, Ni-Mo-Ti formulations was controlled by carbon, nickel, manganese, and silicon contents.

With the influence of composition on both the yield strength and impact toughness well defined, compositional ranges providing the optimum combination of strength and toughness were identified. The influence of individual elements on yield strength and CVN energy were evaluated simultaneously. This was accomplished by first setting molybdenum, and any two of the four elements found to control the impact toughness, at constant levels. CVN energies were then calculated as a function of the content of the remaining two elements. The CE(YS) equation presented in Fig. 1 was then used to determine yield strength as a function these elements.

This method is presented graphically in Fig. 8. In this example, carbon, molybdenum, and silicon were fixed at the levels shown. Iso-yield strength lines were developed over a range of manganese and nickel levels. The constant carbon and silicon contents, and the boundaries of acceptable impact toughness in Figs. 6 and 7 were then used to determine the combinations of manganese and nickel that provide acceptable CVN energies. The shaded area shows the range of manganese and nickel levels expected to give acceptable strength and impact toughness.

Using this method, the following range of weld metal compositions was estimated to provide acceptable yield strength and impact toughness: carbon at 0.04 to 0.06 wt%; manganese at 1.5 wt% to 1.7 wt%; nickel at 2.8 wt% to 3.3 wt%; molybdenum at 0.4 wt% to 0.6 wt%; and silicon at approximately 0.20 wt% to 0.25 wt%. The optimum titanium range was not identified due to the inability to control the titanium level in the experimental formulations.

FUTURE RESEARCH

As mentioned previously the solid wire verification phase is underway. Based on the results obtained in Phase 3, the optimum weld metal compositions identified above are being evaluated in solid wire form. The target titanium level was somewhat arbitrarily set at approximately 0.015 wt% to 0.02 wt%. The evaluation includes mechanical property testing at weld metal cooling rate extremes and weldability testing to evaluate cracking sensitivity.

LOW-CARBON BAINITIC WELD METALS

INTRODUCTION

The goal of this task is to develop a new gas metal arc welding (GMAW) wire for use in welding high strength steels. Currently used high strength weld metals were metallurgically designed to take advantage of the martensite microstructure which can be produced in carbon steel through the control of certain alloying elements such as manganese, chromium, and primarily carbon [Linnert, 1965]. This system can easily develop the high strength needed and can be modified to give good fracture toughness. The two main drawbacks of such a metallurgical design are that the formation of martensite requires high cooling rates and possesses high as-deposited hardness levels which are sensitive to hydrogen cracking. The cracking problem is usually controlled with the application of a high preheat temperature which drives off surface moisture and reduces the cooling rates. Thus, the cure for the cracking problem hampers the formation of the martensite. The reduced cooling rate can result in replacement of the martensite formation with bainite formation. The bainite structure of steel is less tough than the martensite structure when the carbon level of the steel is above about 0.04 wt% [Vassilaros *et al*, 1990; Irani and Tither, 1967].

The strength of bainite has been shown by Irvine and Pickering [1965] to be a function of the bainite transformation temperature. Unlike martensite, the bainite

transformation temperature often extends over a large range of cooling rates. Additionally, the measured transformation temperature for a given bainitic steel is relatively constant. Therefore, the formation of bainite occurs over a much larger range of cooling rates than martensite, which produces a microstructure less sensitive to cooling rate than materials designed to be martensitic. The toughness of bainite is comparable to martensite if the carbon level in the steel remains low. Low-carbon bainitic (LCB) steel plates have been commercially produced in the 60 ksi to 80 ksi yield strength range with excellent toughness [Nakasugi *et al*, 1980]. A 100+ ksi yield strength LCB steel developed by CDNSWC and University of Pittsburgh had a 0.02 wt% carbon level and toughness equivalent to HSLA-100 steel. The LCB steel system was further extended to develop an LCB-130 steel which provided the toughness of HY-130 steel with a carbon level of 0.02 wt% [Blicharski *et al*, 1988]. The success of these steels provided validity to the theory that a very low carbon steel could be developed with yield strengths greater than 130 ksi. However, these LCB steels had the advantage of advanced thermomechanical processing (TMP) during rolling which produced a fine grain size with improved toughness and strength. An LCB weld metal would not have the benefit of TMP to enhance strength and toughness.

The development of a high strength low carbon weld metal which is bainitic in structure requires a significant advancement of the state of the science. The main questions to be addressed are:

(1) Is a high strength LCB steel weld metal possible?

(2) Is the strength of such a steel less cooling rate sensitive than a martensitic steel with similar strength?

(3) Can the toughness of such a system be comparable to the toughness of a martensitic steel?

The LCB weld metal program will deal with these issues and attempt to develop a model to be used in weld metal design. The current results of the program described here will pertain to the three questions stated above.

EFFECTS OF ALLOYING ON STRENGTH

This program was designed to investigate effects of alloying and processing on generic low carbon steel compositions called "model materials." These model materials were initially designed using the knowledge resulting from the Navy programs which developed LCB-100 and LCB-130 steels. The analysis of these model materials was used to assess the effects of the major alloying elements such as manganese (1 wt% to 2 wt%), molybdenum (1.5 wt% to 5 wt%), nickel (2.5 wt% to 5 wt%), titanium (0.1 wt% to 0.25 wt%), and niobium (up to 0.1 wt%) on the strength, toughness, and cooling rate sensitivity of low carbon steel weld metal.

The effects of alloying elements on the strength of low carbon weld metals was obtained from two different welding procedures. For the first procedure, autogenous welds were produced in a series of low carbon steel model materials using gas tungsten arc welding (GTAW) with an argon - helium shielding gas. Two welding heat inputs (60 kJ/in and 120 kJ/in) were used to vary the cooling rate in the 3/4-in.-thick plate. The resulting weld was sectioned into flat tensile specimens and tested. The results of these tests are summarized in Table 7, which lists the measured strengthening of the alloying elements. A second series of model materials was welded using GTAW of grooved 3/4-in.-thick plates and filler wires of matching chemistries produced from the ends of the same plates. These weldments were sectioned to produce tensile specimens. The results of these tests are shown in Fig. 9. A list of the pertinent results are given as follows:

(1) The weld metal yield strength of low carbon steel can exceed 130 ksi with carbon levels of 0.02 wt%.

(2) The strengthening power of the alloying elements investigated in these low carbon steel weld metals was not predicted by published equations.

(3) The strengthening power of the alloying elements molybdenum and nickel were not a linear function of alloy content. These alloying elements appeared to interact with other alloying elements to produce an increase or decrease in

their effectiveness to increase strength for a given welding condition. This interaction made it difficult to design a low carbon steel weld metal from any linear equation based on only the chemistry of the weld deposit.

(4) Nickel alloying up to 3.5 wt% can be used to increase weld metal strength. This result allowed the addition of nickel for its traditional toughening effects to also aid in strengthening of the weld metal.

EFFECTS OF ALLOYING ON COOLING RATE SENSITIVITY

One of the postulated benefits of an LCB weld metal was the assumed cooling rate insensitivity. The evaluation of the effects of alloying elements on changes in weld metal strength with different cooling rates was investigated. Weld metal cooling rate changes were produced by varying travel speed and, therefore, heat input. The results of the first cooling rate study, performed on a 100 ksi yield strength low carbon weld metal, autogenously welded with heat inputs of 35 kJ/in to 120 kJ/in, are shown in Fig. 10. Also shown on this figure are some typical results for a MIL-120S [MIL-E-23765/2C, 1983] GMAW weld metal. A summary of the conclusions is as follows:

(1) Cooling rate sensitivity of weld metal strength was reduced with low carbon.

(2) Molybdenum and nickel were best at producing a strengthening effect which had the least cooling rate sensitivity.

(3) Increasing manganese, niobium, and carbon increased the strength and cooling rate sensitivity of the weld metal.

EFFECTS OF TITANIUM-BEARING INCLUSIONS ON TOUGHNESS

The investigation of the effects of titanium-nitride inclusions on the strength and toughness of LCB weld metals was performed on a series of low carbon weld metals with titanium levels of 0.012 wt%, 0.017 wt%, and 0.024 wt%. The three materials were grooved and GTA welded with matching filler metal. Tensile specimens machined

from the welds displayed a yield strength of approximately 120 ksi. The CVN 50% fracture appearance transition temperature (50% FATT) for CVN specimens machined from the welds is shown in Fig. 11.

Titanium levels greater than 0.012 wt% were found to increase the CVN 50% FATT without a significant change in strength. This indicated that the production of fine TiN inclusions, which have been shown to hinder grain growth and help toughness in weld heat affected zones (at low Ti levels) [Houghton *et al*, 1982; Kanazawa, 1976], was not enhanced with increased levels of titanium.

SUMMARY

The low-carbon bainitic weld metal system, with a carbon level of 0.02 wt%, has demonstrated the capability of obtaining yield strengths as high as 130 ksi. These chemistries have also demonstrated strengths which have a reduced cooling rate sensitivity compared to weld metals with a slightly higher carbon content (*e.g.* 0.04 wt% to 0.07 wt%). Although these results are promising, the toughness of such weld metals is still a significant question. The control and maximization of toughness in LCB weld metals will require further research before such materials can be fully developed.

FUTURE RESEARCH

Current research in the low-carbon bainitic weld metal program is focused on the metallurgical conditions that control the toughness of the weld metal. The relationship between chemistry and impact toughness is being investigated using the previously mentioned model materials to produce GTAW and GMAW welds which are evaluated for strength, ductility, impact toughness transition temperature, and upper shelf impact energy. The research will also address the effects of welding parameters such as heat input rate and cooling rate on the toughness of low-carbon bainitic weld metal. The formulation of a new welding consumable will require

knowledge of the chemistry/toughness relationship in addition to the strength/ chemistry relationships.

WELDABILITY METHODOLOGY

INTRODUCTION

In order to reduce the cost of fabricating naval structures the U.S. Navy is currently developing high strength low alloy (HSLA) steels which exhibit better weldability compared to the quenched and tempered HY steels. Significant fabrication cost savings were realized by substituting HSLA-80 for HY-80 in surface ship construction due to relaxation of welding controls such as minimum preheat temperature requirements. As the yield strength of the weldment system increases, the propensity for hydrogen cracking problems also increases. At the 100 ksi yield strength level, the results of weldability tests using HSLA-100 steel indicated that although the base plate exhibited good weldability, the weld metal was found to be the weak link in the weldment system. Improvement in the hydrogen cracking resistance of welding consumables for 100 ksi yield strength applications appears to be necessary in order to fully attain the cost saving afforded through the use of the highly weldable HSLA-100 steel.

Several mechanisms have been proposed to explain how hydrogen causes embrittlement and cracking [Zapffe and Sims, 1941; Louthan *et al*, 1972; Petch and Stables, 1952; Beachem 1972]. Despite different viewpoints on the exact mechanism used to describe hydrogen cracking, all of these models recognize that hydrogen cracking is caused by a combination of hydrogen, microstructure, and tensile stresses. The objective of this weldability program is to determine the effect of welding variables such as base plate and welding consumable composition, and weld metal diffusible hydrogen (H_D) content on resistance to hydrogen cracking.

Tensile stresses develop during welding due to welding shrinkage stresses that develop as each weld bead solidifies and cools. Satoh *et al* [1977] reported both

measured and calculated "restraint intensity" values for a variety of joint configurations and weldability test specimens based on mechanical factors such as plate size and thickness. Matsuda *et al* [1984] reported on the effect of the chemical composition on the development of restraint stress in a Y-groove-type weldability specimen (Fig. 12). They found a significant decrease in final restraint stress as the strength level of the weldment system was increased from a mild steel to a high strength steel system. This trend was attributed to "transformation expansion" due to lower temperature transformation products such as bainite or martensite. Transformation expansion at lower temperatures is believed to offset the welding shrinkage stresses.

TRANSFORMATION EXPANSION

Instrumented Y-groove tests were conducted in order to determine if this mechanism may be useful for increasing weld metal cracking resistance when welding HSLA-100 steel. Strain gages were placed on the flank of the specimen, as shown in Fig. 12. The total average strain, ϵ^*, was then calculated for each of the welding consumables used, as follows:

$$\epsilon^* = \epsilon_1 + \epsilon_2 - (\epsilon_3 + \epsilon_4) \quad (4)$$

where ϵ_1, ϵ_2, ϵ_3, and ϵ_4 are strain measurements from the strain gages in Fig. 12. The results shown in Fig. 13 indicate that there was a decrease in final restraint strain from -1930 μ for a MIL-70S [MIL-E-23765/1D, 1981] consumable to -1523 μ for a MIL-140S [MIL-E-24355B, 1992] consumable. Assuming linear elastic conditions, the restraint stress corresponding to these strains was found to decrease from 56.9 ksi for the MIL-70S product to 44.9 ksi for the MIL-140S product.

The decrease in strain was believed to be due to an increased volume fraction of lower-temperature transformation products (*e.g.*, martensite and bainite) when using the higher alloyed MIL-140S product compared to the MIL-70S product. Matsuda *et al* [1984] illustrated this phenomenon by conducting similar instrumented Y-groove

tests on a variety of materials ranging from mild steel to a 150 ksi yield strength steel. This phenomenon is important to understand, because tensile stresses are required to initiate hydrogen cracking. If tensile stresses (and accompanying strains) can be reduced, the likelihood of cracking may also be reduced.

This potential benefit of lowering the strain by using a higher alloy system is offset by the fact that the resulting lower temperature transformation products, such as martensite, are more susceptible to hydrogen cracking. The results of weldability tests discussed in the following section indicate that the higher alloyed weld metal was found to be more susceptible to hydrogen cracking compared to a lower alloyed deposit. This suggests that the increased hydrogen cracking susceptibility of the lower temperature transformation products that are formed when the alloying elements of the weld deposit are increased outweigh any slight improvement (reduction) in restraint stresses which may have occurred as a result of using a richer chemistry system.

WELDABILITY TESTS AND CRACKING MODELS

The Welding Institute of Canada rigidly restrained cracking (WIC) test [Chakravarti and Bala, 1989] was employed to evaluate weld metal cracking resistance under a variety of weld metal chemistry and welding preheat conditions. This test simulates a tack weld or root pass in a butt weld. Cracking in this test occurs parallel to the direction of welding. In general, the results of the weld cracking tests indicate that the hydrogen cracking susceptibility increased as the alloying and strength level of the welding consumables increased. These results suggest that the hydrogen cracking susceptibility of the higher alloy weld metal outweighs the slight reduction in final restraint stresses for higher strength consumables. Gapped bead on plate (GBoP) [Chakravarti and Bala, 1989] weldability tests, which promote weld cracking perpendicular to the direction of welding, exhibited results that were consistent with those of the WIC tests. Figure 14 shows a compilation of

WIC and GBoP results for several shielded metal arc welding (SMAW) consumables.

Several hydrogen cracking models have been proposed [Hart and Jones, 1982]. One feature common to these models is a cooling time parameter, such as $t_{100}cr$, or "critical time to cool to 100°C." This parameter is a measure of how quickly the weldment may cool and still avoid hydrogen cracking. The smaller the $t_{100}cr$ value for a given material system, the better its hydrogen cracking resistance. A small $t_{100}cr$ infers that the weld metal can cool rapidly and still avoid cracking. Welding conditions which result in faster cooling times include lower heat inputs and lower preheat temperatures. Consequently, a material which can be safely welded without a minimum preheat temperature requirement exhibits a smaller $t_{100}cr$ than one requiring a high preheat temperature in order to avoid cracking.

Weld metal P_{CM} [Ito and Bessyo, 1968] values were compared to t_{100} values measured for preheated and non-preheated WIC tests. The cracking results from the WIC tests shown in Fig. 14 demonstrate that, as the P_{CM} value increases, higher t_{100} values are required to avoid cracking. The minimum t_{100} value required to avoid cracking is called $t_{100}cr$. One estimation of $t_{100}cr$, based on Y-groove-type JIS-y tests were reported by Suzuki *et al* [1983]. In that study, $t_{100}cr$ for high restraint conditions was given by:

$$t_{100}cr = (1{,}145)*(P_{HA})^2 + (864)*(P_{HA}) - 171 \qquad (5)$$

where: $P_{HA} = \log(\lambda * H_D) + (11.9)*(P_{CM}) + (0.000089)*(R_{FY}) - 2.55$

$\lambda = 0.6$ for low hydrogen electrodes

H_D = weld metal diffusible hydrogen

R_{FY} = intensity of restraint.

For the high restraint WIC specimen, R_{FY} was approximated as 4000 kgf/mm*mm. Application of this model to the welding conditions investigated here produced the theoretical $t_{100}cr$ curve shown in Fig. 14. The predicted cracking/no cracking conditions were conservative compared to those observed for the WIC and GBoP tests. Use of the P_{HA} model in this case would result in higher and more costly preheats

than may actually be required.

DIFFUSIBLE HYDROGEN

The effect of H_D on weld metal cracking in WIC tests is currently being investigated using the GMAW process. H_D was varied using a nominal 98% argon - 2% oxygen shielding gas premixed with 0.3%, 0.6%, or 0.9% hydrogen. The initial results of H_D measurements using a gas chromatography technique [AWS A4.3-86, 1986] indicated that the H_D level can be varied from approximately 4 ml/100g for straight shielding gas to 16 ml/100g for 0.6% or 0.9% hydrogen additions (depending on the specific consumable employed). The results of WIC tests performed at ambient temperature (21°C) using HSLA-100 base plate and a MIL-100S-1 [MIL-E-23765/2C, 1983] welding consumable is provided in Table 8. The extent of cracking was estimated as the total combined crack length in three WIC specimens divided by the total test region length.

FUTURE RESEARCH

The results in Table 8 show that the combination of HSLA-100 base plate and MIL-100S-1 consumable result in a relatively low calculated P_{CM} value of 0.22 to 0.25. Under this low P_{CM} condition, weld metal cracking was not observed until H_D levels of over 7.5 ml/100g were imposed. Additional WIC tests are planned for different strength level consumables and HY-100 base plate to broaden the P_{CM} range. These studies will identify critical H_D levels under different P_{CM} conditions. The next phase of this study will combine the effect of critical H_D on cracking for a given P_{CM} condition, with the effect of P_{CM} on cracking, resulting in a cracking/no cracking response surface as a function of weld metal H_D and P_{CM} values (Fig. 15). This analysis will be used to recommend safe and rational welding parameters based on knowledge of the initial welding conditions.

WELDING FLUXES

INTRODUCTION

The U.S. Navy has a need for very close controls on the welding consumables and welding procedures used to fabricate its high strength steels at the 100+ ksi yield strength level. The pedigree of the materials being used plays a large role in the welding conditions that can be used to deposit weld metal that performs in a satisfactory manner.

The primary objective of this task is for the Navy to gain a more fundamental understanding of the welding fluxes used in the construction of its surface ships and submarines. The Navy has traditionally known very little about these materials, relying on the manufacturer to provide the correct product and performing a series of welding tests to confirm that satisfactory properties can be achieved. The technology base established by this work will provide a means of verifying flux integrity for quality conformance and assurance, provide a possible means of making more appropriate flux selection for future submerged arc welding (SAW) applications based on the knowledge gained, and, possibly, allow for the design and production of optimized fluxes for special applications.

FLUX COMPOSITION

Commercial fluxes, all classified as high basicity, agglomerated fluxes, were characterized by several analytical techniques to determine their compositions. Unlike the low alloy steels used by the Navy, very little is known about the composition of the fluxes used to weld them. Therefore, rather than requesting the amounts of specific elements present (*e.g.*, those known to be used for alloying in steels, with the remainder being iron), it was necessary to request an analysis for as many elements as each technique could measure. A second consideration was that fluxes are composed mainly of minerals of the earth and would not simply be an alloy of several elements, but a mixture of complex constituents. Flux compositions were

determined by optical emission spectroscopy (OES), x-ray fluorescence (XRF), and inductively coupled plasma (ICP), which are described elsewhere [Siewert and Franke, 1990]. OES was able to provide relative concentrations, on a semi-quantitative basis, for 35 elements; XRF provided results for ten elements; and ICP provided analysis for 44 elements. The selective ion electrode technique [ASTM C698-80a, 1980] was used to determine fluorine since it was known that, as high basicity fluxes for welding high strength steels, these fluxes would contain a significant amount of CaF_2 for weld metal oxygen control. XRF data, originally reported as simple oxides, was recalculated to include the F data, assuming the F was present as CaF_2. When consolidated into a single list, these techniques were able to provide data for 46 different elements in the flux samples, providing a "fingerprint" for each manufacturer's flux. A partial list of flux compositions is presented as simple compounds of oxygen and fluorine in Table 9.

The fluxes are all similar in the content of their major constituents. This is understandable, as they all come from the *same* class of fluxes. Trace amounts of constituents can be identified, as well. This provides the "fingerprint" of each flux. MnO, TiO_2, and Fe_2O_3 are found to vary markedly from flux to flux. The amounts and ratios of the binders (K_2O, Na_2O) also vary. The analysis also identified a higher concentration of rare earths (Ce, La, Nd) in flux C2 than in the other fluxes, indicating that they are probably intentionally added. The use of rare earth additions to modify weld metal inclusions has been shown to reduce H_D, refine non-metallic inclusions, and improve weldability [Musiyachenko *et al*, 1988]. It is possible the manufacturer of flux C2 considered this attribute in formulating this flux. The basicity index (BI) [Tuliani *et al*, 1969] covers a relatively large range within this class of fluxes.

SUBMERGED ARC WELDMENTS

The fluxes shown in Table 9 were then used to produce a series of SAW weldments

to discern if this information could be used to better understand the performance of weld metals produced with them. One-in.-thick weldments were produced with HY-100 plate [MIL-S-16216J, Amend. 1; 1982] and a MIL-120S [MIL-E-23765/2C, 1983] welding electrode. Welding parameters were held as constant as possible to achieve a heat input of 55 kJ/in and preheat and interpass temperatures between 250°F and 275°F, so that the only variable in the weldments was the flux.

The resulting weld metal chemistry is presented in Table 10. Figure 16 shows the correlation between flux MnO and weld metal Mn. Since Mn is constant in the welding consumable and the plate, interactions between the weld pool and the molten slag account for the changing weld Mn levels. Mn increases as MnO increases. These results are in agreement with those of Chai and Eagar [1981], which identified a correlation between flux MnO content, BI, and the resulting weld metal Mn. Figure 17 shows how the flux CaF_2 content affects the resulting weld metal oxygen content. Increased amounts of CaF_2 reduce the level of oxygen in the weld metal.

Table 11 presents the weld metal properties achieved with these fluxes. While all tensile yield strength values are at least 100 ksi, flux F2 produced weld metal much stronger than the rest. Flux F2 exhibits a high CaF_2 and MnO content, and a high BI. The resulting weld metal also has a high P_{CM} and low oxygen.

Figures 18 and 19 show CVN performance as a function of weld metal Mn and oxygen, respectively. Evans [1980] reported similar Mn performance for C-Mn weld metals. Both elements appear to exhibit an optimum value for the best CVN performance at a given temperature. In combination with the previous figures, these show the correlation between flux composition, resulting weld metal composition, and the weld metal performance.

SUMMARY

This initial phase of work has indicated much can be learned from a more fundamental knowledge of the welding fluxes the Navy uses to weld its high strength

steels. From flux composition the philosophy, or strategy, of the flux manufacturer can be hypothesized. Correlations can be drawn between flux composition, weld metal chemistry, and weld properties. This knowledge of the Navy's welding fluxes can lead to better selection of the appropriate flux, or possibly the design of optimized fluxes, for specific welding applications.

FUTURE RESEARCH

Work is proceeding to determine the most appropriate flux composition for welding the Navy's high strength steels. One drawback of this study pertained to the use of commercial welding fluxes from several manufacturers. While performance of actual materials the Navy uses was of great help, correlations for certain flux constituents were most likely confounded due to their interactions with other constituents that changed for each flux, possibly masking the full effects of those constituents. The next phase will examine a series of fluxes in which only MnO content is varied from about 0.5 wt% to 3.5 wt%. Performance of other constituents is also desired, but MnO content appeared to play a significant role in determining weld metal Mn content and weld metal performance. Future series will examine other flux constituents as well as synergic effects for several constituents.

SUMMARY AND CONCLUSIONS

With regard to the overall goals of the program the following summarizes the current status of each of the four programs described and provides direction and emphasis of future work.

The HSLA-100 Welding Consumables Development program used a statistically designed experiment to formulate and develop a wire for gas metal arc welding of HSLA-100 steel. Electrode formulations were based on low carbon, Ni-Mo-Ti systems with yield strength goals of 102 ksi to 122 ksi and minimum CVN toughness goals of 45 ft-lb at -60°F and 60 ft-lb at 0°F. Regression analyses and systematic, non-

statistical evaluations were conducted to determine the effects of elements and interactions of elements on the strength and toughness of the deposited weld metal. Regression analysis was used to analyze the effects of elements on the yield strength of the weld metals. The results of the regression indicate that C, Mo, Mn, and Ni are the strengtheners, with the results being consistent with literature. Regression analysis did not explain the results of the impact toughness tests; thus, systematic, non-statistical evaluations were conducted and indicate that, over a wide range of yield strengths and oxygen contents, the impact toughness variation of the low C, Ni-Mo-Ti formulations was controlled by carbon, manganese, and silicon contents. Once the influence of composition on both the yield strength and toughness were defined, compositional ranges providing the optimum combination of strength and toughness were identified as: carbon at 0.04 wt% to 0.06 wt%; manganese at 1.5 wt% to 1.7 wt%; nickel at 2.8 wt% to 3.3 wt%; molybdenum at 0.4 wt% to 0.6 wt%; and silicon at approximately 0.20 wt% to 0.25 wt%.

For the Low-Carbon Bainitic Weld Metals program, "model materials" were used. Both autogenous and cold wire gas tungsten arc welding were used at 60 kJ/in and 120 kJ/in to identify the potency of alloying elements on the yield strength, cooling rate sensitivity, and impact toughness of the resulting weld metals. The results indicate that the yield strength of low carbon (<0.02 wt%) steels can exceed 130 ksi; that the strengthening power of the alloying elements was not predicted by the literature; that the strengthening power of Mo and Ni were not linear functions of the alloy content; and that Ni could be added up to 3.5 wt% to increase weld metal strength as well as toughness. The cooling rate sensitivity of the weld metal was identified to decrease with low carbon levels; Mo and Ni strengthened the weld metal with the least effect on cooling rate sensitivity; and increases in Ni, Nb, and C increased the strength and cooling rate sensitivity of the weld metal. The toughness of these weld metals was lower than desired, requiring additional research.

The goal of the Weldability Methodology program is to develop a response surface which provides a boundary between crack-sensitive and non-crack-sensitive regions of a three dimensional space consisting of weld metal cooling time, diffusible hydrogen, and alloy content of the weld metal. In order to achieve this predictive capability, the results of a number of weldability tests were evaluated to identify weld metal cracking resistance for a variety of weld metal chemistry and welding preheat conditions. The Welding Institute of Canada (WIC) and Gapped Bead on Plate (GBoP) tests were chosen to identify the susceptibility of the weld metals to hydrogen cracking in the longitudinal and transverse directions, respectively. The weld metal alloy content was represented by P_{CM} and the measure of ease with which hydrogen can diffuse from the weld metal was represented by t_{100}; the diffusible hydrogen measurements, H_D, were made using a gas chromatography technique. The results indicate that for the combination of HSLA-100 base plate and MIL-100S-1 consumable chosen, weld metal cracking was not observed until H_D levels of 7.5 ml/100g were imposed.

The primary objective of the Welding Fluxes program was to gain a more fundamental understanding of these materials. Techniques were developed to determine the composition of the fluxes: optical emission spectroscopy to provide relative concentrations for 35 elements; x-ray fluorescence for ten elements; and inductively coupled plasma for 44 elements. In addition to these techniques, the selective ion electrode technique was used to determine fluorine content, since significant amounts of fluorine were present in these fluxes to control weld metal oxygen content. The combination of these techniques provided data for 46 different elements in the flux, thus, providing a fingerprint for each flux. The results of these analyses indicate that the MnO, TiO_2, and Fe_2O_3 varied markedly from flux to flux. Additionally, using these techniques, apparent intentional additions of rare earth elements Ce, La, and Nd were found in one flux, possibly to reduce the amount of diffusible hydrogen, refine non-metallic inclusions, and improve weldability. A

series of submerged arc welds was produced to identify the performance of weld metal produced using the various fluxes. Results indicate that weld metal manganese content correlates with flux MnO content, and that weld metal strength is a function of weld metal Mn content. The weld metal toughness is directly affected by the weld metal oxygen content, which is reduced with increasing amounts of CaF_2 content in the flux. Thus, correlations between flux composition, weld metal chemistry, and weld properties can be drawn, and a design of an optimum flux for Naval use may be achieved.

Work will continue in all four of the cited programs. The HSLA-100 Welding Consumables Development program will advance to evaluating weld metal produced by GMAW using solid wires based on the target chemistry identified in this report. The Low-Carbon Bainitic Weld Metals program will focus on increasing the impact toughness of candidate materials which have demonstrated adequate strength levels. The Weldability Methodology program will be used as an underlying methodology to assess the weldability of all candidate consumables. The Welding Fluxes program will systematically vary the flux MnO content in order to optimize weld metal yield strength while maintaining weld metal impact toughness.

Efforts to rigorously understand the metallurgical aspects of the weld metals produced in these programs will include fundamental evaluations of titanium bearing inclusions. Finally, solid state phase transformations, as well as inclusion and phase stability, associated with the relatively rapid cooling rates and reheating which occur during multi-pass fusion welding will be investigated as a means to maintain strength while attempting to improve toughness.

Table 1. Phase 1 target weld metal compositions (wt%).

Formulation	C	Ni	Mo	Ti	Mn
1	0.03	2.0	0.4	0.015	1.5
2	0.03	2.0	0.4	0.030	1.5
3	0.03	2.0	0.6	0.015	1.5
4	0.03	2.0	0.6	0.030	1.5
5	0.03	3.0	0.4	0.015	1.5
6	0.03	3.0	0.4	0.030	1.5
7	0.03	3.0	0.6	0.015	1.5
8	0.03	3.0	0.6	0.030	1.5
9	0.06	2.0	0.4	0.015	1.5
10	0.06	2.0	0.4	0.030	1.5
11	0.06	2.0	0.6	0.015	1.5
12	0.06	2.0	0.6	0.030	1.5
13	0.06	3.0	0.4	0.015	1.5
14	0.06	3.0	0.4	0.030	1.5
15	0.06	3.0	0.6	0.015	1.5
16	0.06	3.0	0.6	0.030	1.5

Table 2. Weld metal compositions (wt%) and mechanical properties from Phase 1.

Formulation	C	Ni	Mo	Ti	Mn	Si	O	N	Yield Strength (ksi)	CVN at 0°F (ft-lb)	CVN at -60°F (ft-lb)
1	0.040	2.18	0.42	0.011	1.53	0.19	0.041	0.014	90	88	50
2	0.036	2.14	0.44	0.018	1.41	0.14	0.041	0.010	83	57	45
3	0.040	2.31	0.64	0.012	1.49	0.21	0.043	0.008	94	63	41
4	0.035	2.20	0.64	0.017	1.42	0.14	0.068	0.013	90	66	34
5	0.040	3.26	0.44	0.011	1.54	0.20	0.042	0.008	102	65	51
6	0.037	3.12	0.45	0.017	1.50	0.14	0.041	0.012	93	79	47
7	0.040	3.23	0.63	0.011	1.50	0.21	0.050	0.011	104	69	43
8	0.035	3.18	0.63	0.022	1.53	0.18	0.042	0.013	101	69	49
9	0.050	2.13	0.44	0.012	1.52	0.19	0.041	0.008	98	55	33
10	0.059	2.25	0.46	0.013	1.36	0.09	0.053	0.011	87	50	29
11	0.060	2.35	0.65	0.012	1.55	0.20	0.042	0.007	105	45	17
12	0.060	2.21	0.65	0.023	1.53	0.18	0.033	0.011	102	67	40
13	0.060	3.15	0.44	0.014	1.50	0.20	0.041	0.009	107	55	43
14	0.060	3.12	0.45	0.022	1.57	0.18	0.029	0.010	103	62	54
15	0.060	3.15	0.63	0.014	1.52	0.20	0.041	0.008	111	64	53
16	0.066	3.11	0.63	0.018	1.40	0.15	0.034	0.010	106	60	50

Table 3. Regression equations developed from Phase 1 data [Wong and Hayes, 1990].

Regression Equation (elements are in wt%)	Regression Coefficient (R^2)	
Yield Strength (ksi) = 88.86 + 387(C) - 3249(Ti) + 634(Ni)(Ti) + 1763(Mo)(Ti)	0.915	(1)
CVN at 0°F (ft-lb) = 346 - 4020(C) - 210(Mo) - 156(Ti:N) + 152(C)(Ni) + 1831(C)(Ti:N) + 127(Mo)(Ti:N)	0.914	(2)
CVN at -60°F (ft-lb) = 381 - 4886(C) - 49(Ni) - 323(Mo) - 107(Ti:N) + 829(C)(Ni) + 1281(C)(Ti:N) + 46(Ni)(Mo) + 114(Mo)(Ti:N)	0.917	(3)

Table 4. Phase 2 target weld metal compositions (wt%).

Formulation	C	Ni	Mo	Ti	Mn	Ti:N
17	0.045	2.5	0.5	0.023	1.50	1-3
18	0.045	2.5	0.5	0.023	1.25	1-3
19	0.045	2.5	0.5	0.023	1.75	1-3
20	0.060	4.0	0.6	0.023	1.50	1-3
21	0.060	3.0	0.6	0.025	1.50	2.0
22	0.060	3.0	0.6	0.035	1.50	2.8
23	0.060	3.0	0.6	0.045	1.50	3.6
24	0.072	3.0	0.4	0.024	1.50	2.0
25	0.035	7.0	0.6	0.023	1.50	1-3

Table 5. Weld metal compositions (wt%) and mechanical properties from Phase 2.

Formulation	C	Ni	Mo	Ti	Mn	Ti:N	Si	O	Yield Strength (ksi)	CVN at 0°F (ft-lb)	CVN at -60°F (ft-lb)
17	0.048	2.57	0.50	0.018	1.52	2.25	0.23	0.042	99	51	38
18	0.044	2.68	0.51	0.015	1.27	1.25	0.22	0.039	99	38	25
19	0.049	2.50	0.49	0.021	1.70	2.10	0.23	0.020	108	60	47
20	0.057	3.90	0.62	0.016	1.50	1.78	0.22	0.042	122	33	27
21	0.064	3.15	0.61	0.023	1.57	2.55	0.25	0.036	118	48	39
22	0.058	3.06	0.61	0.022	1.52	3.14	0.24	0.038	110	46	39
23	0.057	2.99	0.59	0.023	1.48	2.30	0.23	0.043	111	48	36
24	0.065	3.30	0.46	0.015	1.39	1.67	0.25	0.023	112	48	40
25	0.033	6.90	0.58	0.018	1.46	2.57	0.24	0.049	129	24	21

Table 6. Comparison between calculated and measured mechanical properties for Phase 3 formulations.

Formulation	Yield Strength (ksi)		CVN at 0°F (ft-lb)		CVN at -60°F (ft-lb)	
	Calc	Meas	Calc	Meas	Calc	Meas
17	94	99	56	51	46	38
18	96	99	67	38	40	25
19	91	108	56	60	43	47
20	116	122	69	33	72	27
21	110	118	90	48	86	39
22	106	110	99	46	99	39
23	104	111	72	48	64	36
24	109	112	57	48	49	40
25	140	129	70	24	71	21

Table 7. Change in weld metal yield strength (ksi) per percent alloying.

	Element Addition															
	<3.5 wt% Ni		>3.5 wt% Ni		Mo		C		<1.5 wt% Mn		>1.5 wt% Mn		<0.045 wt% Nb		>0.045 wt% Nb	
Heat Input (kJ/in)	60	120	60	120	60	120	60	120	60	120	60	120	60	120	60	120
at low Mo (2.5 wt%)	6.25	4.17	-2.3	-1.1	---	---	76.9	230	22.5	18.4			261	130		
at high Mo (3.5 wt%)	4.52	9.03	4.52	9.03	---	---	875	625	30.6	21.3	1.3	-1.3	267	133	467	333
at low Ni (2.5 wt%)	---	---	---	---	4.72	4.72										
at high Ni (3.5 wt%)	---	---	---	---	-1	2							140	70		
at high Mn (2.0 wt%)					-3.1	-2.1			---	---	---	---				
Heuschkel [1964]	3.9	3.9	3.9	3.9	17.6	17.6	115	115	15.3	15.3	15.3	15.3	---	---	---	---
Pickering [1967]	1.1	1.1	1.1	1.1	17.4	17.4	247	247	12	12	12	12	168	168	168	168

Table 8. Results of WIC weldability tests performed at ambient temperature for HSLA-100 steel and a MIL-100S welding consumable.

Hydrogen in Shielding Gas, %	H_D ml/100g	P_{CM}	Extent of Cracking, %
0	4.7	0.23	0
0.3	7.5	0.22	0
0.6	10.0	0.25	96
0.9	14.7	0.25	100

Table 9. Submerged arc welding flux composition (wt%).

Compound	Flux A7	Flux E2	Flux C2	Flux F2	Flux H2
SiO_2	14.7	13.3	16.5	14.5	15.6
MgO	28.9	32.5	28.3	32.1	35.0
CaO	10.2	11.6	17.3	9.9	8.1
CaF_2	22.2	22.2	14.0	20.9	17.2
Al_2O_3	18.2	14.1	13.1	15.4	16.9
Fe_2O_3	1.09	1.09	1.19	1.05	2.00
TiO_2	0.72	-	-	0.52	0.68
MnO	0.89	0.45	0.05	1.60	2.52
K_2O	0.99	1.08	1.24	1.34	0.39
Na_2O	0.94	1.01	1.99	1.20	1.23
BaO	0.050	0.017	0.027	0.009	0.035
SrO	0.009	0.013	0.014	0.006	0.025
CeO_2	0.004	-	0.326	0.006	0.014
La_2O_3	0.006	0.002	0.269	0.007	0.009
Nd_2O_3	-	-	0.093	-	0.007
ZrO_2	0.04	0.03	-	0.08	0.04
BI	2.66	3.40	2.75	2.97	2.63

Table 10. Chemistry (wt%) of welding consumable and submerged arc weld metal.

Element	Wire	Flux A7	Flux E2	Flux C2	Flux F2	Flux H2
C	0.081	0.062	0.063	0.056	0.064	0.064
Mn	1.50	1.45	1.33	1.28	1.54	1.51
Si	0.40	0.38	0.28	0.42	0.34	0.28
Ni	2.25	2.33	2.33	2.51	2.34	2.34
Mo	0.42	0.46	0.45	0.47	0.49	0.47
Cr	0.28	0.25	0.41	0.40	0.50	0.47
Ti	0.014	0.006	0.004	0.004	0.006	0.005
Cu	0.011	0.020	0.020	0.017	0.026	0.021
Al	0.012	0.013	0.012	0.011	0.014	0.011
V	0.001	0.003	0.002	0.002	0.003	0.004
B	0.004	0.004	0.004	0.004	0.001	0.004
O	0.003	0.030	0.032	0.034	0.032	0.035
P_{CM}	0.269	0.250	0.249	0.248	0.256	0.264

Table 11. Properties for submerged arc weld metals.

Property	Flux A7	Flux E2	Flux C2	Flux F2	Flux H2
YS (ksi)	101	100	100	120	102
CVN (avg, ft-lb)					
@ -60°F	71	73	49	66	44
@ 0°F	91	101	86	87	63
Upper Shelf	106	130	120	92	94
50% FATT (°F)	-65	-75	-49	-99	-52

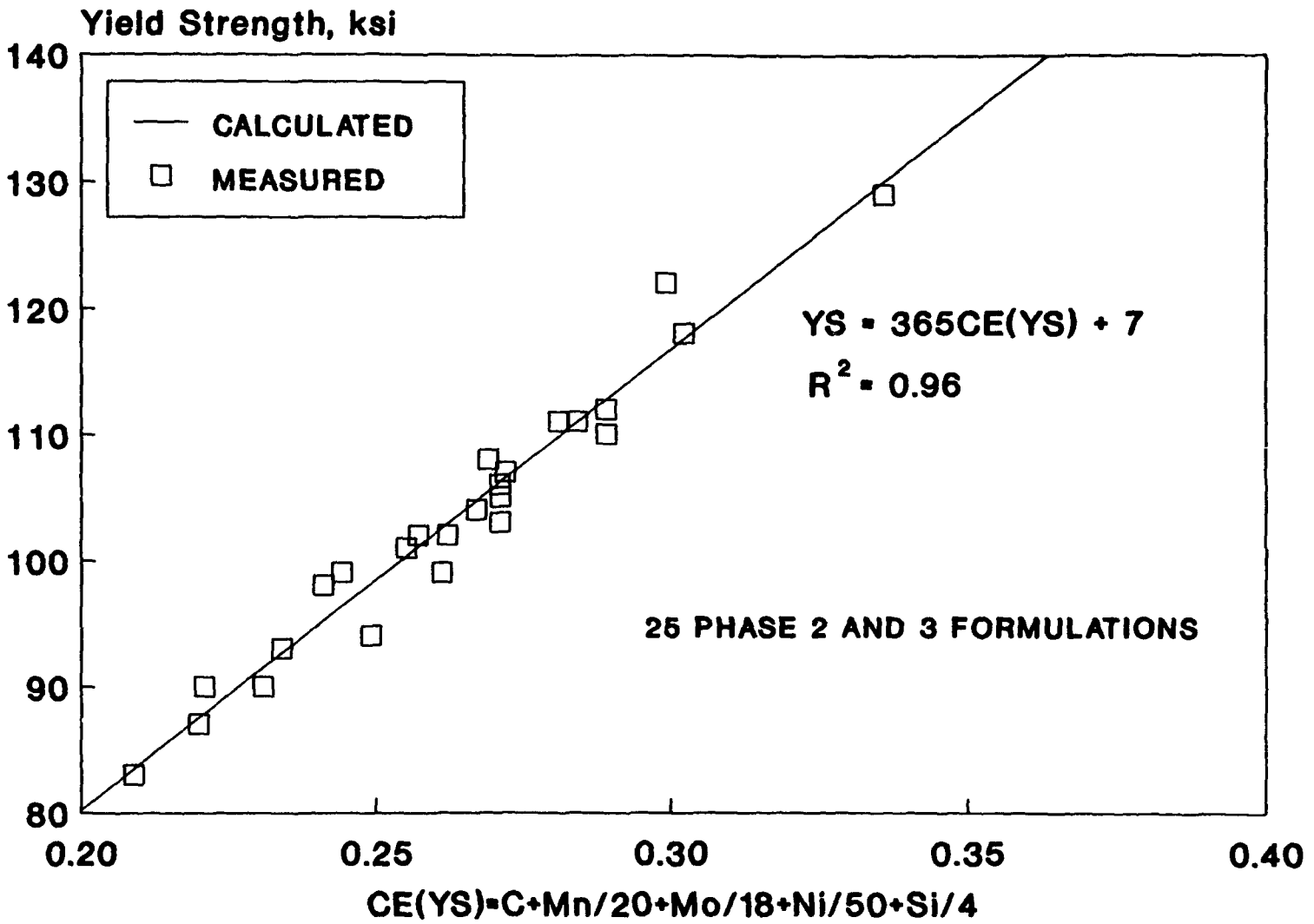

Fig. 1. Effect of composition on yield strength.

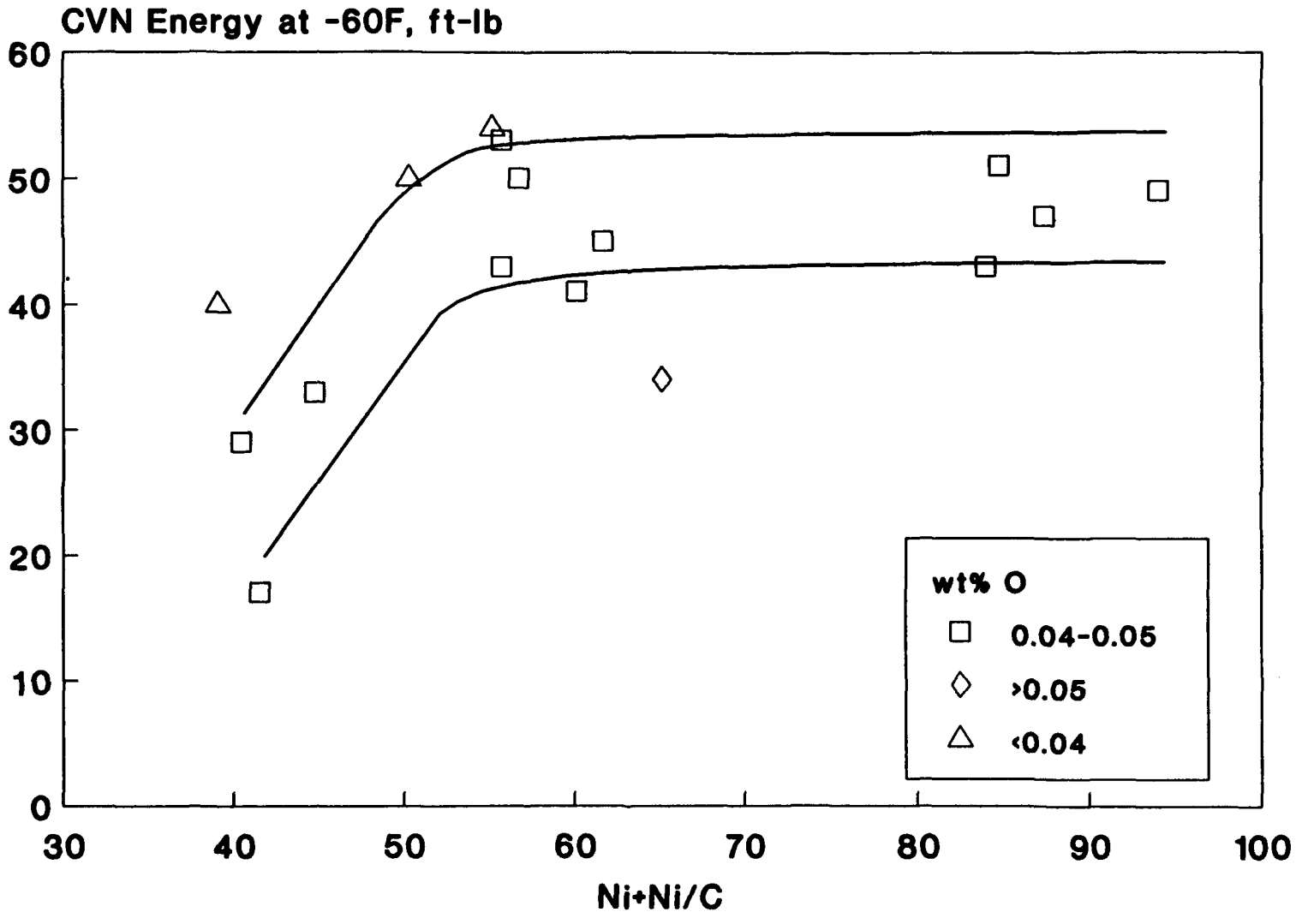

Fig. 2 Effect of carbon and nickel on CVN energy at -60°F.

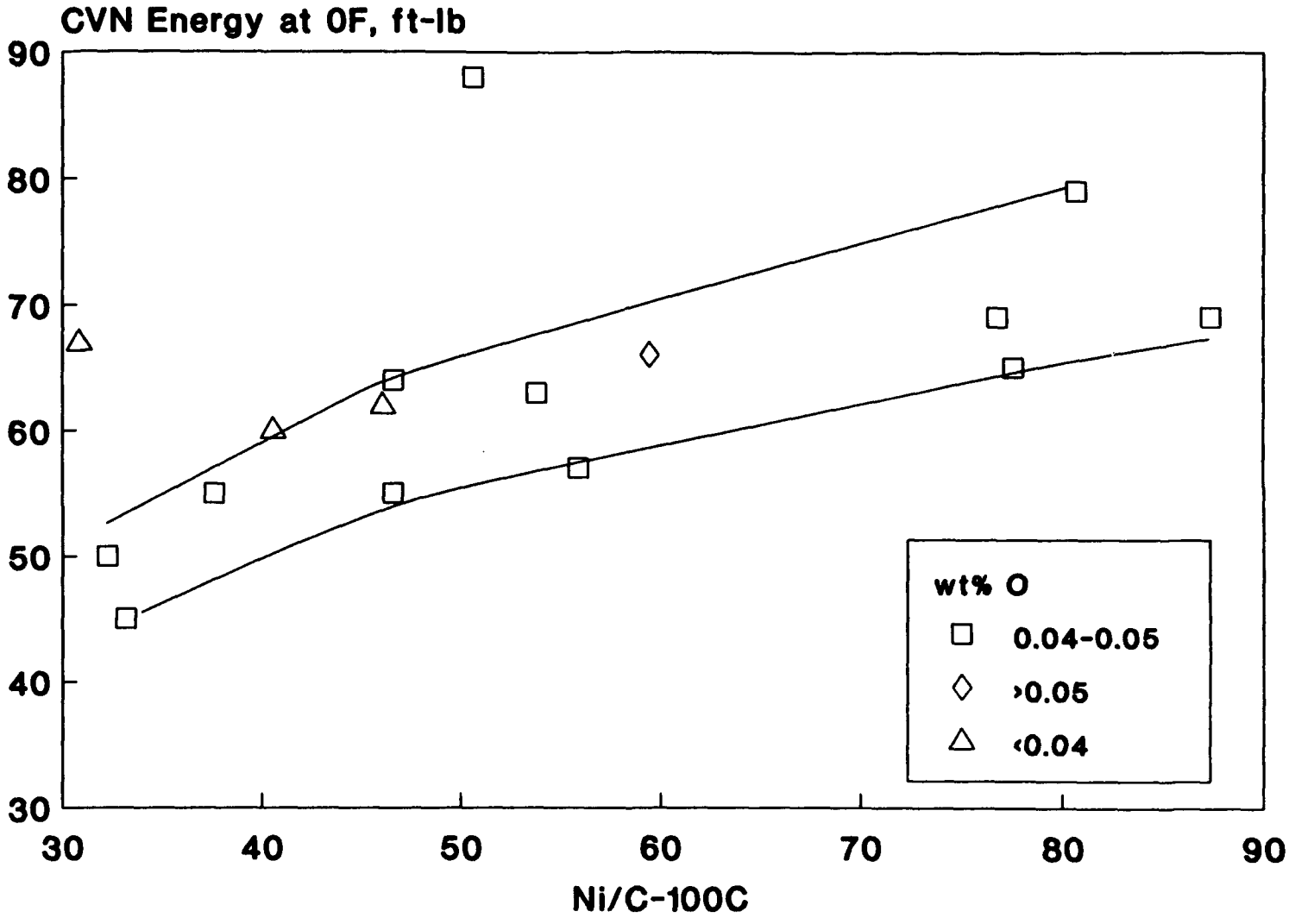

Fig. 3. Effect of carbon and nickel on CVN energy at 0°F.

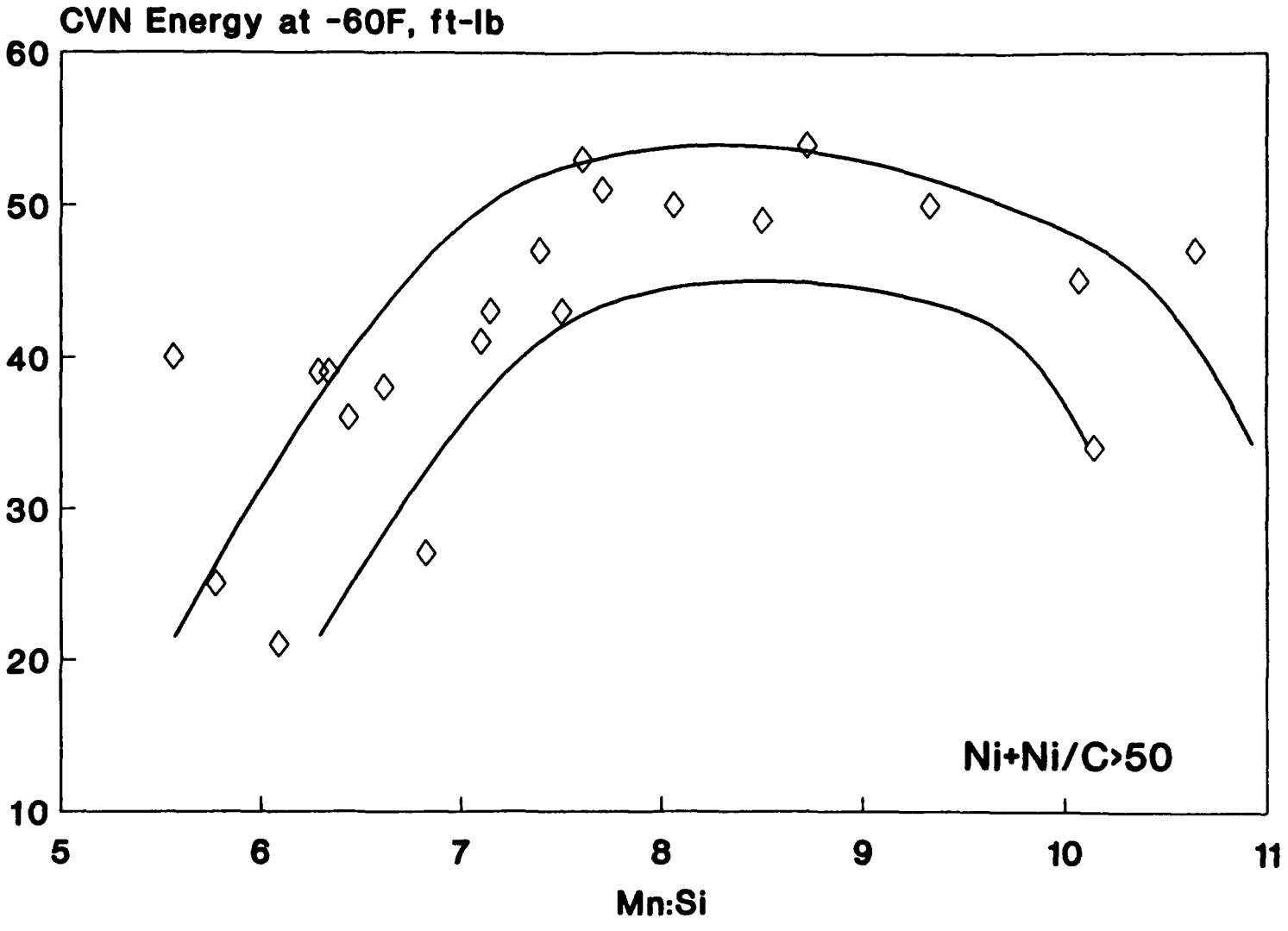

Fig. 4. Effect of manganese and silicon on CVN energy at -60°F.

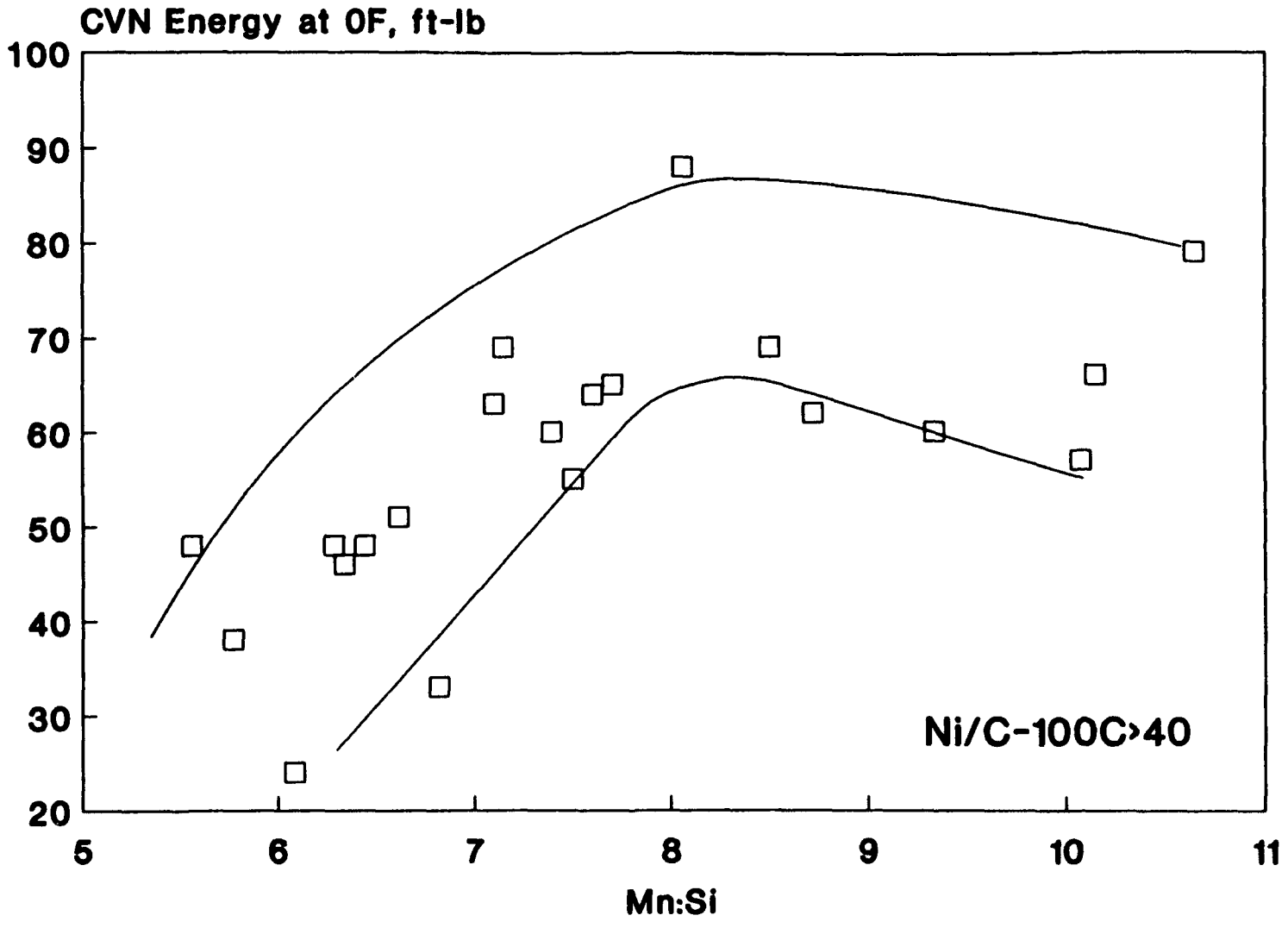

Fig. 5. Effect of manganese and silicon on CVN energy at 0°F.

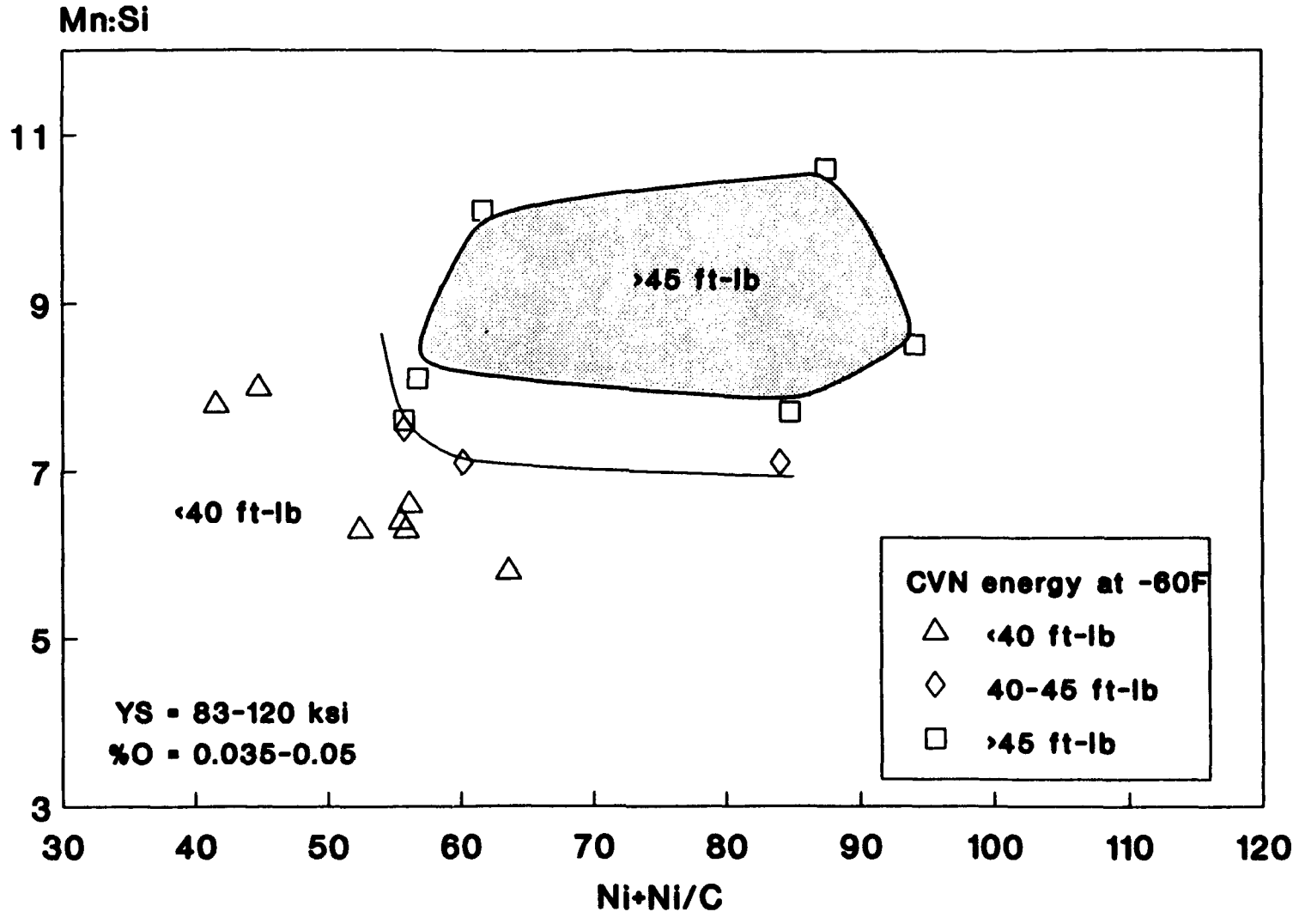

Fig. 6. Effect of composition on CVN energy at -60°F.

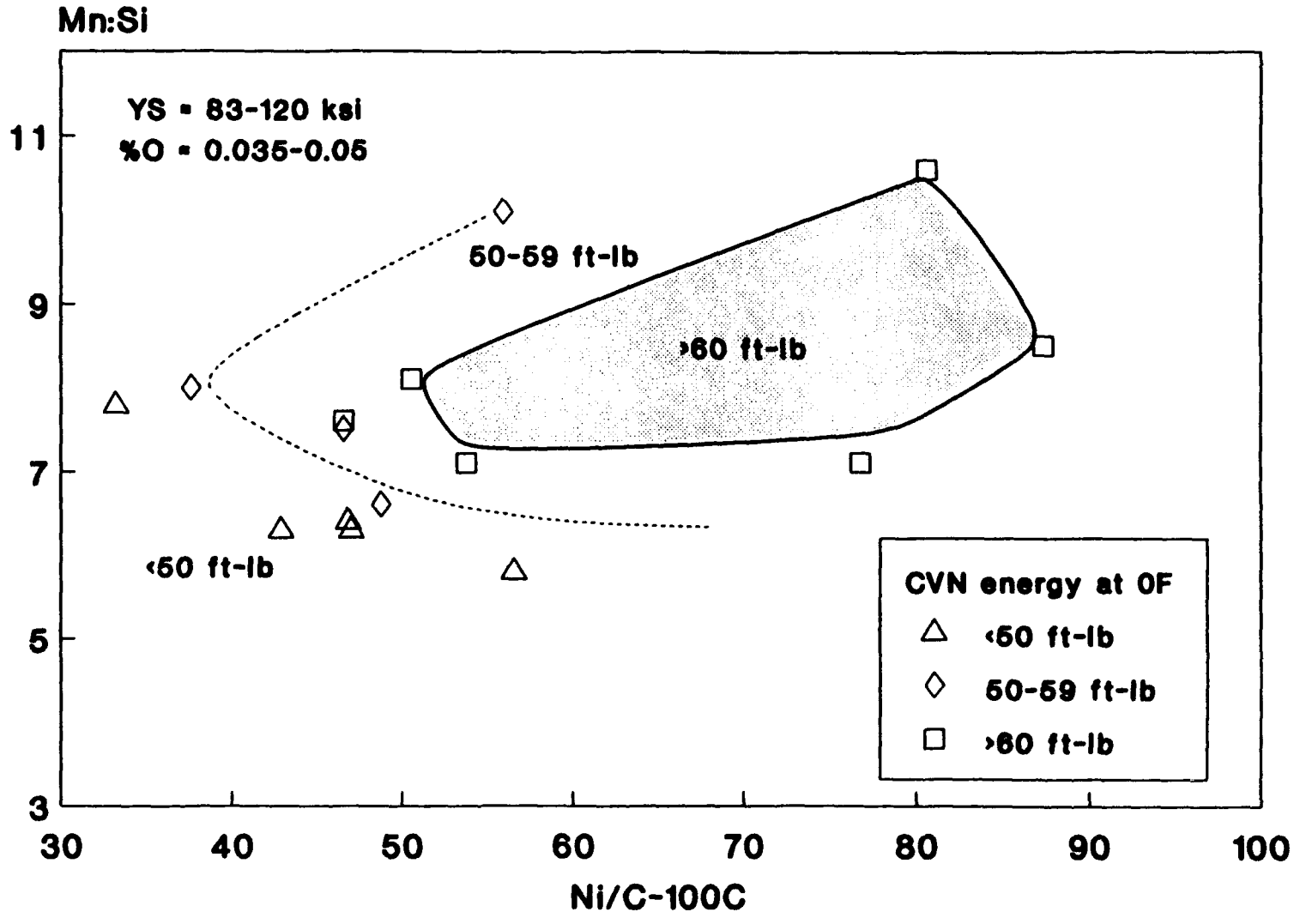

Fig. 7. Effect of composition on CVN energy at 0°F.

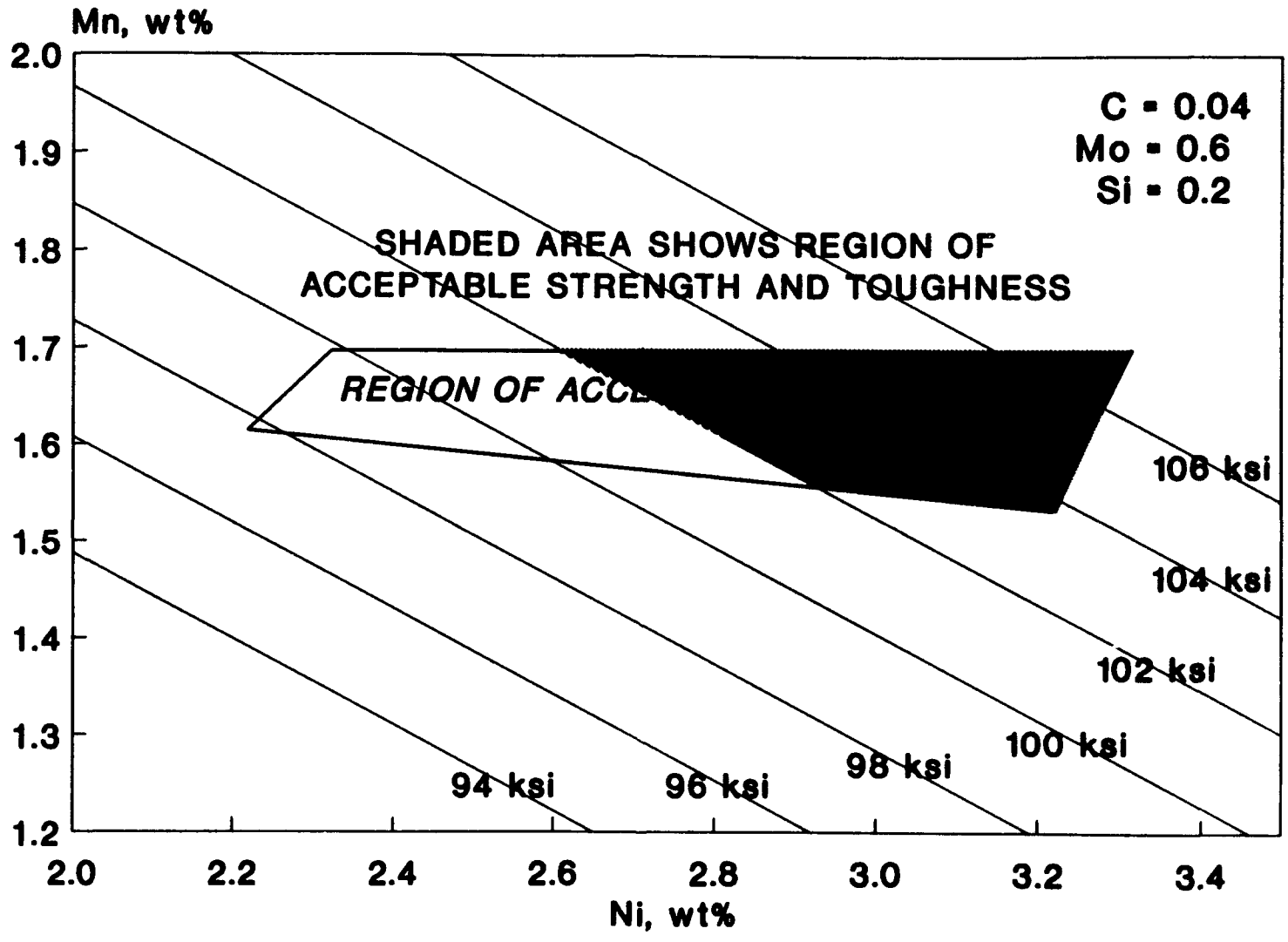

Fig. 8. Example of graphical method of identifying the compositional range that provides acceptable mechanical properties.

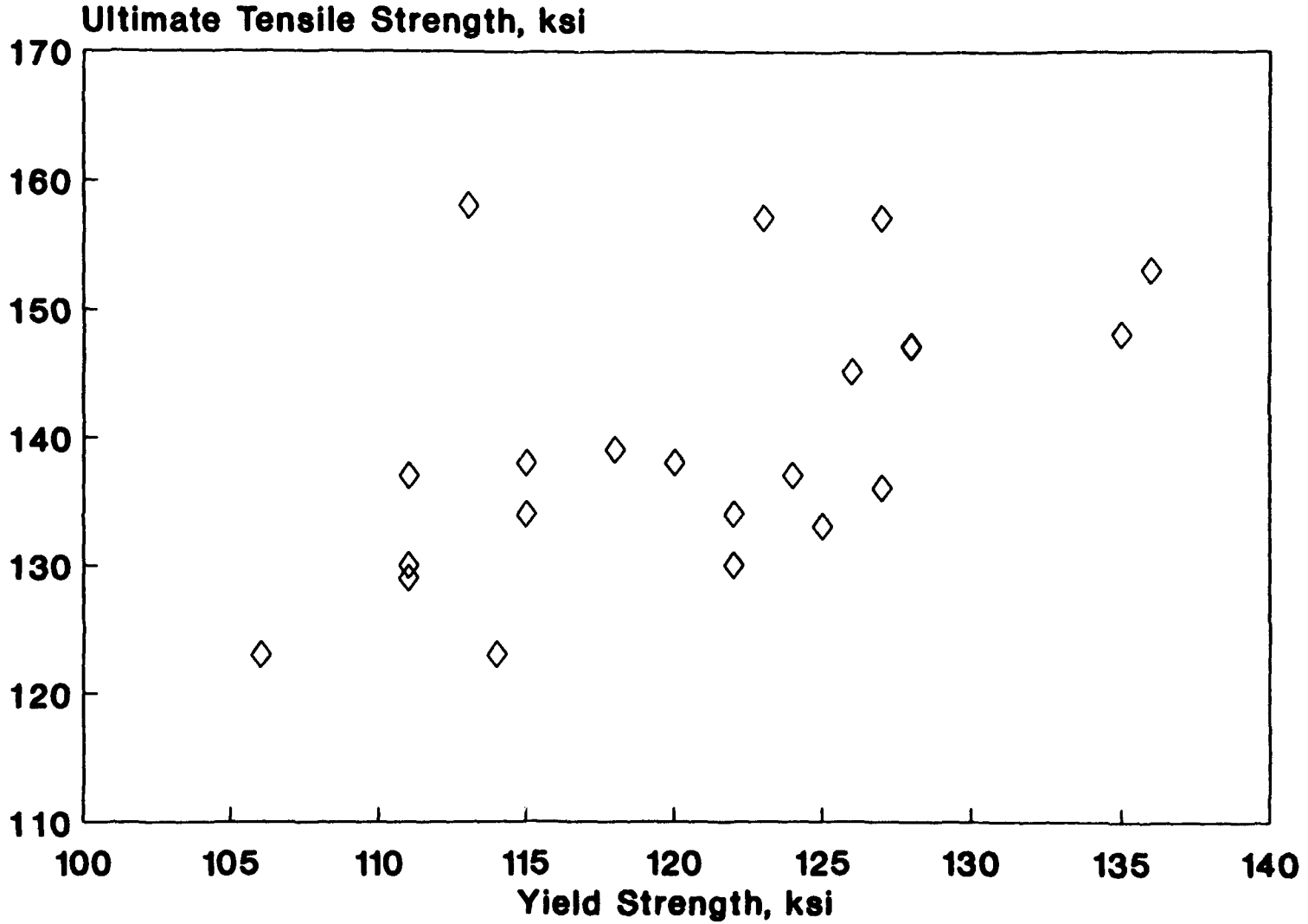

Fig. 9. Tensile test results for LCB weld metals.

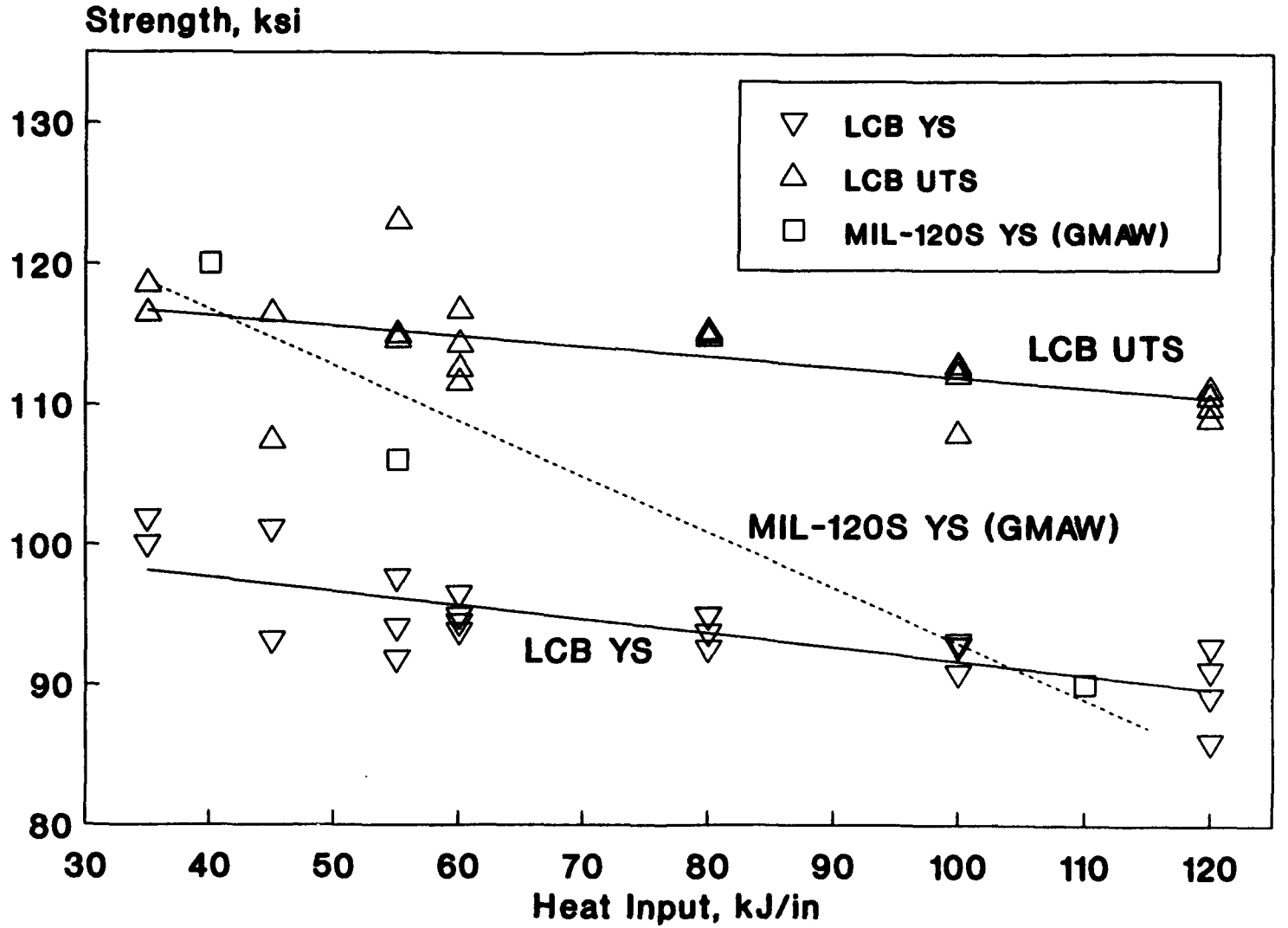

Fig. 10. Effect of welding heat input on LCB weld metal strength.

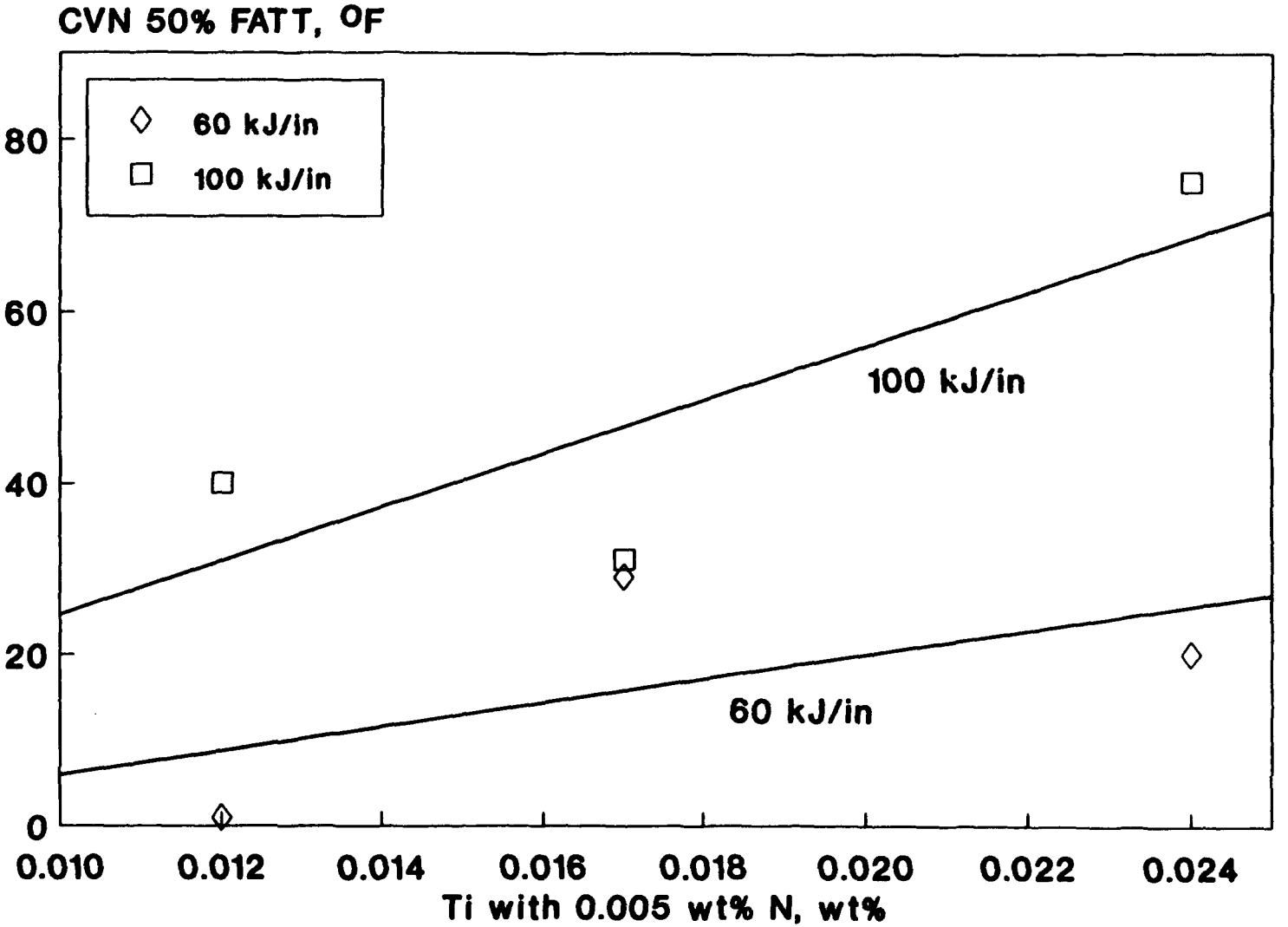

Fig. 11. Effect of TiN on LCB weld metal impact performance.

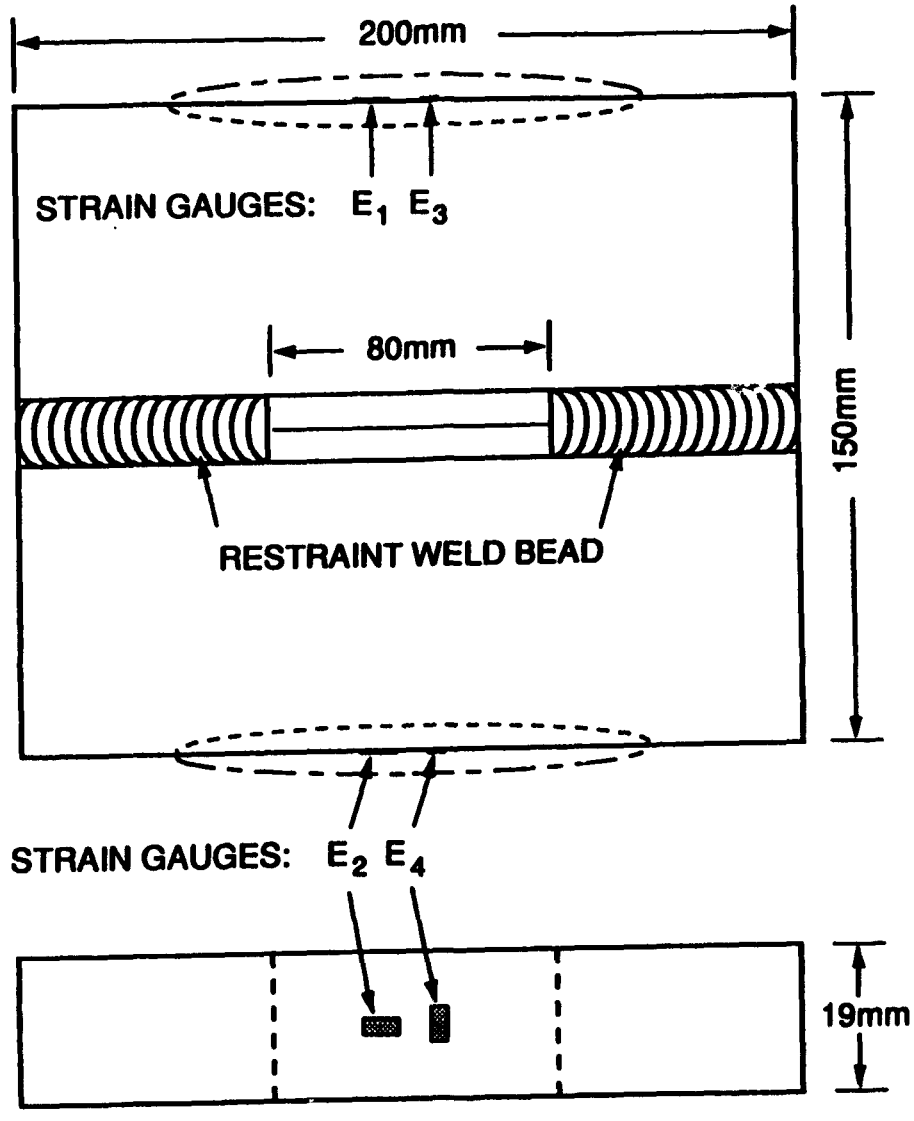

Fig. 12. Schematic illustration of a strain gauge instrumented Y-groove specimen (after Matsuda *et al* [1984]).

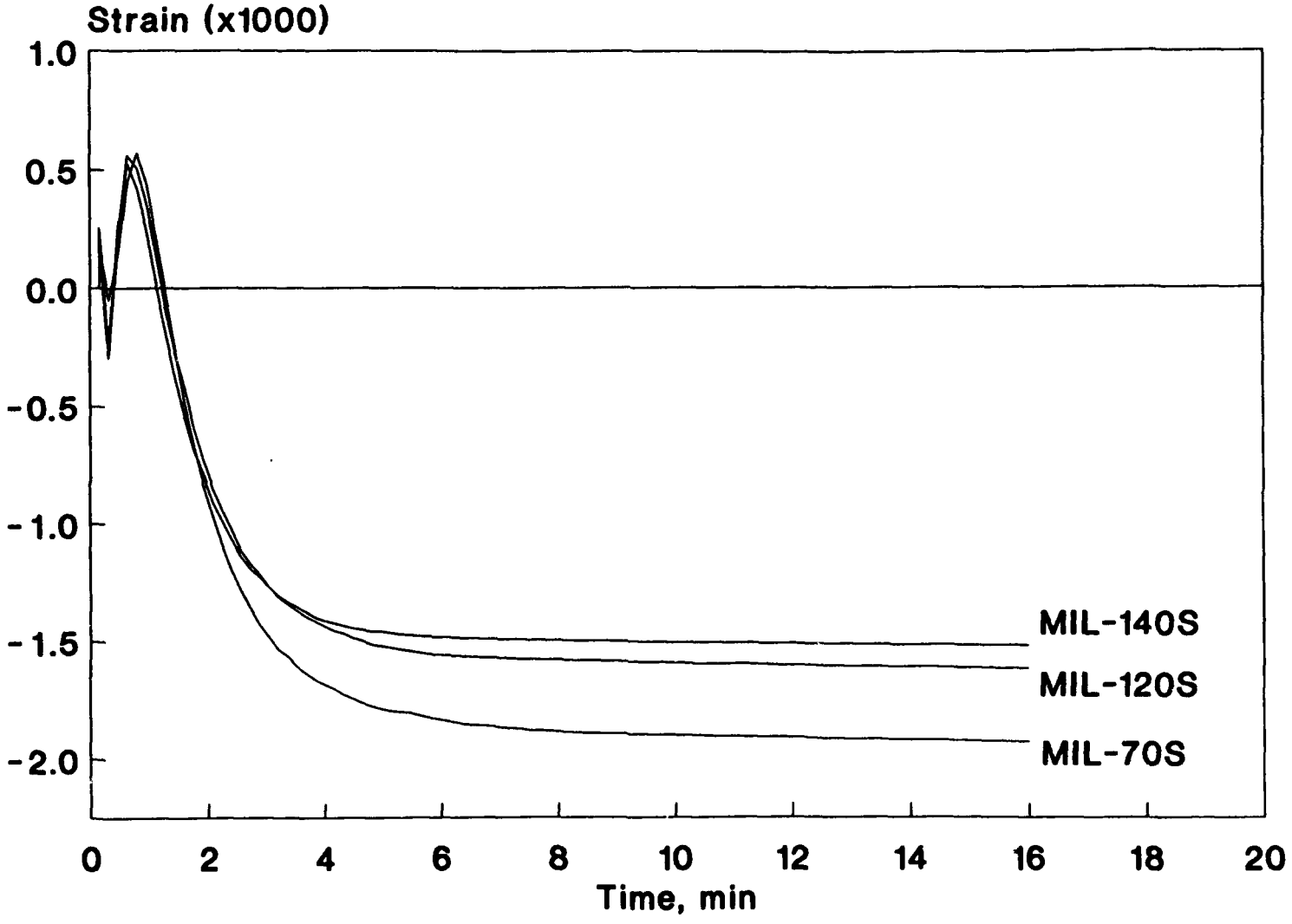

Fig. 13. Development of strain on the flank of a Y-groove specimen using GMAW consumables.

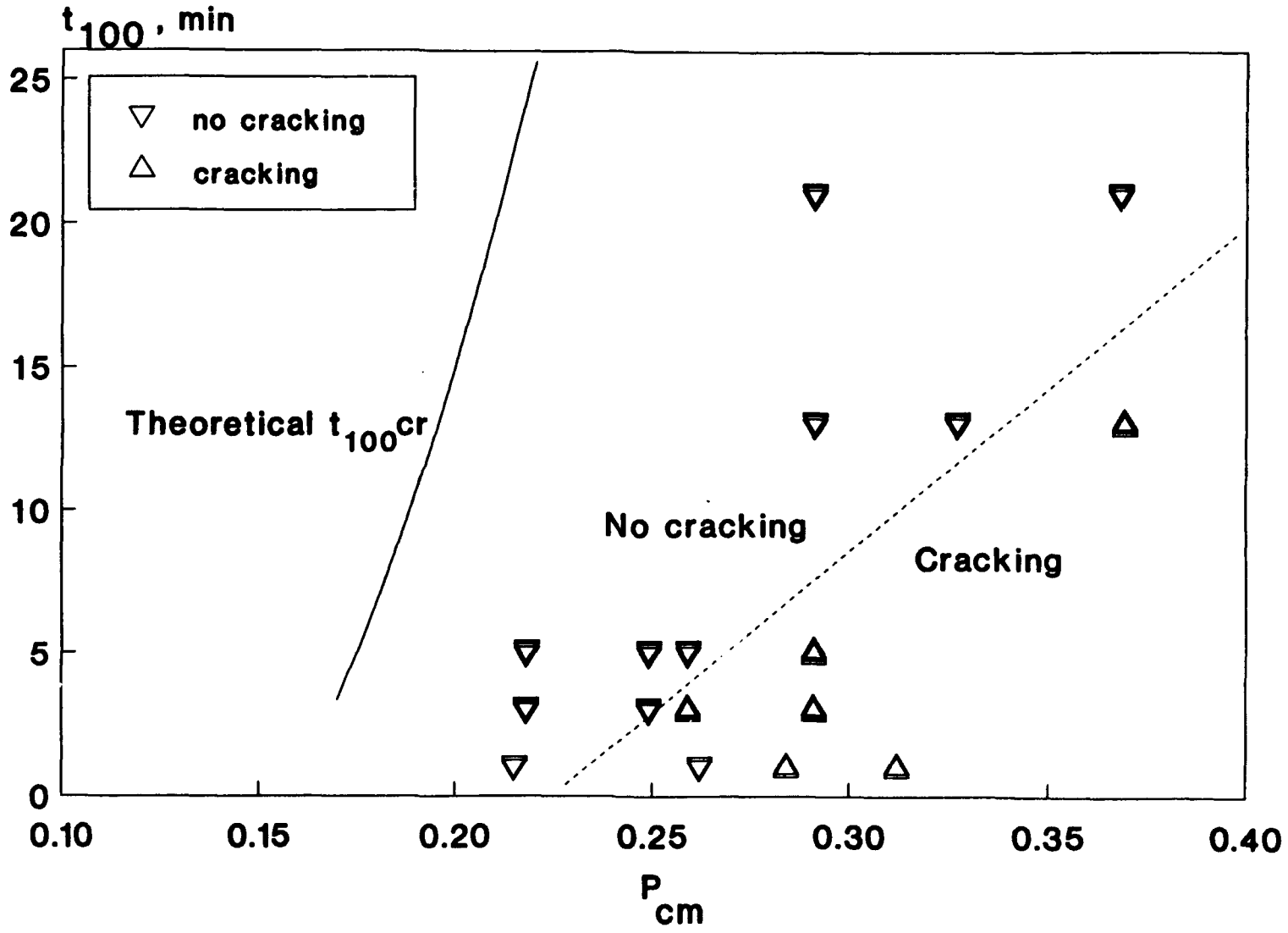

Fig. 14. Results of WIC and GBoP weldability tests using SMAW consumables.

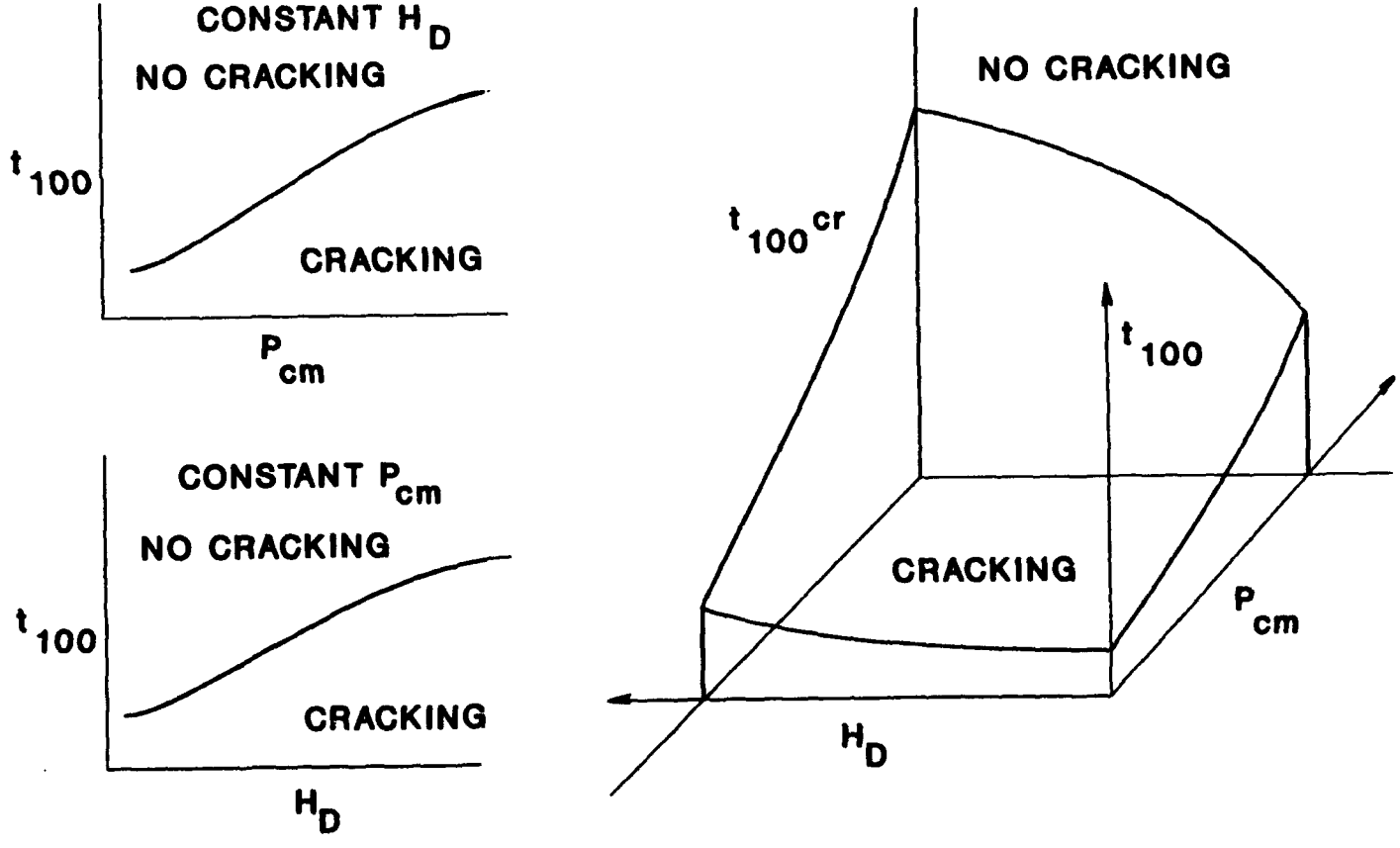

Fig. 15. Schematic response surface for hydrogen cracking resistance.

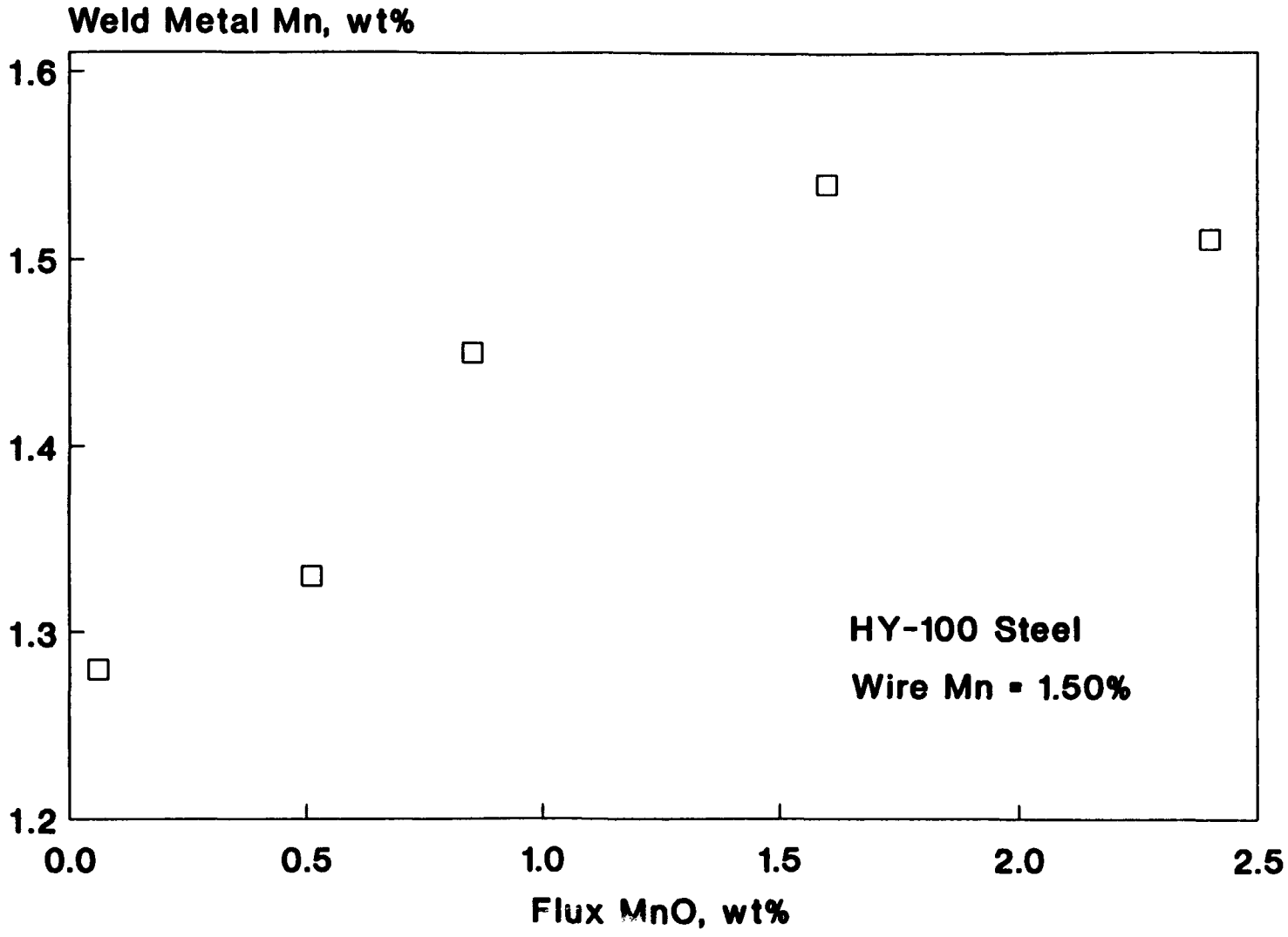

Fig. 16. Effect of flux MnO content on weld metal manganese content.

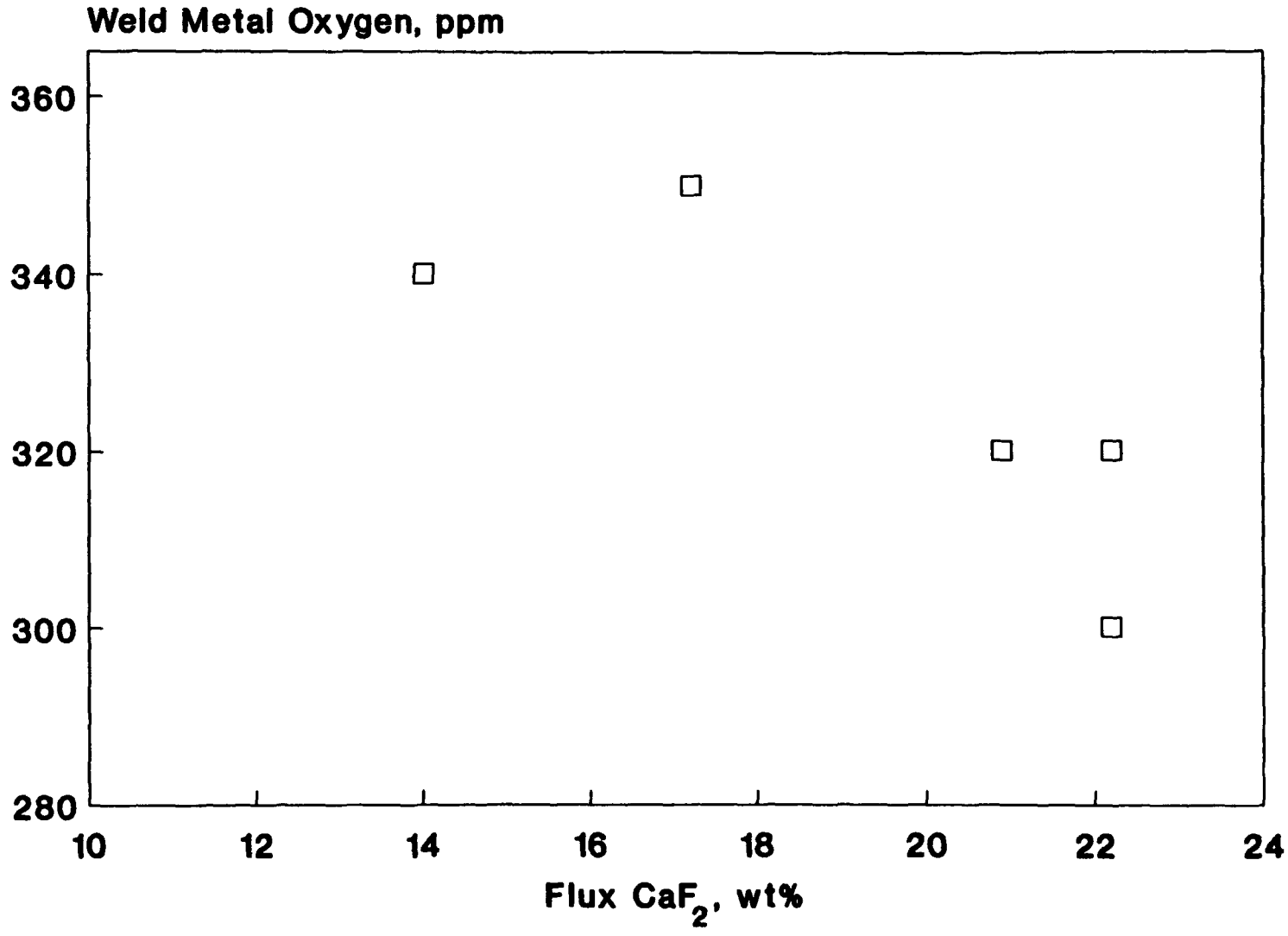

Fig. 17. Effect of flux CaF_2 content on weld metal oxygen content.

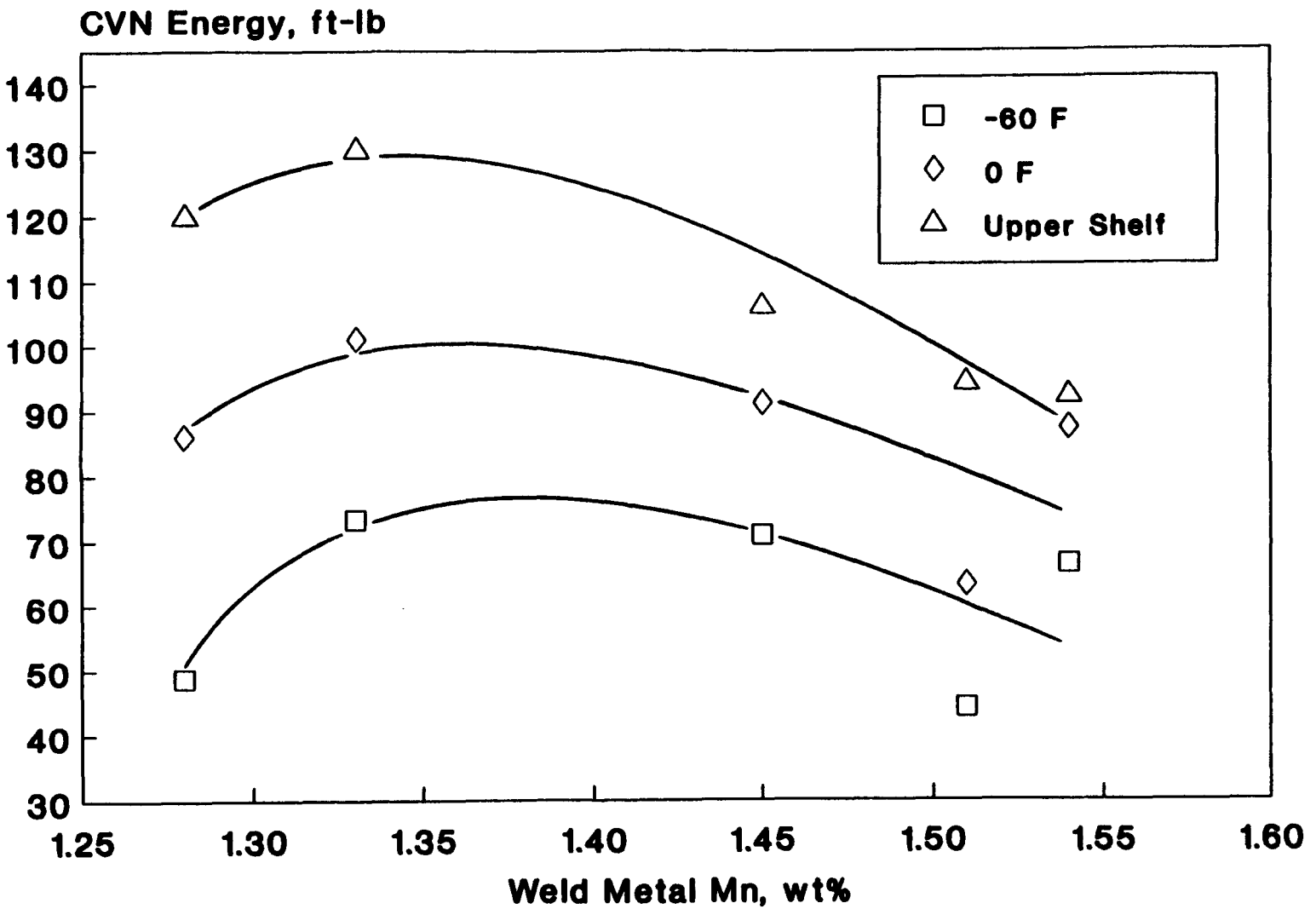

Fig. 18. Effect of weld metal manganese content on CVN performance.

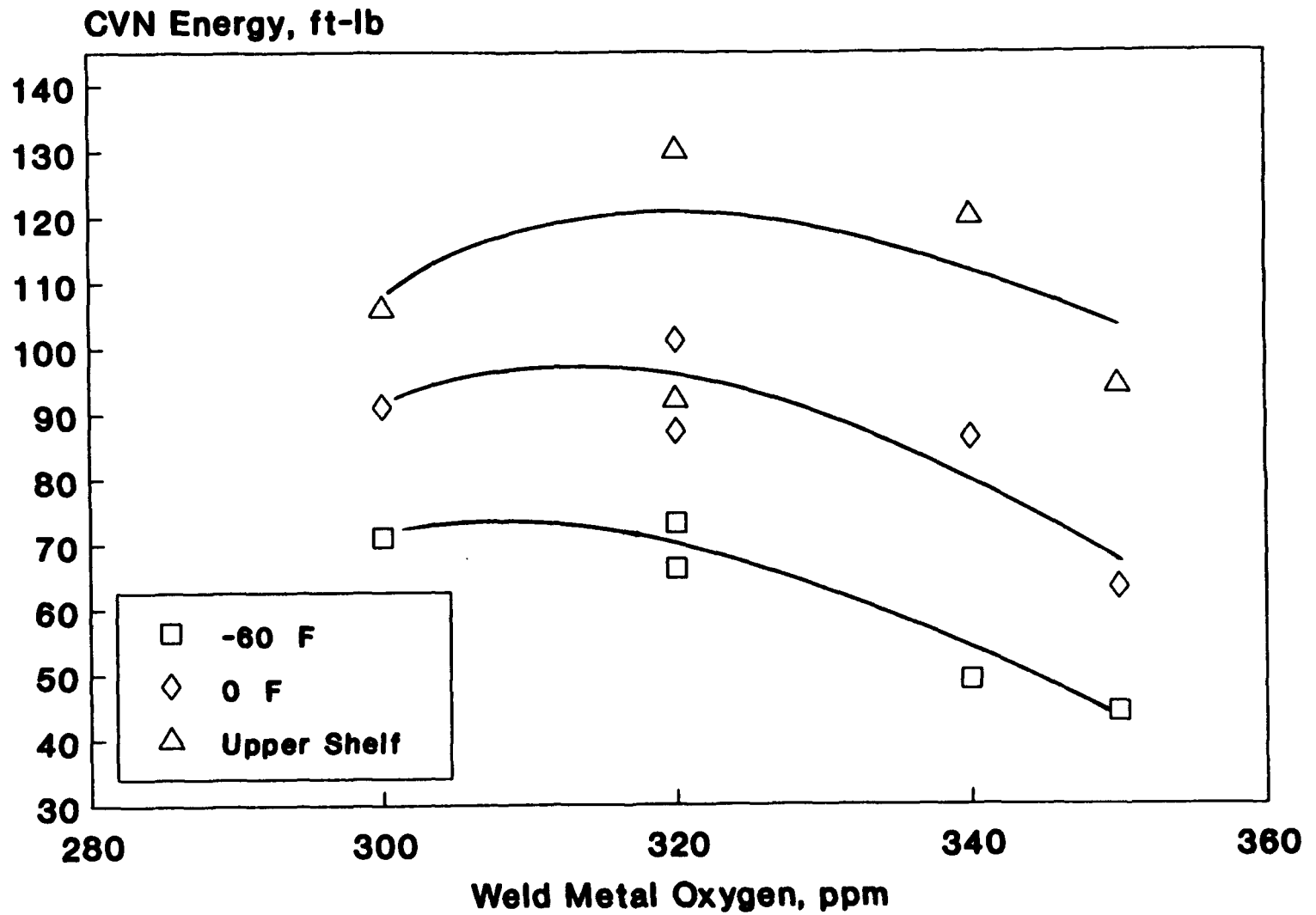

Fig. 19. Effect of weld metal oxygen content on CVN performance.

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13. ABSTRACT *(Maximum 200 words)*

One of the thrusts of U.S. Navy research is directed toward providing new and advanced material systems with improved properties, and developing methods and materials to construct current and future generation naval vessels more economically. It focuses on materials of construction, fabrication methods and consumables, and methods to ensure and enhance structural integrity. One specific area in this thrust is that of welding the Navy's high strength steels. Toward this end, work is being conducted on a number of topics pertaining to welding Navy steels with yield strengths exceeding 100 ksi.

Four tasks are discussed in this presentation. HSLA–100 Welding Consumables Development addresses evaluation of experimental compositions, data analysis, and identification of optimum compositions. Low-carbon Bainitic Weld Metals discusses the effects of alloying on weld metal strength and cooling rate sensitivity, and the effect of titanium-bearing inclusions on weld toughness. Weldability Methodology addresses transformation expansion, weldability tests, diffusible hydrogen, and cracking models. Welding Fluxes discusses determination of flux composition and correlations with weld performance. Future research for each task is also described.

14. SUBJECT TERMS
Weld metal, Welding consumables, HSLA–100, Low-carbon bainite, Weldability, Diffusible hydrogen, Welding flux, Steel

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