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FACTORS THAT DETERMINE THE APPLICABILITY OF HIGH STRENGTH QUENCHED AND TEMPERED STEELS TO SUBMARINE HULL CONSTRUCTION

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ABSTRACT

An analysis is given of the potentials and limitations of high strength, quenched and tempered (Q&T) steels for welded construction of submarine hulls. The experience that has been developed in the use of such steels in a wide variety of applications provides the essential background for preassessment of fabricability and design problems in hull construction, as a function of increasing strength level. With increasing strength level, there is a decrease in fracture toughness level which requires an upgrading of fabrication and design quality levels for purposes of maximizing structural reliability.

The explosion tear test and the drop-weight tear test and new methods designed for the evaluation of the fracture toughness of plate materials and welds. The explosion tear test is used to estimate the flaw size-fracture stress relationships. Correlations of these data with drop-weight tear test data and Charpy V test data provide an indirect assessment of these relationships. These tests cover an extensive range of materials and strength levels.

Collective consideration of these factors lead to the conclusion that Q&T steel hull fabrication capability at the 120-130 ksi yield strength level may be attained by a short term development program. The attainment of similar capabilities at the 150 ksi yield strength level is a "ceiling aim" requiring research solutions relating to weld metals, prior to entering a development stage. The attainment of a practical fabrication capability at yield strength levels in excess of 150 ksi is not feasible. The best promise of hull steels with strength levels in excess of 150 ksi is in the development of an entirely new family of materials based on new metallurgical concepts of hardening.

PROBLEM STATUS

This is a final report on one phase of this problem; work on other phases of the problem is continuing.

AUTHORIZATION

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FACTORS THAT DETERMINE THE APPLICABILITY OF HIGH STRENGTH QUENCHED AND TEMPERED STEELS TO SUBMARINE HULL CONSTRUCTION

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The aim of this presentation is to provide an analysis of the potentials and limitations of high strength quenched and tempered (Q&T) steels for welded construction of submarine hulls. As for any other class of materials, the important considerations resolve to an assessment of the relative level of construction difficulty, service reliability, and cost. For a given class of materials, these factors are interrelated to greater degree than is commonly recognized. The basic parameters which commonly affect these factors involve the physical metallurgy and fracture toughness characteristics of the metal. Thus, as the strength level of Q&T steels is increased, the attainment of desired levels of structural reliability requires the attainment of certain minimum values of fracture toughness which, in turn, require closer control of steelmaking and fabrication practices, resulting in increased costs.

It should be emphasized that there is no absolute measure or gauge of the "feasibility" of fabrication for any structure, particularly if "feasibility" is considered to represent factors other than those of technological attainment. A technological attainment "feasibility" that does not recognize facility or cost feasibilities may serve no useful purpose. Thus, the analysis that is required involves questions of relative constraints to the construction of a reliable structure, i.e., fabricability.

For the case of Q&T steels, the relative constraints for fabrication at various levels of strength have been evaluated over the past 25 years for a wide variety of structural applications. This broad background of information may be assembled in the form of a continuous spectrum of fabrication constraints related to the spectrum of strength levels ranging from the 80 ksi to over 250 ksi* levels. When this is done, the fabrication constraints for the strength range of primary interest (HY-100 to HY-150 ksi) become evident; the reasons for primary interest in this range of strength level also become evident.

FABRICABILITY CONSIDERATIONS

Discussions of fabricability cannot be divorced from considerations of design complexity, size of structure, service stress levels, and allowable costs. Thus, a component such as an aircraft landing gear which is designed by the ultimate of photoelastic stress analysis techniques, and for which high fabrication costs are acceptable, cannot be compared on equal basis with a complex structure such as a submarine hull. Relatively brittle steels which pose severe fabrication problems may be considered for the aircraft component case and not for the submarine hull case. In evaluating the fabricability of high

*All strength levels quoted relate to yield strength. The term HY is applied in reference to a material considered to relate specifically to submarine hull construction.

strength steels for submarine hull applications, the boundary conditions are much more stringent – the steel has to conform to metallurgically unfavorable fabrication practices. In the case of the aircraft landing gear, the fabrication practices are made to conform to the limitations of the steel.

Conforming to the limitations of the steel involves increases in fabrication costs. Thus, questions of fabricability limits are basically questions of cost acceptability limits. The cutoff occurs at a point such that a sharp discontinuity in the increase in welding difficulty, resulting from decreased weldability, causes recourse to welding minimization procedures which are always expensive.

Much has been said about the welding difficulties of HY-80 and the need to avoid such difficulties for "new" steels or other high strength metals. The facts of the case are that the submarine yard welding difficulties of the HY-80 lead (early) submarine construction period derived from attempts to fabricate a strong Q&T steel with minimum modifications of the low order, quality control procedures formerly used for the 50 ksi, normalized, HTS steel. The resulting "flap" was predictable and the "fix" simply involved the rigorous application of quality control. In other words, the submarine yard fabrication methods had to be upgraded from the low levels of the practices that were tolerable by a low strength steel to those that were tolerable by a steel of higher strength level.

It is an unalterable fact of life that with increasing yield strength there must be concomitant increase in the quality level of fabrication practices – these may be described broadly as the characteristic practices of the trade involved with fabrication of steels of specific levels of strength. Figure 1 illustrates the experience developed from the late 1930's to date in the various "trades" with projections based on the experience trends.

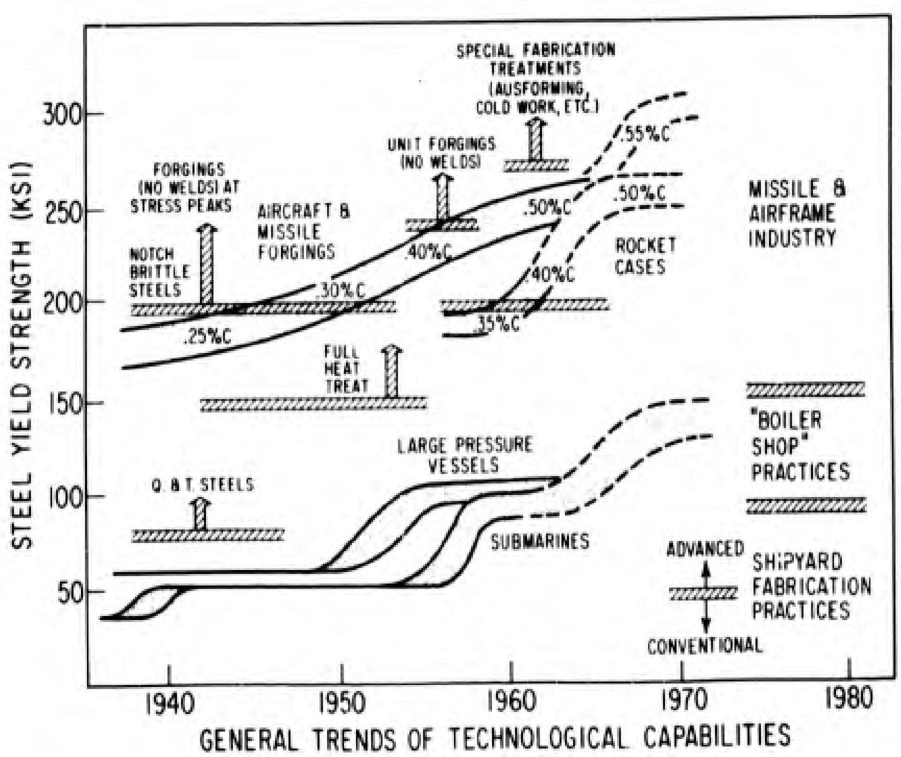

Fig. 1 - Illustrating the general requirement of increased level of fabrication quality with increase in strength level

The aircraft industry developed technological capabilities for using steels of increasing yield strength following the trends indicated by the upper shaded band. In the late 1930's, the use was restricted to 160 to 180 ksi steels of 0.25% C level – all weldments of strength greater than 150 ksi were found to require full Q&T heat treatment following welding. For this strength range, it was established that welds could be located at regions of relatively high stress level. A great deal of development effort was expended in raising the strength level that could be utilized in aircraft design. It was established further that all steels of over 200 ksi strength were notch sensitive (basically brittle steels), requiring the elimination of welds at stress peak points in the 200 to 240 ksi strength range, and complete elimination of welds at strength levels in excess of 230 ksi. The noted requirement for increased carbon content reflects a delicate metallurgical trade-off between strengthening and embrittling effects of increased carbon content. For example, at 240 ksi the use of higher or lower carbon contents than the noted 0.40% C would result in a more brittle (more sensitive to minor flaws) steel. A similar experience was obtained in the last 1950's for rocket case construction. Thus, the experience in the 160 to 250 ksi range was that the increasing levels of strength could be utilized only by increased sophistication in fabrication of the material, which ranged from permitting welding at stress peaks to complete elimination of welding. The fabrication practices required for this range of strength levels may be categorized as those of the missile and airframe industry.

The pressure vessel industry began to utilize Q&T steels in the 90 to 120 ksi range in the early 1950's. The steels were found to require "boiler shop" welding practices, involving accurate fit up, controlled preheat, and high quality fabrication. Such "boiler-shop" practices may be described as highly sophisticated for the structural steel industry but crude by standards of the airframe and missile industry.

The use of steels in the 40 to 60 ksi range has been characterized by welding fabrication practices of the lowest order of attention to details and quality control. In other words, these steels had a demonstrated tolerance for mistreatment in fabrication that was not tolerable for the Q&T steels of the 80 to 120 ksi strength range. Thus, the change from HTS (50 ksi) submarine fabrication to HY-80 (80 to 90 ksi) fabrication required a change from "conventional" to "advanced" (present) shipyard fabrication practices. It is clear that submarine hull fabrication of 120 to 150 ksi steels will require "boiler shop" practices, which are considerably more demanding of quality control practices than the present "advanced" submarine yard practices. There is no magic wand or mysterious ingredient for the steel that will provide for fabrication of such Q&T steels by present levels of submarine yard technology.

RELIABILITY CONSIDERATIONS

The requirements for upgrading of fabrication practices, with increased yield strength, derive in part from the increase in propensities for cracking, fissuring, and other forms of weld-area defects. The basic causes for such tendencies are directly relatable to the increases in hardenability and transformation product hardness, which is consequent to the alloy modifications required for developing increased strength. A second effect must now be recognized, that of a decrease in fracture toughness to the level of semibrittleness and then to the brittle state. Thus, the problem involves not only an increased disposition to flaw development with increasing strength but also a drastically decreased tolerance for the presence of flaws. In the face of such a double jeopardy – the recourse must naturally be not only the improvement of welding practices but also the elimination of welds from positions of high stress and the refining of the design features so as to eliminate positions of high stress. The higher the level of strength and the lower the tolerance for defects, the greater is the necessity for concurrent use of all three methods for increasing the reliability of the structure to acceptable levels.

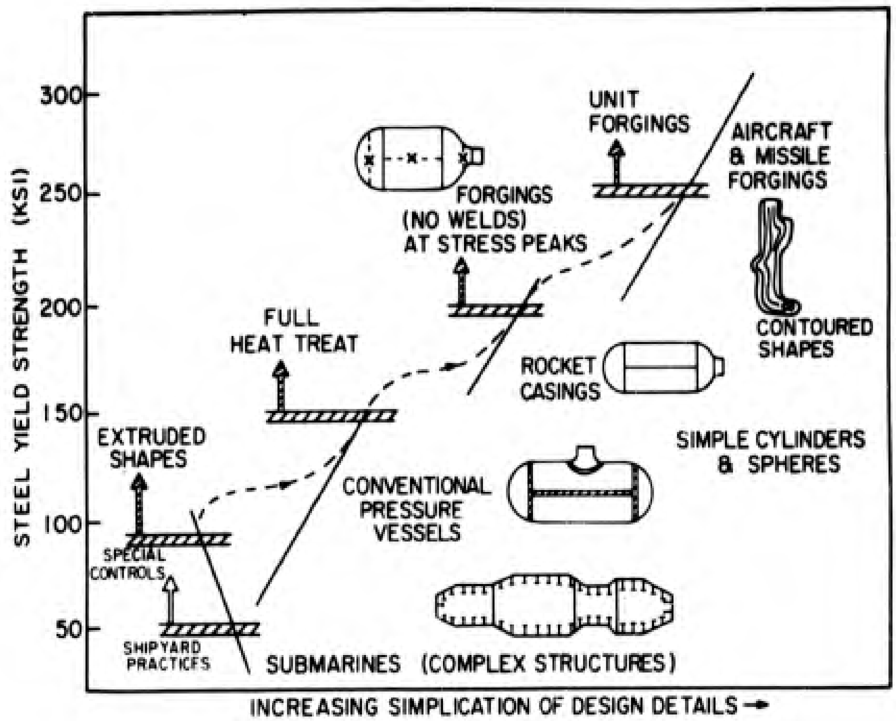

Fig. 2 - Illustrating the general requirement for increased design quality with increase in strength level

The intimate relationships between upgraded fabrication practices and upgraded design features are illustrated in Fig. 2. Again, we are discussing limitations that are documented by extensive experience. Design use of ultrahigh strength steels is restricted to highly contoured (stress peaks eliminated) unit forgings (no welds). Rocket case pressure vessels constructed of 190 to 230 ksi steels must be designed so as to eliminate welds from points of high bending moment – nozzle ports must be 3D machined from forged stubs, etc. Such extreme requirements are not necessary for rocket case pressure vessels constructed of 160 to 180 ksi steel – for this strength range, welds may be placed at points of high bending moments. Conventional pressure vessels of 100 to 120 ksi steel may be designed with much less attention to the transition of the stress flow lines from heads to cylinder and to nozzles, i.e., a minimum of detailed stress analysis is required.

That the use of steels of increased yield strength requires increased attention to simplification of the design details, is another unavoidable fact of life that was not recognized in the change from HTS to HY-80 construction of submarines. The present submarine structures are much more complex than the old ones. This is due to the requirement for the use of transition cones in areas of cylinder diameter change. The unique feature of a decreased simplification of design with increased strength must now be rectified by a switchback to designs more closely representing simple cylinders if steels of higher strength levels are to be used.

One may now ask, what is it that characterizes a given yield strength level to relatively specific and narrow ranges of design and fabrication finesse? Do we know enough to be sure that the modern day miracle of a "breakthrough" cannot be achieved? The answer to both of these questions is that we know enough regarding the basic factors that dictate the fracture toughness of Q&T steels to be realistic about the prospects for the next decade. These prospects are that there are no fundamental concepts that would suggest a change in the described situation for Q&T steels.

METALLURGICAL CONSIDERATIONS

The steel that we know as HY-80 evolved from 50 years of research and development, first as armor plate and then as a weldable structural material. The Ni-Cr-Mo family of armor steels was identified circa 1900 as the type that developed the highest level of toughness in ballistic testing. The composition adjustments that were made for the thickness range of present interest (1 to 3 inches) were first established empirically by high obliquity, ballistic tests which entailed a glancing blow and therefore required the utmost of fracture toughness to deflect the uncapped test projectile. At this stage of development, the steel was known as STS (special treatment steel, i.e., a Q&T steel). Interest in the use of STS for structural weldments developed during the late 1940's and explosion bulge tests of weldments were conducted in the early 1950's. Explosion test comparison of STS weldments with those of a material of lower carbon content, known briefly as Low Carbon STS, indicated much improved performance, particularly for the heat affected zone. Moreover, the general level of weldability was much improved, especially with respect to resistance to cracking when welded under restraint. These tests were paced by the development of a new series of notch tough electrodes of the 100 to 110 ksi yield strength class. There was no case in prior history of the development of a weldable hull structural material in which a similar level of research and test effort was expended in proving out a material in a welded state.

In late 1955, a decision was required between the steels that were contenders for hull fabrication use in the new classes of submarines. The materials and their characteristic performance in explosion bulge tests at 30 °F were as follows:

(1) HTS – brittle fracture in the presence of small flaws,

(2) Code Y (proprietary structural steel) – always resulted in complete separation in the HAZ band,

(3) STS – limited shear tearing always occurred in the HAZ band –.difficult to weld, and

(4) HY-80 – outstanding resistance to fracture, even in the presence of brittle, crack-starter welds.

In effect, there was no contest, the superiority of HY-80 was clearly obvious. All that remained was to "marry" this material and associated welds to a structural design appropriate for the strength level and to adopt controlled welding procedures required for a steel of high hardenability. The followthrough in these respects was not entirely adequate and the ensuing fabrication difficulties experienced by the lead yards may now be recognized, dispassionately, as a lesson for the future. A new material is not fully developed until a suitable "marriage" also is made with a proper level of design quality and with proper quality control procedures in the construction yards.

The popular opinion that the HY-80 fabrication difficulties derived from an inadequate balance of the alloy elements for welding or that somehow the weld was not "quite right" for the steel, are without foundation in fact. The dramatic advances in physical metallurgy of the 1930's provided the metallurgist with all necessary tools for improving and adjusting the HY-80 composition. The principal adjustment that was necessary, involved the lowering of the carbon content, following well-established scientific principles. Inasmuch as these principles are involved in the development of Q&T steels of higher strength level, we shall now discuss such factors.

Figure 3 illustrates the primary effects of the alloying elements present in HY-80. The carbon content determines the strength level and ductility – for the 80 ksi level of strength, the optimum carbon content is in the range of 0.13% to 0.18%.

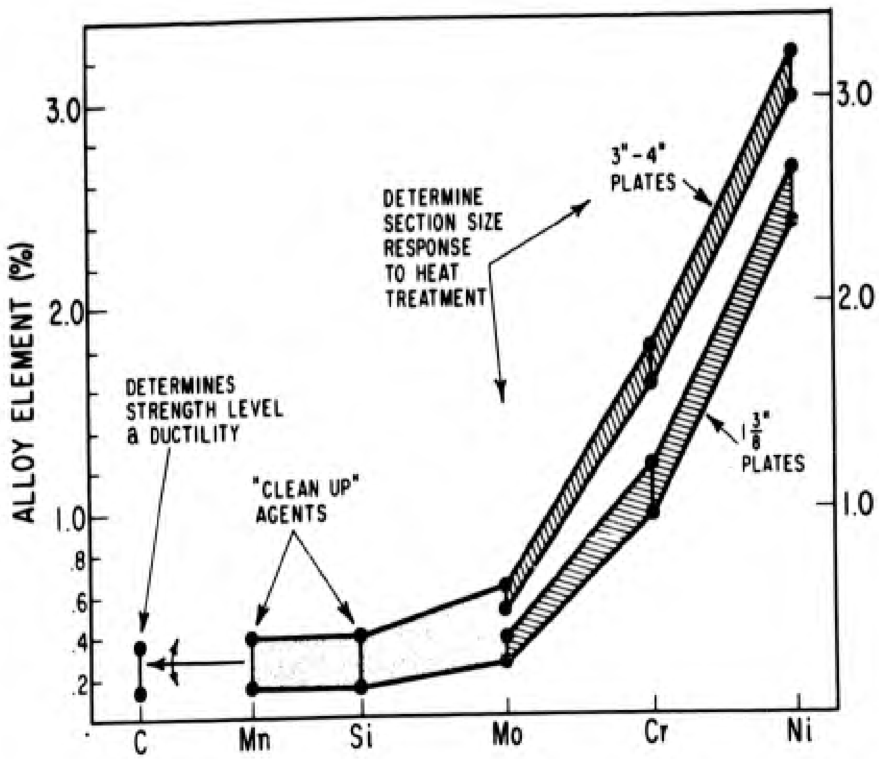

Fig. 3 - Primary effects of the alloy elements present in Q&T steels of the HY-80 family

Manganese and silicon are "clean up agents," in the sense that undesirable constituents (P,S,FeO) are removed from solution and either floated out of the bath or rendered harmless by fixation as inclusions in the steel.

Molybdenum, nickel, and chromium are the "alloy" elements that determine the transformation of the steel to "best" metallurgical structures on quenching. As the section size increases, it is necessary to increase the "alloy" content, as indicated in the figure for thin and thick plates – the bands represent the "low" and "high" levels of the chemistry range of HY-80, as presently used.

The transformation features of HY-80 are illustrated by Fig. 4. At high temperatures, the steel is in the austenitic (crystal form) state. On cooling, the austenite transforms to a variety of possible transformation products, characteristic of the temperature of transformation. For the diagram shown (specific to a given Ni-Cr-Mo composition) the cooling rate obtained by water quenching of a 1-inch plate is such as to result in transformation at 700 °F – at this temperature the product is the hard and brittle structure termed martensite. For the cooling rate obtained by water quenching of a 3-inch plate, the product is largely upper bainite – an undesirable product of medium strength and low ductility. For the particular composition, 1- and 2-inch plates transform to a desirable hard product but 3-inch plates do not. Accordingly, for 3-inch plates it is necessary to increase the Ni, Mo, and Cr contents, thus "pushing" the transformation diagram to the right and resulting in transformation at the 3-inch plate cooling rate, to the desired lower bainite and martensite structures.

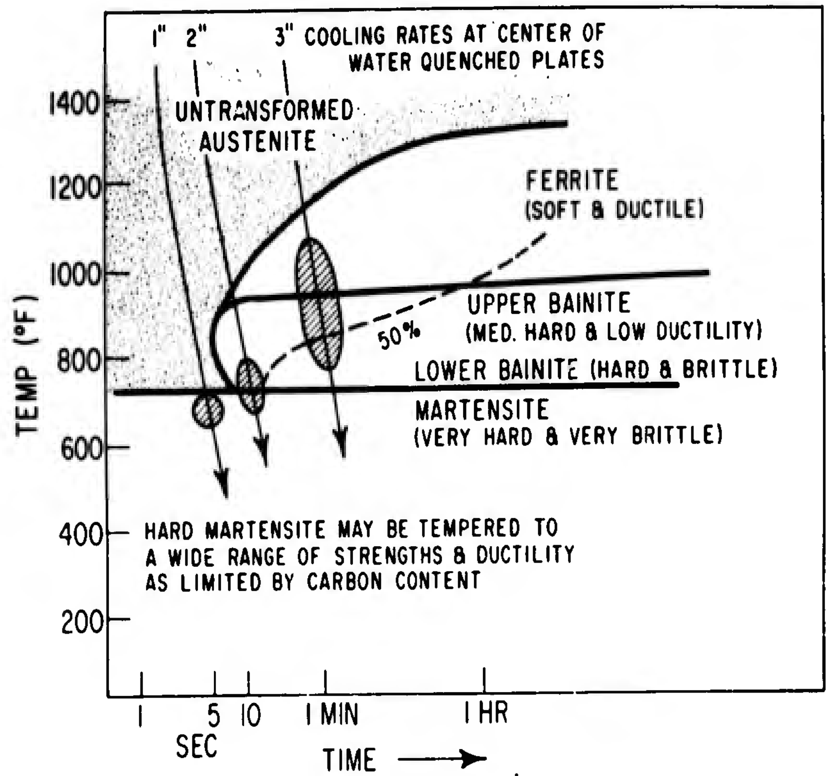

Fig. 4 - Transformation features of HY-80

Following proper quench transformation to martensite (or lower bainite), it is then necessary to temper the hard and brittle product to a softer and more ductile product, known as tempered martensite. Figure 5 illustrates that the carbon content of the martensite and not the alloy element content, dictates the strength level of the tempered steel. Thus, all steels of a given carbon content that have been properly transformed to martensite (or the equivalent lower bainite) on quenching will resort to essentially the same strength level when tempered at 1200°-1250°F (lower line). The strength of HY-80 is determined by the fact that the carbon content is in the range of 0.13% to 0.18% and the highest possible tempering temperature (1200°-1250°F) is used. This combination results in a maximum of fracture toughness. Also shown in the figure is the strength level of common aircraft forgings which are tempered at the lowest temperature (400°-500°F) that is expected to provide for a minimal relief of quenching stresses. By varying the tempering temperature between the limits of 400°F to 1250°F, it is possible to obtain a wide range of strengths and fracture toughness – the trend being a decrease in the one property with an increase in the other.

For any specified yield strength, the maximum toughness is obtained by a trade-off between carbon content and tempering temperature. For example at 150 ksi, a carbon content of approximately 0.20%, coupled with a tempering temperature of 1050°F, provide the optimum tradeoff. Lowering of the tempering temperature to 950°F, as would be required if the carbon content was 0.14%, would give a 150 ksi steel of decreased toughness – the same would result for increasing the carbon and tempering temperature levels.

As the strength level is increased to the HY-150 to HY-200 range, the experience of aircraft steel producers, based on physical metallurgy principles, should be considered. Maximizing of the fracture toughness in this range requires the use of increased amounts of carbide forming elements, beyond the levels required for hardenability (quench hardening) purposes. Thus, added amounts of Mo and Cr may be required in proper balance.

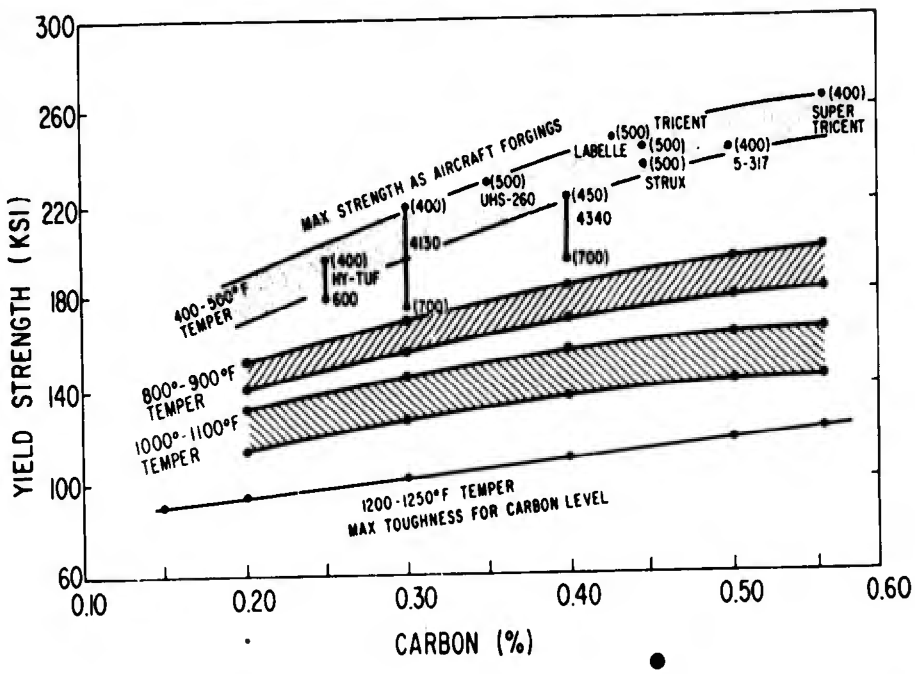

Fig. 5 - Relationships of carbon content and tempering temperature required to develop specific strength levels for Q&T steels

The "secondary hardening" effect provides for using lower carbon and higher tempering temperature combinations than are otherwise possible for the strength level. The addition of silicon and of vanadium provide toughness benefits at the higher levels of strength, however, at the level of approximately 150 ksi, such additions are of questionable merit. Major increases in nickel or other solid solution hardening elements, such as cobalt, may be considered for the 150 to 180 ksi range, however, the effects on increased strength are relatively small.

These relationships are well known, thus the problem of developing the optimum Q&T steel for a given strength level becomes one of considering the more subtle effects of second order variables – no miracles are to be expected, only a "finer" scale optimization of toughness for the strength level. The necessity for investigating second order variables (purity, special melting practices, etc.) varies with the difficulty of attainment of the desired level of toughness, for example:

(1) The attainment of a desirable level of toughness for combatant submarine hull purposes at 200 ksi is clearly outside of the feasibility of second order variables – i.e., it is too brittle to start with.

(2) There is no need for recourse to fine scale optimization at the 120/130 ksi strength level – a high level of fracture toughness is available.

(3) In 150 to 180 ksi range the fracture toughness level is marginal – any improvements obtained by fine scale optimization would be desirable and possibly worth the cost.

WELDABILITY CONSIDERATIONS

The basic problem of weldability is that the process of welding requires melting, solidification, and heat treatment under conditions that are less than ideal for the purposes of developing an optimum material for the strength class. Accordingly, the weldability problem is directly related to the intrinsic decrease in fracture toughness as the strength level is increased. At strength levels that require a fine scale optimization of steel making and heat treatment practices for the plate, the possibilities of nearly equal attainment under welding conditions for either the weld metal or the HAZ are remote. At strength levels such that considerable leeway is available in melting practice and heat treatment, the possibilities of closely equal attainment under welding conditions are good. Thus, the increase in strength level from HY-80 to HY-120/130 ksi, which does not require major upgrading of steel making and heat treatment practices, may be expected to be attainable for the weld and HAZ by reasonable upgrading of welding practices. This does not say that the weldability problem at HY-120/130ksi is not more difficult than that of the HY-80 level, but that the increased difficulty can be coped with by reasonable adjustments in fabrication practices and quality control. In effect, it may be concluded that existing yard facilities may be upgraded without major difficulties to HY-120/130 fabrication. The increase in strength level from 130 to 150 ksi is another matter – the significant upgrading in melting and heat treatment practices required for optimizing the properties of the plate product presages a marked upgrading requirement for welding practices. The implications are that new facilities may be required for fabrication at such strength levels.

We may now develop a breakdown of three principal areas of metallurgical control problems that exist with respect to the weld metal. These are:

(1) Restrictions on the carbon content of the weld metal because increasing carbon content above a maximum of 0.10% to 0.12% results in uncontrollable weld cracking.

(2) Requirement for developing strength levels equal to those of the plate with weld metal of much lower carbon content. Such an off-optimum approach to the strength level requires upgrading of melt quality, and particularly a decrease in the tolerable levels of O_2, N_2, P, and S.

(3) Restrictions in the range of cooling rates of the solidified metal and of the range of temperatures developed in tempering of the "hard" deposited weld metal by subsequent passes.

The attainment of a desirable control in each of the described areas becomes more difficult with increasing strength level. For example, at 120/130 ksi, weld metal of high fracture toughness can be developed with reasonable control. All known weld metals of the 150 ksi strength level are brittle even with the best of laboratory control practices. The difficulty of attainment does not increase gradually, but is involved with "critical edge," discontinuity situations.

We may now discuss the welding control problem with respect to cooling rates developed in the weld and the HAZ. The essence of the cooling rate problem is illustrated in Fig. 6, in relation to the transformation diagram of HY-80. The range of cooling rates for normal operating conditions of stick electrodes are indicated in association with a range of preheats that vary from 70°F (no preheat) to 500°F (normal maximum is 300°F). In this range, transformation products of good quality will be attained. Slower welding progression and higher heat inputs will result in transformation to bainite structures of poor ductility. Faster welding progression and lower heat inputs will result in development of martensite with consequent HAZ cracking. The normal trends, in the development of higher strength steels will result in moving the transformation diagram to the right, such that the "knee like" region will move into the described desirable cooling rate band for the HY-80 case. This then results in a requirement for a much narrower band of cooling rates for the avoidance of cracking in the higher strength steels. In other words, the controls required for the welding of steels of increasing strength levels will tend to move to limits that are too narrow for human control and necessitate automatic, machine controlled welding. Such welding cannot be done out of position (other than downhand) except

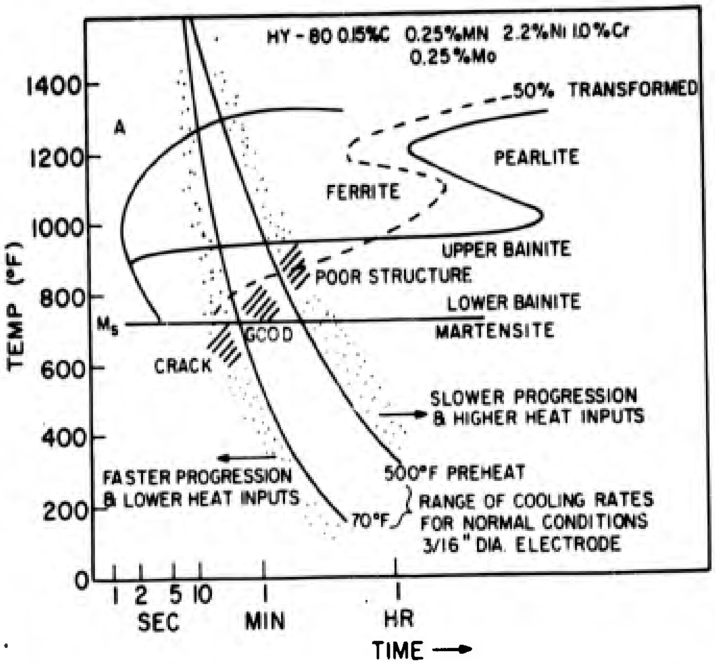

Fig. 6 - Relationships of cooling rates of weld region to nature of transformation product

with very special and difficult set-up procedures. The 150 ksi level is at the borderline of such a requirement for automatic welding techniques, that would be quite similar to those used for the fabrication of rocket cases.

Another aspect of this problem is the increased finesse required for applying the proper tempering treatment during welding for high strength steels that are heat treated to low tempering temperature. Figure 7 illustrates the basic effects of the "subsequent" pass in tempering the hard martensite region created by the previous pass. Proper placing of the next pass (tempering pass) is required to soften the hard zone that would otherwise crack in service or on cooling. For steels that are tempered at high temperatures (such as HY-80 which is tempered at 1250°F) this is enough of a trick, to call for reasonably expert welding and adherence to specified welding practices. For steels of 120/130 ksi level this will be more difficult and probably require greater use of automatic welding. For steels of the 150 ksi level, the expected difficulty of the operation is such as to reasonably question the practicability of nonautomatic welding.

FRACTURE TOUGHNESS ASPECTS

The ultimate aims of fracture toughness studies is the development of information that provides for evaluating flaw-size, stress, and temperature relationships of the fracture process. The engineer, as the user of this information, desires to know the flaw-size that will result in the development of a fracture at specific levels of stress and specific temperatures of services. After a decade of intensive study, NRL investigators developed a practical engineering approach which provides for answering such questions for the case of structural steels that are used in the transition temperature range, i.e., in the temperature range of brittle and semibrittle fracture. The "fan" in the diagram of Fig. 8 illustrates the flaw size that is required for the development of fractures at various levels of stress relative to the yield strength of the steel, at temperatures below the NDT temperature.

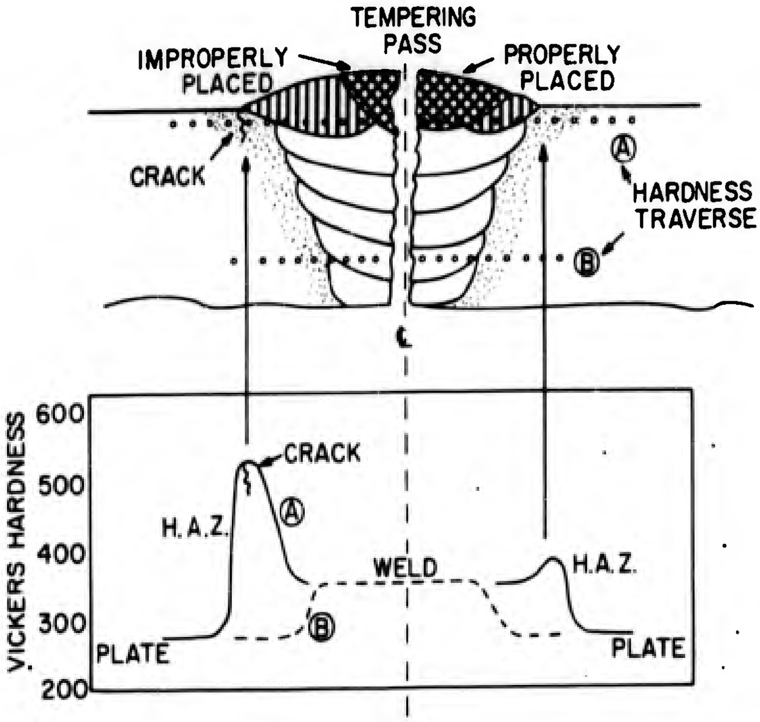

Fig. 7 - Illustrating the critical nature of tempering pass techniques

It is noted that fracture stress decreases with increasing flaw size. For a given flaw size the fracture stress increases markedly with increasing temperature at temperatures above the NDT temperature. Another interesting aspect is the increase in the crack arrest temperature (CAT) with increasing stress, at temperatures above the NDT temperature. A complete description of the "fracture analysis diagram" and its application is provided in Ref. 1.

The improvement in the fracture toughness which resulted from the shift from HTS to HY-80 hull materials may now be interpreted in terms of the fracture diagram. Figure 9 illustrates the shift in transition temperature as indicated by Charpy curves of average quality material. The NDT temperature shift is indicated also by correlation with Charpy V curve. Thus, for the lowest operating temperature of a submarine hull (assumed 30°F) the fracture toughness properties of average quality materials may be represented as being respectively:

(a) 20° to 30°F above the NDT for HTS

(b) 120°F or more above the NDT for HY-80

These relative positions are indicated by the large vertical arrows of Fig. 8. It may be noted that a wide range of flaw sizes and stress levels would result in brittle fracture for HTS. Conversely, fracture (ductile) of HY-80 would require loading to the ultimate tensile strength levels, even in the presence of very large flaws. This extreme resistance to fracture has been repeatedly demonstrated by underwater explosion tests of HY-80 structures.

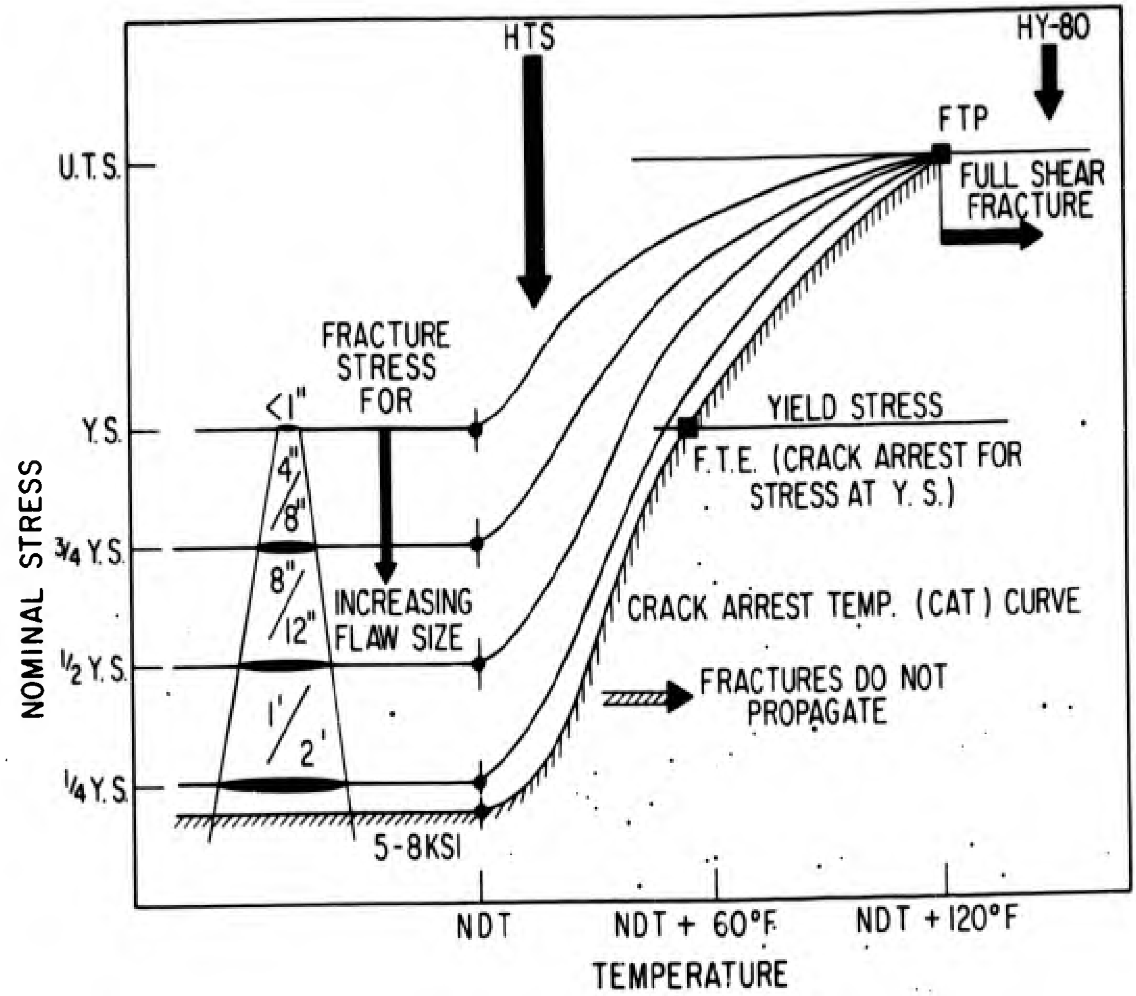

Fig. 8 - Features of NRL fracture analysis diagram

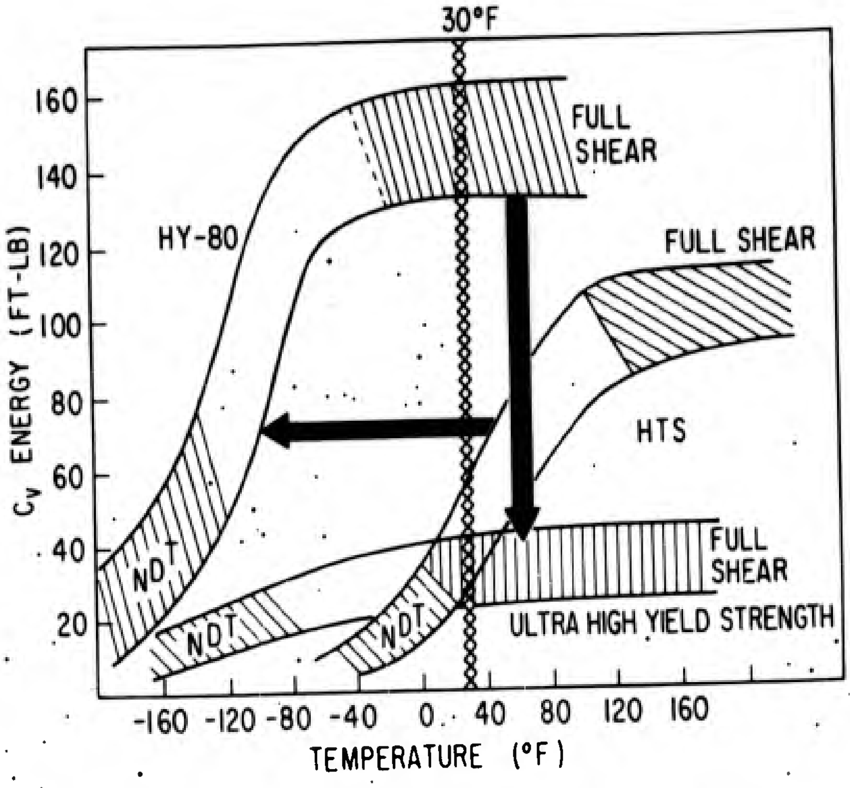

Fig. 9 - Representative Charpy V test energy transition curves for HTS and HY-80 compared to those of ultra high strength (180-200 ksi) steels

Such a high level of fracture toughness at temperatures of full shear fracture (above FTP) is characteristic of steels that have a high Charpy V upper shelf energy (to be designated in this report as C_V-SE). The problem that we now face in the use of steels of high yield strength is that the shelf energy may drop to very low values, as indicated by the curve labeled ultrahigh yield strength steels. In effect, the Charpy test is reporting that such steels may now be ruptured at ductile tearing temperatures with energy absorptions that are in the same order as those obtained for brittle fracture of low strength steels. The suggestion is that the various fracture initiation curves of the diagram of such "low energy shear" steels are "rotated" downwards into the position indicated by Fig. 10, as compared to a steel that is characterized by "high energy shear" properties.

NRL "post mortem" failure analysis of steels that displayed C_V-SE values of less than 30 ft-lb disclosed that such was the case. One example is that of shear fracture failures of solid propellant rocket cases on hydrotest, which initiated from flaws of very small size; Fig. 11 illustrates a typical failure (2). In fact, these initiating flaws were found to be of a size that was close to limit of detection possibilities for hydrotest stress levels slightly under the yield strength of the 190 to 200 ksi steels. The flaw-size fan on the right side of Fig. 10 illustrates this condition of small flaw fracture initiation at elastic levels of stress. Another example is the elastic load, shear fracture of large forged rings (3) that featured C_V-SE in the order of 23 ft-lb.

A review of the available knowledge in 1959 clearly indicated that:

(1) Ultrahigh strength steels of the 200 ksi plus range could develop "low energy shear" fractures at stresses equal to or less than the yield strength in the presence of very small flaws. There was no apparent alloy combination that would provide for better behavior.

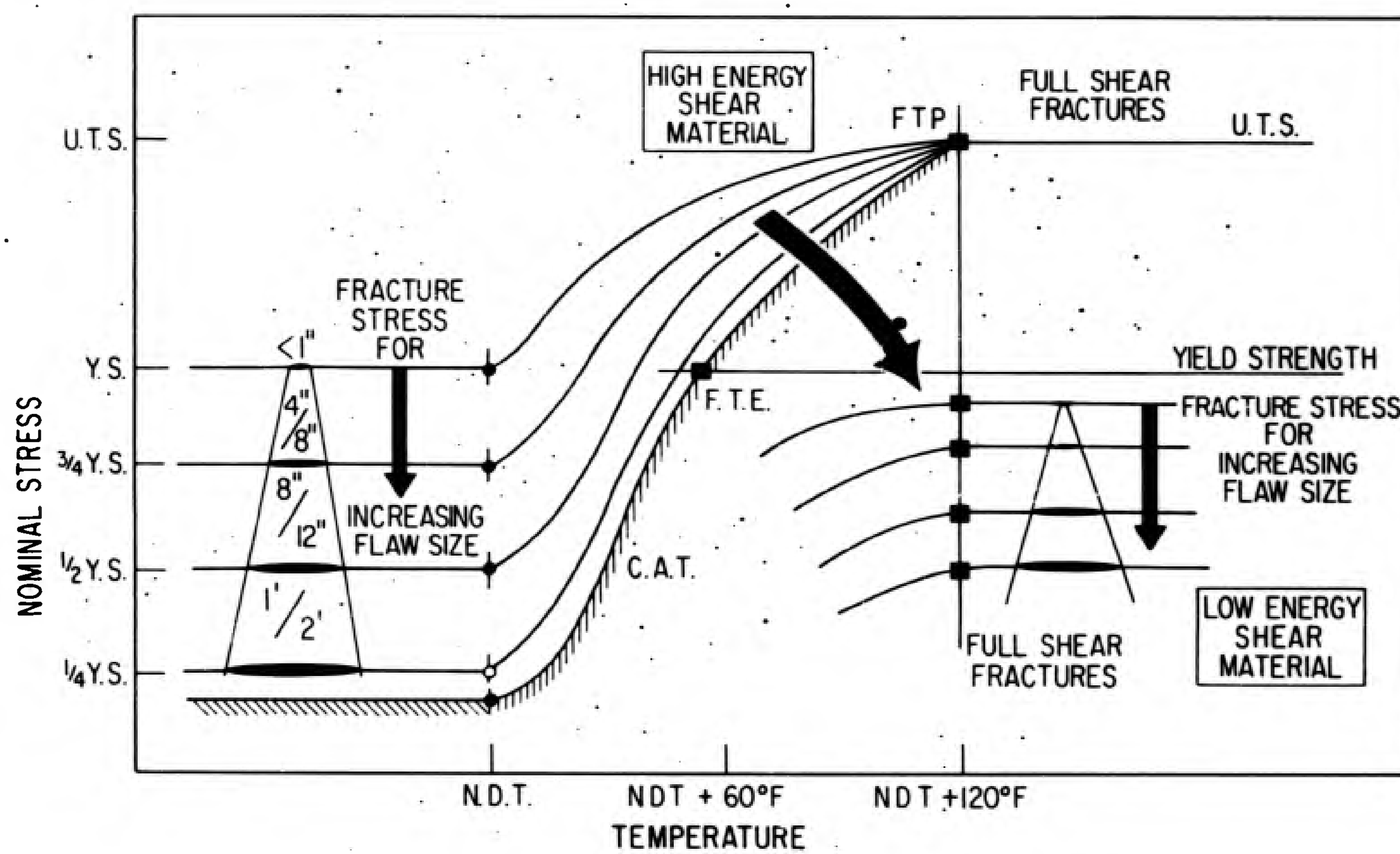

Fig. 10 - Features of NRL fracture analysis diagram modifications for steels of "low energy shear" characteristics

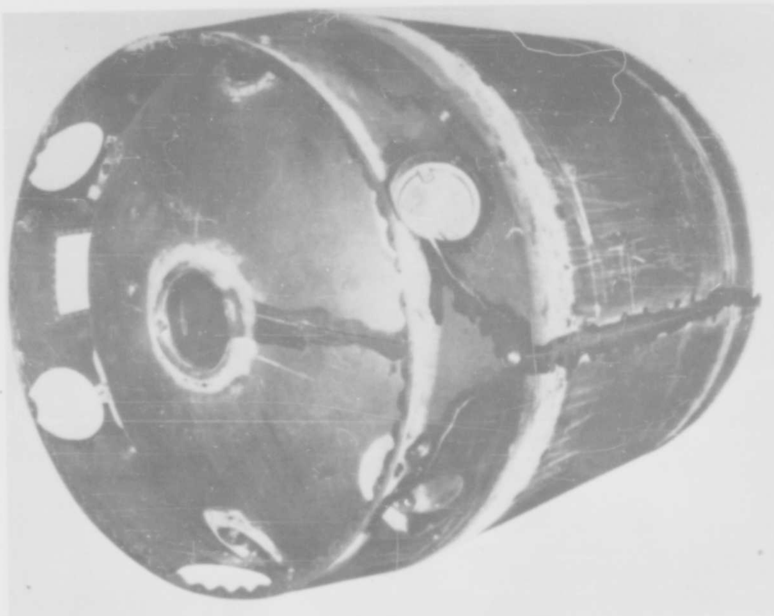

Fig. 11 - Typical hydrotest failure of rocket cases by shear fracture at hoop stress levels below those required for development of general yielding

(2) A wide variety of steels of less than 120 ksi yield strength could be expected to demonstrate very high levels of fracture toughness, as deduced from Charpy test data and armor plate test experience.

(3) Very little specific information was available for plate steels in the range of 130 to 180 ksi yield strength. The drastic decrease in C_v -SE shown by aircraft forgings of the high end of the strength range implied a serious lack of fracture toughness.

The rapidly developing interest in the use of ultrahigh strength steels for submarine hull fabrication, coupled with the general lack of fracture toughness information regarding such steels, clearly indicated the need for a research program in this area. Accordingly, a research plan was developed early in 1960, aimed at evaluating the flaw size – fracture stress relationships of high strength steels. The established goal was to develop a maximum amount of useful engineering information within the time span of an anticipated BuShips requirement period of two years. In the absence of any existing practical method for evaluating the fracture toughness of steels in the strength range of interest, it was necessary to improvise a "short circuit" approach based on entirely new test methods. This goal has been met and we are reporting today on the only available practical method that can be used to provide the information desired by designers or industrial producers. The findings are not an "end" but merely the first significant step in understanding the engineering characteristics of the 100 to 200 ksi strength range for plate material.

EXPLOSION TEAR TEST

Practical considerations dictated the use of explosion loading for evaluating the effects of flaws in thick plate material. By this means a large number of tests could be made at the temperature of interest (30° F), using a relatively simple specimen. A cylindrical bulge configuration (Fig. 12) was decided upon based on a requirement for propagating the fracture in a direction of essentially uniform loading. The top surface of the cylindrical bulge region provided this condition. It was then decided that the investigation of flaw-size effects should start with a "practical" flaw size, representative of the size that would be expected to exist or develop in cyclic load service of welded structures of respectable fabrication and design quality. This requirement dictated the choice of a 2-inch-long weld crack.* The explosion load was planned to be varied, so as to develop different levels of deformation, as a test of the ability of the steel to withstand various levels of plastic overload in the presence of the specified crack.

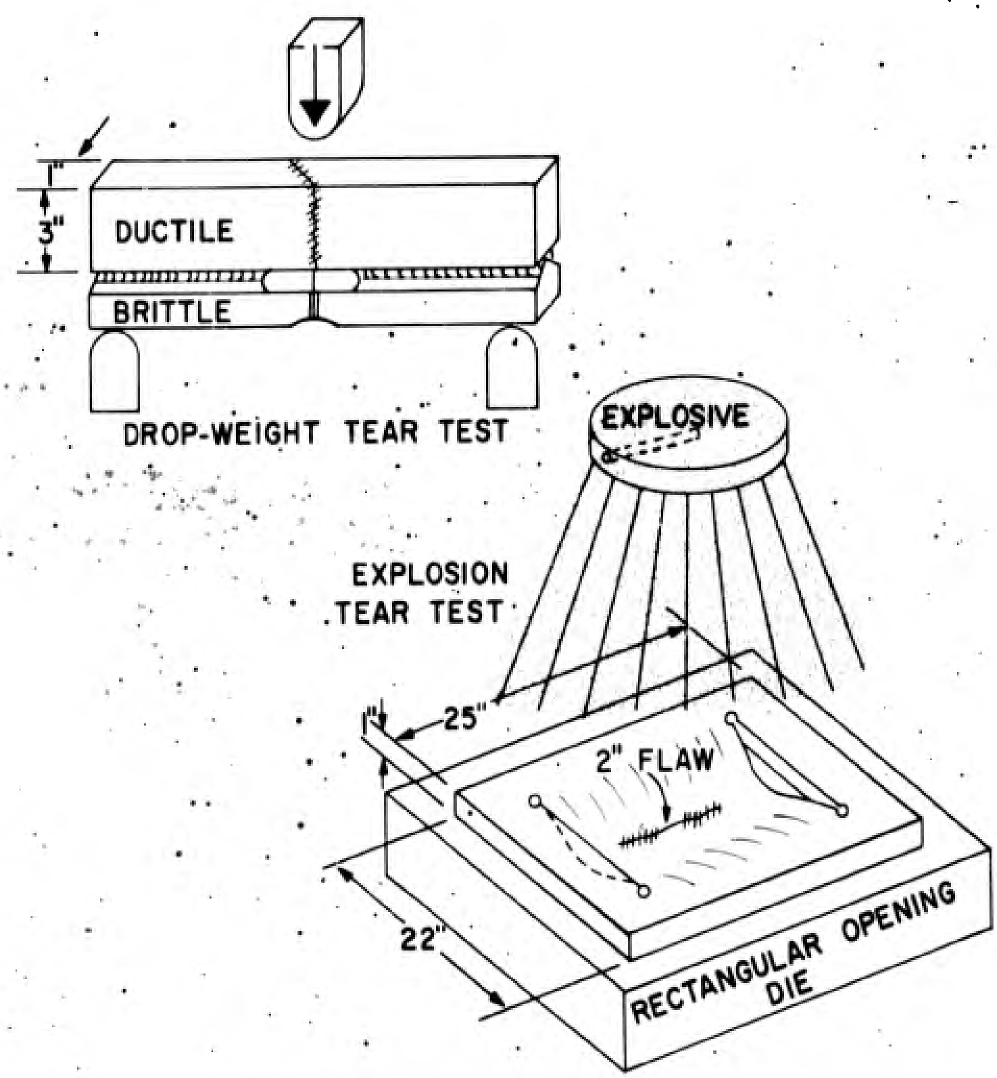

Fig. 12 - Features of drop-weight tear and explosion tear tests

*The 2-inch crack is developed by the constraint of a "patch" weld deposited in a centrally located 1/2 × 2-inch through-thickness slot. The brittle crack-starter electrode, Hardex N, is used for this weld, which is made without preheat and with allowance for cooling between weld passes.

Figure 13 illustrates the performance of HY-220 Maraging steel in the presence of the 2-inch crack. It is obvious that the fracture propagated at elastic levels of load – the plate remained perfectly flat. An opposite extreme is illustrated by a test of HY-80 plate, Fig. 14, which resulted in the propagation of a short, 1-inch-long tear for each of 5 successive load applications. In other words, even with the enlargement of the flaw to almost a 1-foot length and loading which approached the ultimate strength of the steel (deep bulge), fracture propagation is resisted by this steel.

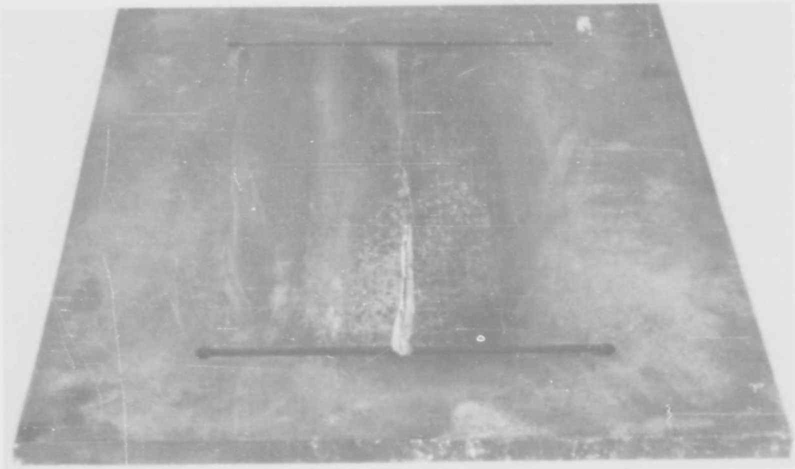

Fig. 13 - Explosion tear test "flat break" fracture, initiating from a 2-in. crack defect in a 1-in.-thick test plate of a 220 ksi Maraging steel of poor quality.

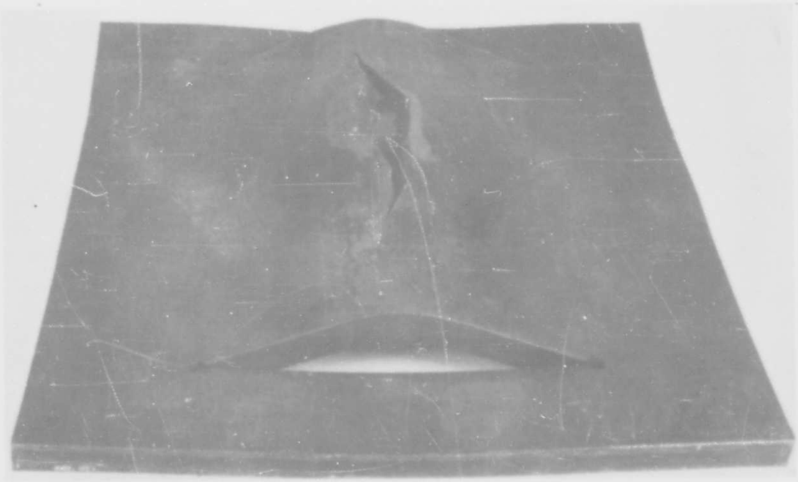

Fig. 14 - Explosion tear test of an HY-80 steel, illustrating limited tearing resulting from the successive application of 5 shots

Typical relationships between the C_V-SE of steels and the total length of tear obtained in the explosion tear test at 30°F are presented in Fig. 15. The point that relates to the "flat breaking" HY-220 steel is located at the top left corner of the diagram, i.e., full break (16 inches) at elastic levels of load. The "first shot" point for the HY-80 steel would be located in the grouping of points encompassed by the square frame, i.e., a 1-inch tear for a load application that resulted in 5% to 6% deformation. The following summary may be made of these data:

(1) Steels that are characterized by C_V-SE values of 50 ft-lb or more develop approximately 2 inches of tearing for load applications which result in 5% to 7% deformation.

(2) For these steels, increasing the level of deformation to 10% to 12% results in increasing the tear length by an amount that is related to the C_V-SE level. The higher the C_V level the less the tear length as noted by the "fan" representation.

(3) Steels that are characterized by C_V-SE values of less than 40 ft-lb develop extensive tearing at deformation levels of less than 5% to 7%. The slope of the curve suggests tear propagation at and below yield strength load levels, for steels of the lowest C_V-SE value.

In all materials which are not uniformly cross rolled in the mill, a great disparity can be expected between C_V-SE and orientation with respect to the principal rolling direction. Accordingly, practical considerations dictate that the "weak" direction of fracture resistance in a material is the controlling factor determining suitability. The data of Fig. 15, relate principally to tests in the "weak" direction of poorly cross-rolled materials.

It is recognized that a "tradeoff" exists between flaw size and deformation level. Thus, the described "fan" effect should be developed by an increase in flaw size as well as an increase in deformation levels. A schematic illustration of the flaw size effect is provided by Fig. 16. The basic effect is to move the "fan" to the left of the diagram and to expand its angular opening. The tear length of the steels of low fracture toughness should be greatly increased by an increase in flaw size. The predictions are based on theoretical considerations and observations of the extension of tears on second shot testing (4). It is obvious that extensions of these studies should now concentrate on exploring flaw size effects, particularly for steels of less than 50 ft-lb C_V-SE. From a practical point of view, ample demonstrations have been made of the high level of fracture toughness of steels that have shelf energies in excess of 50 ft-lb.

It should be noted that the above stated relationships relate tearing characteristics and Charpy V shelf values for a specific direction in the plate. A poorly cross-rolled material that develops say 80 ft-lb in one direction and 30 ft-lb in the other direction must be evaluated on the basis of properties in the "weak" direction. Extreme differences in the fracture resistance for the weak and strong directions may be readily demonstrated by the explosion tear test. For purposes of this paper, we shall illustrate the major effects of directionality by data developed using the drop-weight tear test, to be described in the following section.

DROP-WEIGHT TEAR TEST

The intrinsic property that determines the relative extent of tearing developed in the explosion tear test is the amount of energy that is absorbed in the tearing process. If a

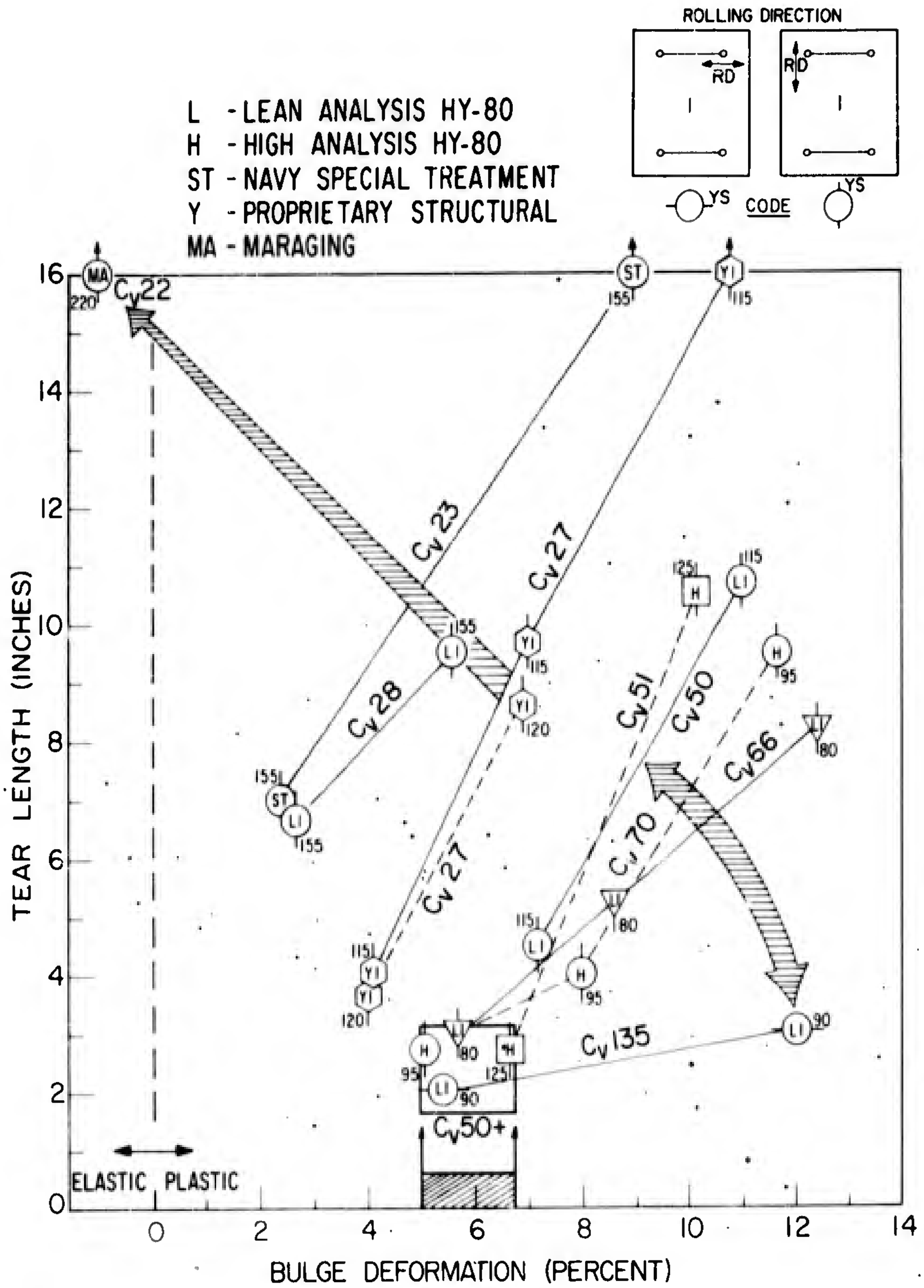

Fig. 15 - Total tear length (summation of the two ends of tear extensions) resulting from explosion loading to various levels of bulge deformation - C_V-SE of the steels are indicated. The connecting lines indicate tests of different plates of the same steel loaded to the indicated levels of deformation.

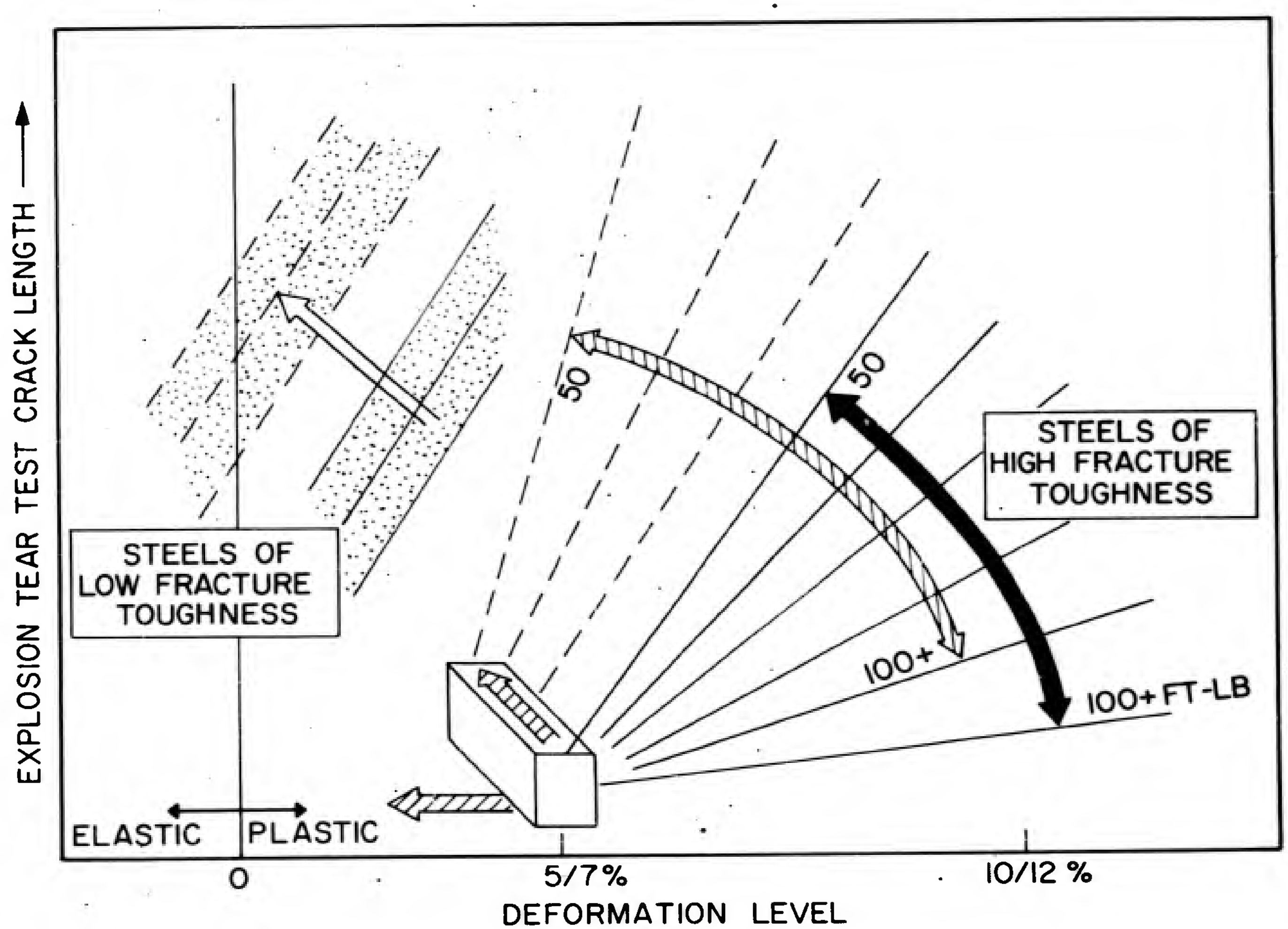

Fig. 16 - Illustrating the anticipated effects of increased flaw size on the explosion test tear lengths of the steels that fall in the "fan region" of Fig. 15, and of the steels that showed markedly lower levels of fracture toughness.

very small amount of energy is required to move the tear for a unit distance, extensive tearing should be expected as the result of loading to relatively low levels of plastic or high levels of elastic strain in the presence of a relatively small flaw. In essence, this is the basis for the correlations of flaw size and stress level relationships to the Charpy V shelf energy – specific levels of energy absorption in the Charpy V fracture of the steel translate (by correlation) to specific fracture toughness performance in the explosion tear test. The Charpy V test has the advantage of providing a "reference number" of tearing energy that can be determined generally by a standardized test procedure. Thus, the fracture toughness quality of a new steel may be assessed independently by any industrial laboratory aiming at the development of materials of Navy interest.

Because of the relatively small test sections offered by the Charpy test and the possibly large contribution of the initiation phase of fracture, related to the relatively "dull" notch of this test, it may be suspicioned that the C_V "reference number" tends to overrate the intrinsic tearing energy of materials with decreasing levels of fracture toughness. A search for a better tool for measuring the tearing energy of steels was therefore indicated, with the following aims:

(1) A full thickness test so as to integrate through-the-plate-thickness quality gradients.

(2) To minimize the amount of energy absorbed in initiation of the fracture to a closely similar, low level, for all steels tested.

(3) To provide for inexpensive "mass production" testing required for broad-field explorations and for specification use.

These aims were achieved with the design of the drop-weight tear test (Fig. 12).

The new test features a composite weldment, formed by joining a 1×3×18-inch test plate to a notched, brittle, cast steel bar. A full thickness weld approximately 4 inches long is deposited opposite the notch with a brittle electrode, Hardex N, to provide a continuous, brittle, crack path into the test plate. The rest of the weld joint consists of two root passes with conventional electrodes normally employed for welding of the test steels. Small steel blocks are tack-welded to one end of the specimen to facilitate vertical alignment on the anvil rounds. The specimen is loaded as a simple beam by the action of a falling weight. Several specimens, usually 3 to 4, are broken to establish the amount of energy required for complete fracture, i.e., by a "bracketing" technique.

Figure 17 shows a broken drop-weight test specimen which illustrates the use of a brittle steel, crack starting bar (brittle fracture region at bottom) which provides for initiation of the fracture with a reproducible energy expenditure of approximately 400 ft-lb. The remainder of the fracture, noted to be a 45 degree shear tear, represents the tearing of the 3-inch-long test plate portion of the particular test piece. In this case, the energy absorption for complete fracture was 7750 ft-lb. Of this amount only 400 ft-lb was used in initiation of the fracture process. This test, as standardized for testing of 1-inch plate, has provided an accurate and highly reproducible method for evaluating the tearing energy absorption of high strength steels. By increasing the size of the specimen and the size of the weight, full thickness tests may be conducted for 2- to 4-inch plates.

The general relationships of DWTT tearing energy values to explosion test performance are illustrated by Fig. 18. This figure is duplicate of Fig. 15, except that the individual curves are referenced to DWTT data obtained from tests at the same (30 °F) temperature as used for the explosion tests. It may be noted that the "fan" region of the diagram is bordered by DWTT values ranging from 3000 to 7750 ft-lb. The materials that developed extensive tearing at low levels of deformation are characterized by DWTT values of

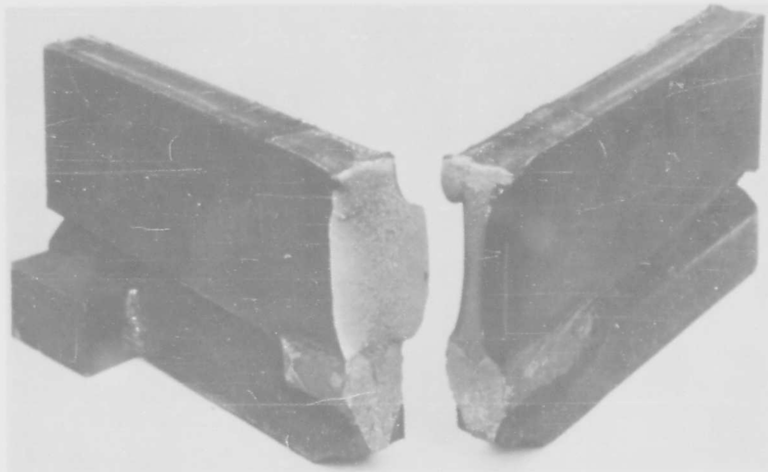

Fig. 17 - Fracture appearance of DWTT specimen representative of a material of high fracture toughness – HY-80 steel sample tested in the "strong" direction

1500 ft-lb or less. It is not possible, at this time, to evaluate the degree to which this test may provide a more accurate correlation with relationships between flaw size and stress than is provided by the Charpy test. Such an analysis must await the development of additional flaw-size test data, particularly for steels of low fracture toughness. Irrespective of such benefits, the test presently provides considerable advantages over the use of Charpy tests, as follows:

(1) May be used to evaluate plate thickness effects.

(2) May be used to compare the tearing energies of widely different materials prior to establishing Charpy V correlation bases.

(3) Less expensive and time consuming than the Charpy test (plate cutting to test cycle).

(4) Provides for evaluating materials at temperatures which entail a mixture of fracture modes. This is not possible for Charpy V tests at the present stage of correlation development.

The correlation of DWTT energy values with C_v-SE values (Fig. 19) is surprisingly good, as indicated by the data band which encompasses all relationships established to date. The data pertain solely to Q&T steels with the exception of the Maraging steels located at the extreme lower portion of the band. The wide range of tear energy values developed by the various steels in the 80 to 220 ksi yield strength range is apparent from the plot. The relative position of the various classes of steels is made evident by the repeat of the data in Figs. 20 to 23.

One of the important findings illustrated by these data is the wide differences in tearing energy that may be obtained in a given plate as a function of orientation. Figure 20 illustrates HY-80 (as received) steel representing highly cross-rolled and poorly cross-rolled material. The high level of tearing energy generally exhibited by HY-80 causes no

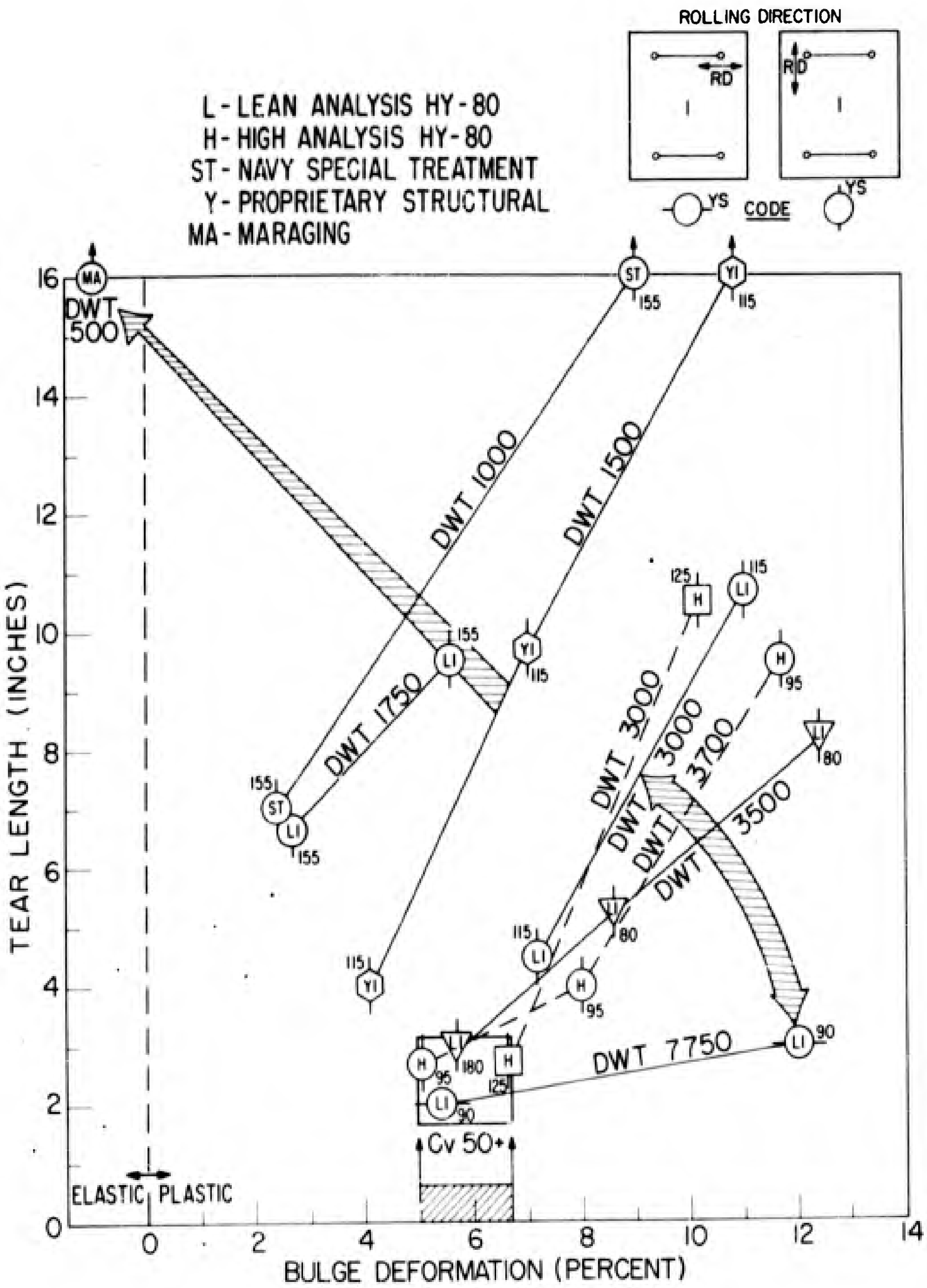

Fig. 18 - Explosion tear test data, previously presented in Fig. 15, referenced to the drop-weight tear test energy adsorption for tests conducted at the same (30°F) temperature as used for the explosion test

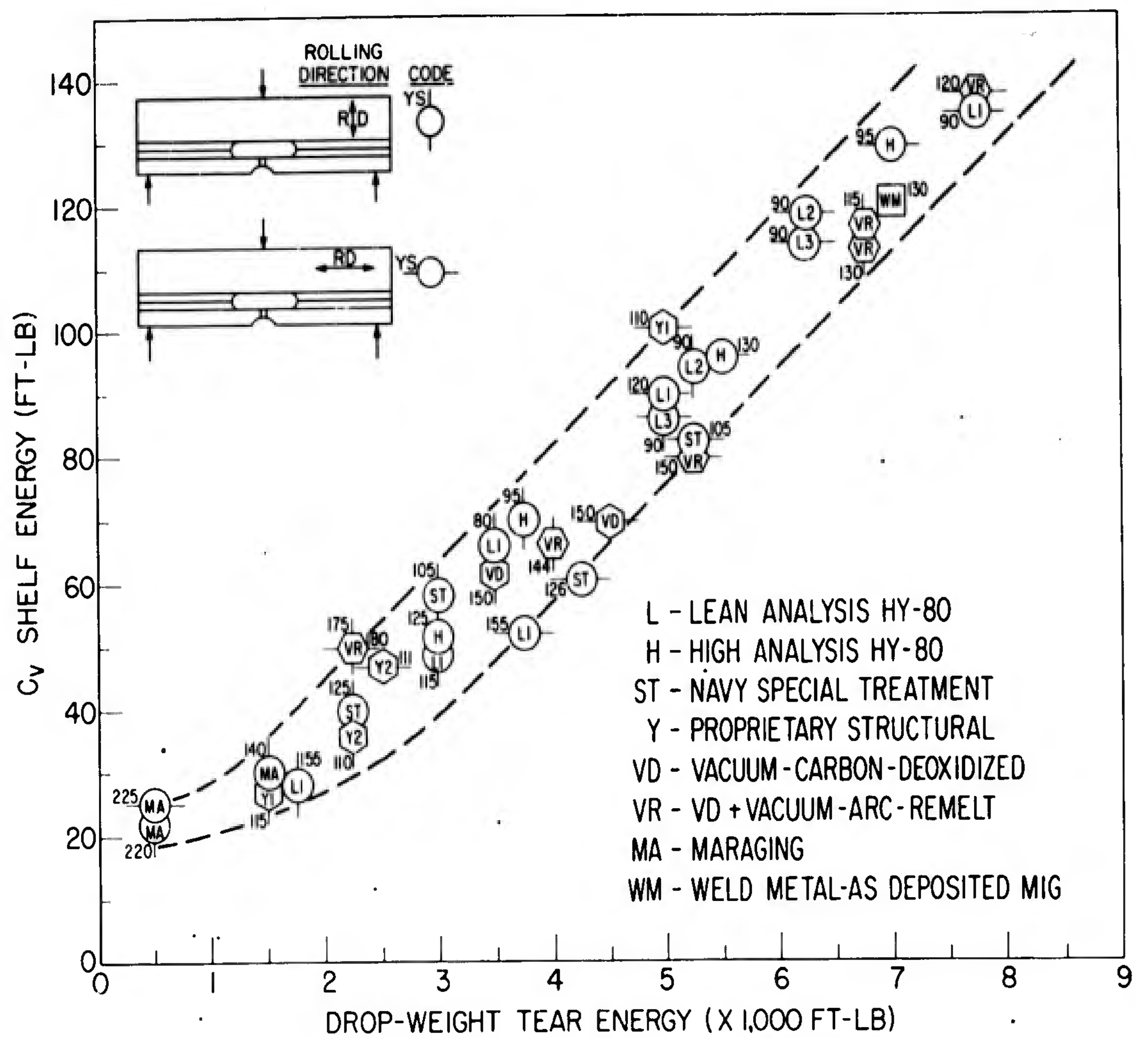

Fig. 19 - Correlation of DWTT energy absorption values with Charpy V test shelf energy values

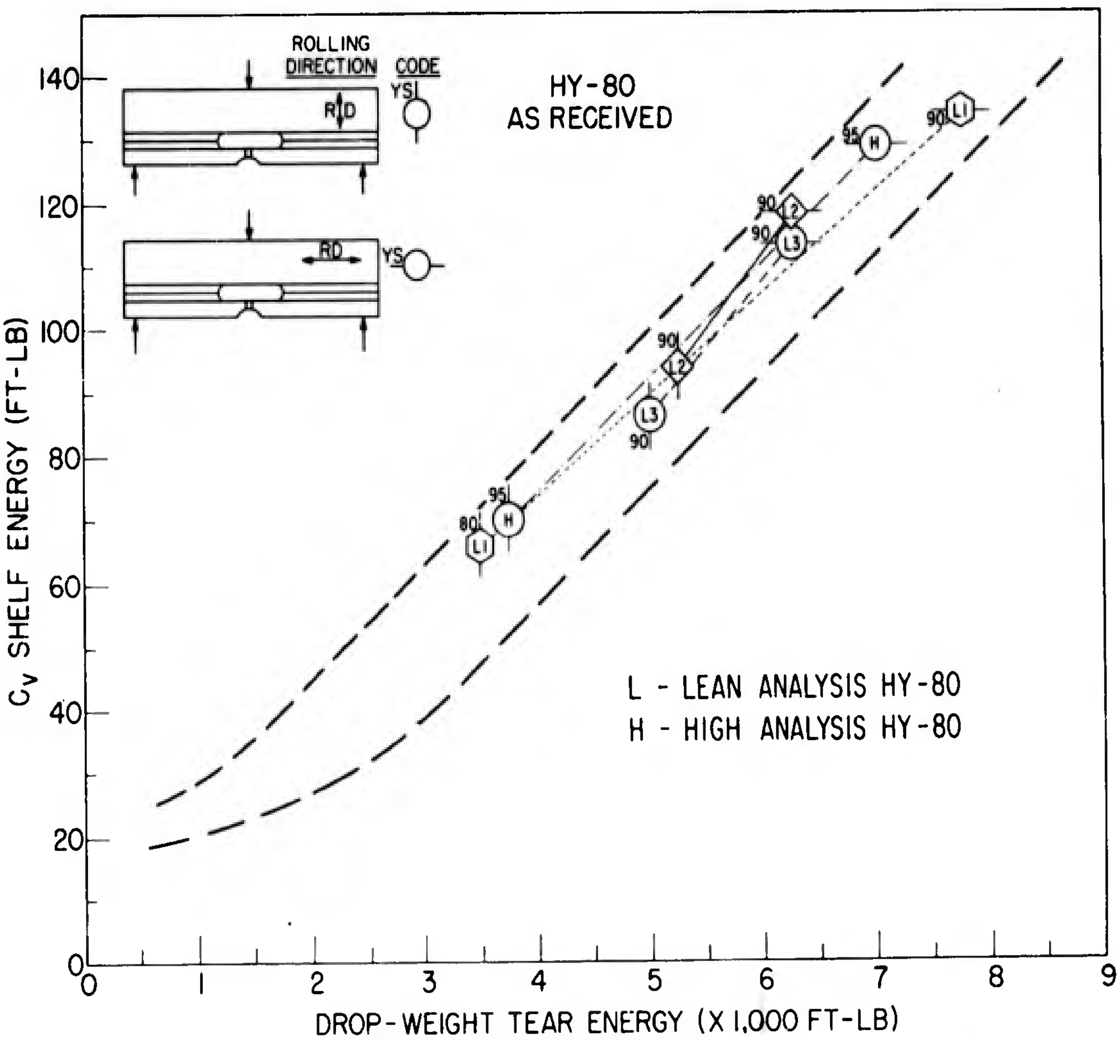

Fig. 20 - DWTT energy values of various HY-80 steel plates - representing highly cross-rolled and poorly cross-rolled materials

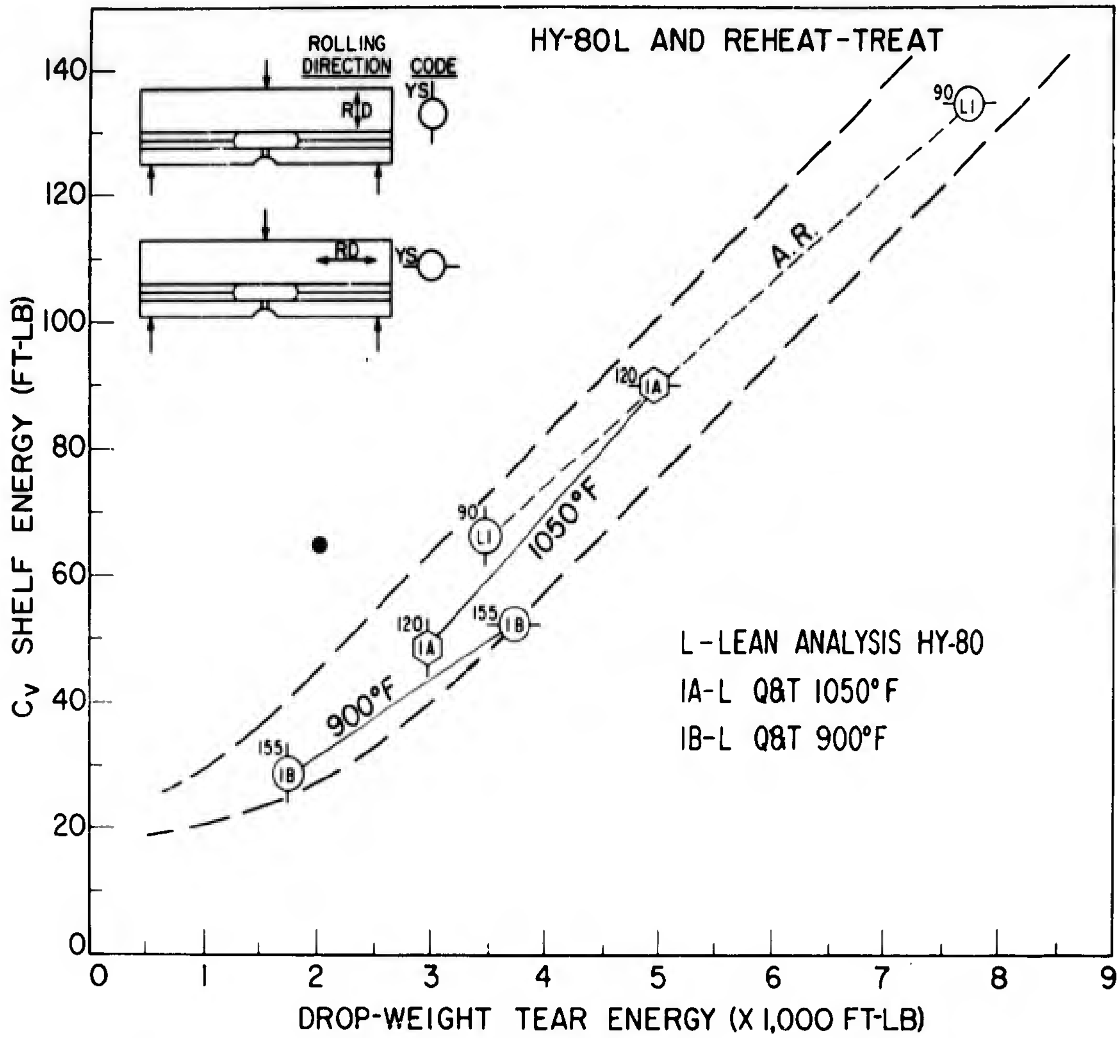

Fig. 21 - Illustrating the general effect of decreased tear energy with increase in strength level; data relate to a poorly cross-rolled HY-80L steel

concern in the use of poorly cross-rolled material, because even the "weak" direction is relatively highly resistant to tearing. Figure 21 illustrates the effects of heat treating the HY-80L (low chemistry range) composition to 120 to 150 ksi levels of strength. The general effect of decreasing tearing energy with increasing strength is such as to cause the "weak" direction of a poorly cross-rolled material to drop to low values of tearing energy at strength levels that would provide for relatively high values for the "weak" direction of highly cross-rolled material. The implications are that, as the strength level is increased, a point is reached such that a markedly "weak" direction cannot be tolerated.

Figure 22 presents a summary of data obtained to date for 100 to 130 ksi material, as coded. The location of a 130 ksi, as deposited, MIG weld metal is particularly interesting in illustrating the state of the art for this strength range. The sharpness of the fracture toughness level change with increasing strength level is indicated by the fact that all known, as deposited, 150 ksi weld metals are highly brittle. In other words, the state of the art is "top of the energy band" for the 130 ksi level and "bottom of the energy band" (expected) for the 150 ksi level.

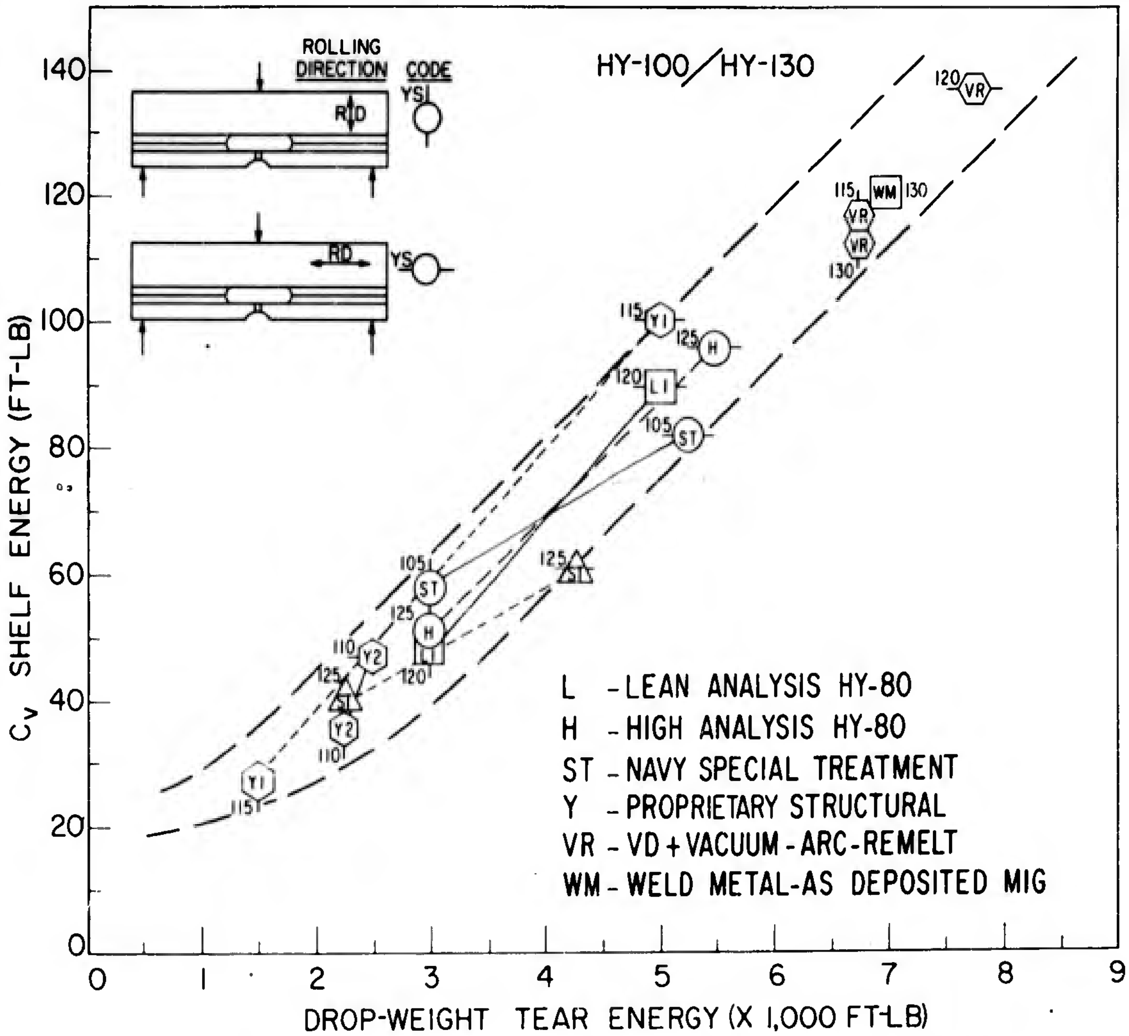

Fig. 22 - Illustrating the wide range of DWTT energy values developed by various steels of the 100 to 130 ksi yield strength range

Figure 23 presents a summary of available DWTT data obtained for steels of 150 to 220 ksi. Of particular interest are the data for the proprietary steel compositions designed for 150 ksi and 170 ksi strength levels. These steels were produced with special steel making practices, coded as VD (vacuum-carbon-deoxidation and air teemed) and VR (vacuum-carbon-deoxidation and consumable electrode remelt) processes. For these steels, deoxidation of the basic electric arc furnace heat was achieved under vacuum principally with carbon, in order to reduce the amount of nonmetallics normally retained in conventional steel making practice using silicon or manganese for deoxidation. The first-melt ingots of carbon-deoxidized steels, however, would be gassy and require special handling to produce sound, gas-free ingots. For the VD steel, the liquid metal was vacuum-carbon-deoxidized in ladle-to-ladle degassing equipment, and then teemed in air into one ingot. The VR steels involved forging of the first-melt (gassy) ingots into electrodes which were subsequently remelted in a vacuum consumable electrode arc furnace to produce high quality, exceptionally clean ingots which are then processed into plate products. One of the VR steels involved ultrahigh purity practice which was restricted to vacuum consumable electrode arc furnaces with the first-melt utilizing an electrode prepared by compacting very high purity powdered metals blended to the desired alloy chemistry. The chemical compositions of all test steels are listed in Table 1.

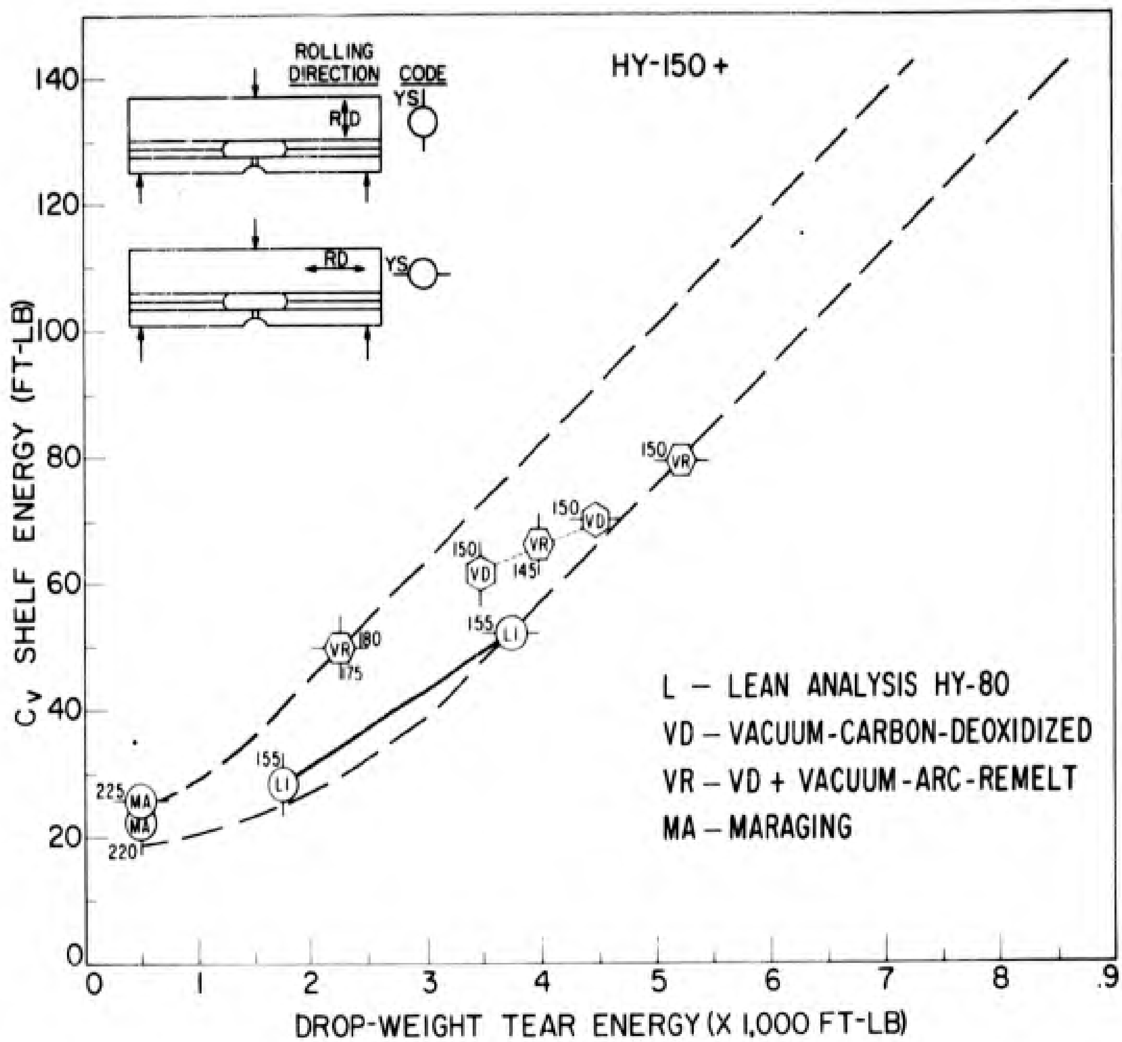

Fig. 23 - Illustrating the range of DWTT energy values developed by various Q&T steels of the 150 to 180 ksi yield strength range. Values obtained for the 220 to 225ksi yield strength Maraging steels are provided for comparison.

Table 1
Chemical Compositions of Test Steels (percent)

Element	HY-80(L)			HY-80(H)	STS
	(L-1)*	(L-2)*	(L-3)*	(H)*	(ST)*
C	0.16	0.15	0.15	0.18	0.27
Mn	0.31	0.20	0.20	0.28	0.20
Si	0.21	0.20	0.20	0.26	0.22
S	0.016	0.007	0.007	0.010	0.025
P	0.015	0.018	0.018	0.015	0.010
Ni	2.16	2.24	2.24	2.76	3.16
Cr	0.93	1.29	1.28	1.35	1.46
Mo	0.29	0.34	0.35	0.44	0.20

Element	Code Y		Special Melting Practice			Maraging
	(Y-1)*	(Y-2)*	(VD)*	(VR-1)*	(VR-2)*	(MA)*
C	0.16	0.17	0.14	0.26	0.26	0.04
Mn	0.80	0.75	0.30	0.13	0.13	0.03
Si	0.23	0.22	0.20	0.01	0.04	0.08
S	0.014	0.017	0.007	0.006	0.006	0.004
P	0.015	0.014	0.004	0.004	0.004	0.001
Ni	0.79	0.79	7.20	9.10	3.12	18.30
Cr	0.54	0.50	0.93	0.39	1.48	0.05
Mo	0.50	0.48	1.07	0.39	0.90	5.3
	0.21Cu	0.26Cu		4.1Co	0.09Co	7.4Co
	0.004B	0.004B				0.24Ti

*Designates steel number on illustrations.

It is apparent that the two types of metal treatment (consumable electrode and vacuum teem) result in steels of 150 ksi strength levels that have closely similar tear energy properties. Both of these materials were highly cross rolled; therefore, direct comparison cannot be made with the HY-80L steel that was not cross rolled to the same high level. A test of the value of special melt treatment, as well as of alloy modifications, will be made when tests are conducted for HY-80H steel that is highly cross rolled. The implications of these data are that expensive melt treatments may not be required for the 150 ksi strength level, insofar as plate properties are concerned.

The explosion tear test results for the "new" steels are presented in Fig. 24. The figure presents the outline of the "fan" defined previously by Figs. 15 and 18. The tear lengths developed by these steels are grouped closely to the "square frame" position of the "fan," i.e., are exactly as expected for the test conditions and for the reported C_V-SE and DWTT energy values. Additional tests of these steels at 10% to 12% levels of deformation are expected to demonstrate significant differences between the 150 ksi and 175 ksi strength level, as predicted by the "fan" relationships to C_V-SE and DWTT energy values.

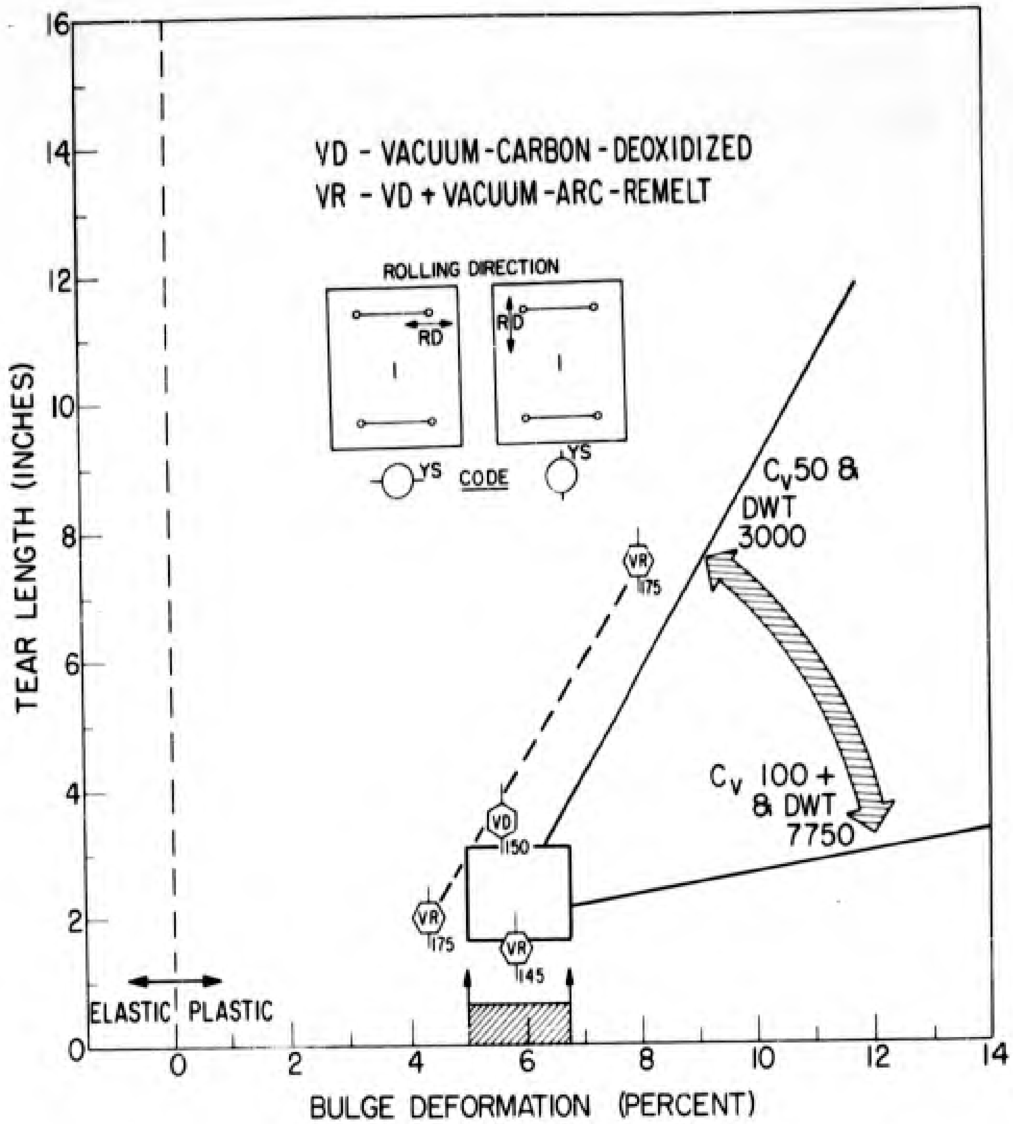

Fig. 24 - Explosion tear test data for the "new" 150 and 175 ksi steels prepared by special melting practices, compared with the "fan" diagram of Figs. 15 and 18

In effect, these data indicate a relatively high level of fracture toughness can be obtained for Q&T steel plates at a yield strength level of 150 ksi. While the relative resistance to deformation overloads, in the presence of large flaws, is not of the same order as that which may be obtained for Q&T steels of the 80 to 130 ksi strength range, there is no cause for concern. The practical problem is clearly not that of the plate material but of the weld and HAZ, i.e., it is a problem of fabricability. To date, no data have been provided with respect to the weldability of these steels.

SUMMARY

The physical metallurgy and mechanical properties of quenched and tempered steels are sufficiently well known to provide for the making of reasonable accurate analyses of the potentials of these steels for future submarine hull fabrication. In effect, this information provides a sound basis for justifiable optimism for attainment of certain strength levels and clearly defines strength levels which are realistically to be considered as unattainable. We may consider this family of materials as having attained a point of maturity such that possibilities for estimation excesses by opponents or proponents are effectively curbed by facts. A similar point of maturity has not been reached by competitive materials which may be considered to equal or exceed the capabilities of Q&T steels.

A broad analysis is presented of the characteristics of Q&T steels covering the strength range of 80 to 200+ ksi yield strength. The experience obtained in the fabrication and use of steels in this broad range of strength levels provides for accurate assessment of the problem of utilizing the 100 to 150 ksi yield strength range for submarine hull fabrication. The "facts of the case" have been presented by consideration of the physical and welding metallurgy characteristics of Q&T steels which apply to the total spectrum of attainable strength levels. This baseline analysis was expanded further by other analyses of the intimate relationships that exist between the metallurgical, mechanical, and service reliability characteristics of the steels over their spectrum of strength levels. Thus, questions of design and fabrication quality levels as well as those of fracture toughness and structural reliability become assessable and definable within a context of fabrication experiences that are outside those of submarine hull fabrication.

In brief, the subject analyses indicate that:

(1) HY-120 to HY-130 hull fabrication capability may be obtained by a development program which basically involves upgrading of the strength level of the electrodes. There is no basic problem in selecting a steel of adequate fracture toughness and weldability. The primary job involves the establishment of a suitable level of design and fabrication quality that is compatible to the strength level of the steel.

(2) HY-150 hull fabrication capability is a "ceiling aim" for Q&T steels. Before such capability can be claimed it is necessary to demonstrate the development of nonbrittle weld metal in the laboratory. In addition, it will be necessary to customize the fabrication of such steels to inherently critical welding controls which will require general use of automatic welding techniques. The development problem does not reside in the finding of a Q&T steel of this strength level which is uniquely insensitive to welding variables; such an aim may be justified only by alloy concepts that are presently unknown. In other words, this is a long range research question.

(3) The development of fabrication capability for Q&T steel strength levels in excess of HY-150 is presently outside of feasibility justification. In order to attain such capability, it is necessary to apply a full cycle heat treatment to the weldment.

(4) The controlling feature in the use of high strength Q&T steels for submarine hull fabrication is not the fracture toughness of the base plate. On this basis the maximum level of projected attainment may be placed at approximately the HY-180 level. The controlling features are the weld and HAZ properties.

(5) The best promise for the use of steels for hull fabrication at levels of 140 to 180 ksi lies in the development of an entirely new family of materials that are based on new metallurgical concepts of hardening. The "carbon-free" Maraging steels are an example of such new approaches. As for any new material, one should be cautious in proceeding with design commitments based on inadequate information. Similar caution should be exercised in considering fabrication attainments in fields such as the large, solid propellant booster program. The overriding premium that is placed on the "inert parts weight fraction" of such vehicles and the allowable costs for fabrication that are therefore acceptable, are in an entirely different feasibility "ballpark." These considerations presently dictate that Maraging steels of the 260 ksi strength level are acceptable for large booster casings, despite the fact that the flaw size for fracture initiation at yield point stress levels is in the order of 3/8 inch for a 3/4-inch-thick plate. The criteria of acceptability is that such flaw sizes are "inspectable" and that the service is "one shot." It is obvious that the same feasibility considerations do not apply to submarines.

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An analysis is given of the potentials and limitations of high strength, quenched and tempered (Q&T) steels for welded construction of submarine hulls. The experience that has been developed in the use of such steels in a wide variety of applications provides the essential background for preassessment of fabricability and design problems in hull construction, as a function of increasing strength level. With increasing strength level, there is a decrease in fracture toughness level which requires an upgrading of fabrication and design quality levels for purposes of maximizing structural

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