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Z. ZUDANS, PH.D.
Senior Vice President and Chief Operating Officer

September 22, 1982

Mr. E. Igne
Staff Engineer
U.S. Nuclear Regulatory Commission
Advisory Committee on Reactor Safeguards
Washington, D.C. 20555

Re: Vessel Fracture in PTS - CE Owners Groups Use of Pellini Fracture-Safe Design Diagrams, Letter from E. Igne dated September 2, 1982.

Dear El:

Pellini's Fracture-Safe Design Criteria, represented by Slide 6 attached to your letter, are largely empirical based on documented cases of in service brittle failure. The experience upon which Pellini design curves are based does not include any nuclear reactor pressure vessels.

The original development of Fracture-Safe Design diagram by Pellini [1] was associated with the increased use of high-strength quenched and tempered (Q&T) steels for fabrication of vessels. Upon recognition of the fact that the pressure vessel steel establishes a range of temperatures (not a specific temperature) within which fracture could occur (or not) depending on the flaw size and stress conditions of the structure [2], Pellini developed Fracture-Safe Design curves applicable to the low-strength steels. For these steels the primary fracture problem is the transition from ductile (shear) to brittle (cleavage) fracture with decreased temperature of metal. Pellini Fracture-Safe Design, is based on unification of the approach of fracture initiation prevention and the approach of preventing fracture propagation, Slide 6. Here, a generalized stress-NDT+temperature curves for both crack-arrest and fracture initiation are given. According to this curve, for temperature to the right of the crack-arrest-temperature (CAT) curve, brittle fractures are prevented by the "crack arrest" properties of the steel.

[1] W.S. Pellini and P.P. Puzak, "Fracture Analysis Diagram Procedures for the Fracture-Safe Engineering Design of Steel Structures," NRL Report 5920, March 15, 1963; also Welding Research Council Bulletin No. 88, May 1963.

[2] W.S. Pellini and P.P. Puzak, "Practical Considerations in Applying Laboratory Fracture Test Criteria to the Fracture-Safe Design of Pressure Vessels," Pressure Vessels and Piping: Design and Analysis, Vol. 3, Materials and Fabrication, pp. 124-139, 1976.

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Mr. E. Igne
ACRS

- 2 -

September 22, 1982

As shown in [1,2], the data base for Failure Assessment Diagram (FAD) does not include nuclear reactor pressure vessels. Accordingly, the direct use of FAD for nuclear pressure vessel assessment requires caution and careful assessment of the FAD, Slide 6, data base. The approach, however, is quite attractive inasmuch as it requires no detailed stress or fracture mechanics analysis. Certainly *FAD represents a useful screening guide for deciding which cases one should subject to detail fracture mechanics analyses* for establishment of more realistic limits of RT_{NDT} .

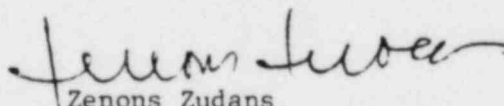
Since FAD's empirical criteria are derived from actual experience (albeit non-nuclear pressure vessels*) it should also provide good benchmarking for validation of detailed fracture mechanics computed RT_{NDT} , (if applied to cases sampled from FAD data base).

Based on the above discussion, I do not believe that FAD curves as given in [1,2] can be used for the demonstration of crack arrest as suggested in C-E analysis attached to your above letter.

I have enclosed Reference 2 for your detailed information. Also, I hope to get Reference 1 (which contains the FAD, Slide 6, data base information) prior to PTS meeting on September 28.

Note that Pellini FAD diagram is also discussed in NUREG/CR-2837 (PNL-4327), pages 6.8 to 6.13. In that discussion more recent work [3] by Pellini on thick sections is presented. I am in the process of obtaining Reference 3 and will report on it later.

Very truly yours,



Zenons Zudans
Senior Vice President and
Chief Operating Officer

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encls.

cc: M. Bender
P. G. Shewmon

*In a nuclear vessel wall temperature and NDT increase through the wall from inside to outside, the stress decreases from inside to outside.

[3] Pellini, W.S., Principals of Structural Integrity Technology, Office of Naval Research, Arlington, Virginia, 1976.

Practical Considerations in Applying Laboratory Fracture Test Criteria to the Fracture-Safe Design of Pressure Vessels

Trends in pressure vessel applications involving higher pressures, lower service temperatures, thicker walls, new materials, and cyclic loading require the development of new bases in the supporting scientific and technological areas. This report presents a "broad look" analysis of the opportunities to apply new scientific approaches to fracture-safe design in pressure vessels and of the new problems that have arisen in connection with the utilization of higher-strength steels. These opportunities follow from the development of the fracture analysis diagram which depicts the relationships of flaw size and stress level for fracture in the transition range of steels which have well-defined transition temperature features. The reference criteria for the use of the fracture analysis diagram is the NDT temperature of the steel, as determined directly by the drop-weight test or indirectly by correlation with the Charpy V test. Potential difficulties in the correlation use of the Charpy V test are deduced to require engineering interpretation of Charpy V test data rather than to involve basic barriers to the use of the test. The rapid extension of pressure vessel fabrication to Q&T steels is expected to provide new problems of fracture-safe design. These derive from the susceptibilities of steels within this family to tear fractures of low energy absorption. This fracture mode does not involve a transition temperature and is therefore relatively independent of temperature. It is emphasized that such susceptibilities are not inherent to the family of Q&T steels of low and intermediate strength levels, but are related to specific metallurgical conditions of the plate and particularly the HAZ (heat-affected-zone) regions of Q&T steel weldments.

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Introduction

As a preface to the presentation of this paper, it is fitting to reflect on the technological trends in the fabrication of pressure vessels and on the evolution of technical knowledge for fracture-safe design. A decade ago the technology was based almost entirely on the fabrication of vessels constructed of low-strength, as-rolled, or normalized steels for low-pressure applications. Williams [1]¹ reports that more pressure vessels for the pressure range 1500 psi to 15,000 psi have been fabricated in the past few years than in the past two decades. A related trend

is the increasing use of high-strength quenched and tempered (Q&T) steels for the fabrication of vessels for both low and high pressures. A decade ago there was no fracture-safe design approach—in its place the safety of pressure vessels was predicted entirely on fabrication and service experience, intuition, and conservatism. The general acceptance of the concept that no laboratory notch test could be developed or adopted for "determining the transition temperature of a structure" cast a pessimistic outlook on the possibilities of developing a practical fracture-safe design approach. This oversimplification completely missed the fact that the steel does not establish a specific transition temperature for a structure (i.e., a fracture temperature) but a range of temperatures within which fracture could occur (or not) depending on the flaw size and stress conditions of the structure. In effect, the laboratory notch specimens were expected to perform an obviously impossible task. Such thinking clearly reflects on the state of technical knowledge of the time. Today there is a greatly improved understanding of these aspects, and practical fracture-safe design principles have been developed

¹ Numbers in brackets designate References at end of paper.
Contributed by the Research Committee on the Behavior of Pressure Vessel Materials and presented at the Winter Annual Meeting, Philadelphia, Pa., November 17-22, 1963, of THE AMERICAN SOCIETY OF MECHANICAL ENGINEERS. Manuscript received at ASME Headquarters, March 12, 1964.

[2]. These principles confirm past experience and clarify the premium derived from fabrication, design, and inspection quality. As such, they provide opportunities for increasing design efficiencies and decreasing metal costs by reduction of unnecessarily high factors of safety deriving from uncertainty.

These advances represent the start rather than the end of the development of fracture-safe design principles. They are applicable to steels of relative metallurgical and fracture simplicity, i.e., low-strength steels for which the primary fracture problem is inherent to their transition with decreased temperature from ductile (shear) to brittle (cleavage) fracture. While we have advanced in these respects, we are far behind in understanding and in the development of similar practical procedures for the more complex steels of the high-strength class. These steels are susceptible to fracture of a noncleavage type in the plates, welds, and heat-affected-zones (HAZ) at moderate temperatures, as well as to cleavage brittleness at low temperatures. Thus, the general outlook for the immediate future is that technical success in one area has been countered by the rapid entry into new areas for which the available technical knowledge is grossly inadequate. Moreover, some of the experience and intuitive background which provided broad-front engineering design guidance for the low-strength steels is not pertinent to the new class of high-strength Q&T steels. A minor example is the general insistence for stress relieving of welded Q&T steel vessels intended for service at temperatures above the transition range of the steel. This follows from past experience that stress relief was beneficial for pressure vessels of the low-strength steels used at temperatures below their transition. It represents misplaced intuition that past experience applies generally to all new steels. For some of these Q&T steels, stress relief is actually detrimental to the fracture toughness of plates, welds, and HAZ. Thus, we are coping with new and relatively unknown problem areas while carrying a burden of unwarranted practices which derive from inapplicable past experience.

These introductory generalizations may be tied to reality by consideration of a spectrum of potential fracture characteristics of pressure vessels illustrated in Figs. 1 to 5. These illustrations dramatize characteristically different failure modes which may be related to the fracture surface appearances and, in turn, to the type of steel and method of fabrication.

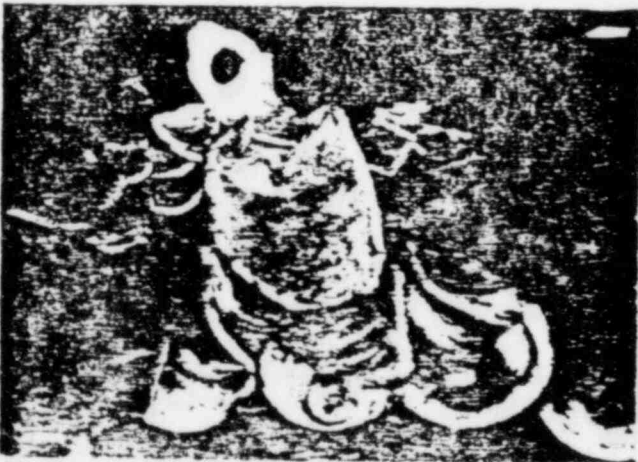


Fig. 1 Fragmentation fracture of a seamless tube, high pressure, compressed-air flask developed at a temperature 10 deg F below the NDT of the steel. The failure occurred under pneumatic loading during a routine test for air-leaks.

Fig. 1 illustrates the "shattering" expected for pneumatically loaded vessels which fracture at temperatures below or near the nil-ductility transition (NDT) temperature of the steel. At these temperatures the fracture surfaces display the classical "square break" (no appreciable shear lip) and the cleavage appearance normally associated with brittle fractures of conventional, low-strength steels. In contrast, the failure of hydrostatically loaded vessels at temperatures below or near the NDT results in fractures involving a few large pieces, Fig. 2. The fracture surface appearance is identical to that of the pneumatic load failures.



Fig. 2 Hydrostatic test fracture of a welded and stress relieved pressure vessel developed at a temperature 10 deg F below the NDT of the steel

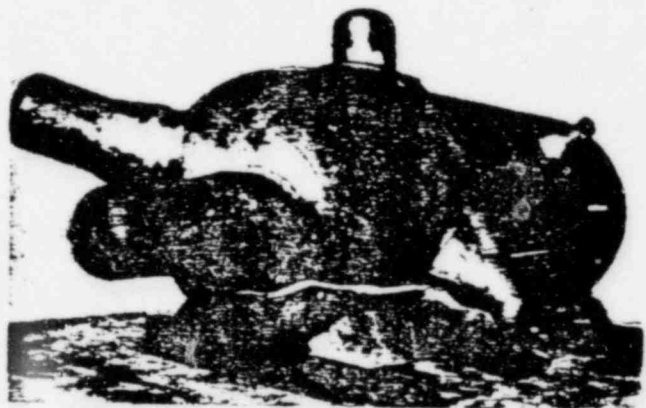


Fig. 3 Hydrostatic test fracture of a welded and stress relieved pressure vessel at a temperature of 50 to 60 deg F above the NDT of the nozzle and plate material. The arrow indicates the location of a 6-in. fatigue crack which initiated the fracture during a low cycle fatigue test.

Fig. 3 illustrates the nature of brittle fractures at temperatures in the order of 50 to 60 deg F above the NDT temperature of the shell plate. At these temperatures the fracture surface is similar to the failures in Figs. 1 and 2, except for 1/8-in. to 1/16-in. shear lip borders at the plate surfaces. The increased energy absorption provided by the shear lips prevents multiple forking of the fracture and limits it to a single or few branches in the direction normal to the principal applied stress. The fractured vessel hangs together if the internal loading is hydrostatic, or

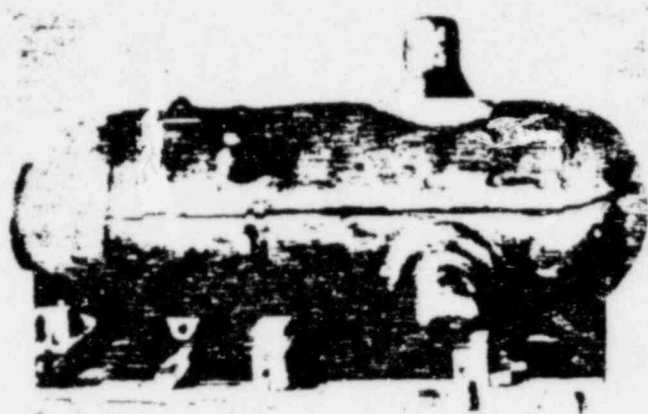


Fig. 4 HAZ "low-energy tear" fracture of a welded and stress relieved Q&T steel pressure vessel resulting from high hydrostatic test pressurization in the presence of a fatigue induced crack



Fig. 5 "Low energy tear" hydrotest fracture of a welded and full Q&T heat-treated rocket case. The fracture initiated from a flaw in the circumferential weld and then propagated entirely in the plate (sheet) material.

separates into a few large chunks if the loading is pneumatic. Such fractures have been relatively rare because of the unlikely combination of large flaws and high stresses required for fracture initiation at temperatures of 50 to 60 deg F above the NDT.

Fig. 4 illustrates a fracture which, at first glance, resembles the Fig. 3 type, but which is characteristically different in that the fracture path followed the weld HAZ. Such fractures may occur in Q&T steel vessels which feature high plate fracture toughness. This behavior is not inherently characteristic of Q&T steels as a family, but derives from improper (specific to the steel) welding conditions or stress relief heat-treatments which degrade the HAZ fracture toughness. In this case the fracture surfaces do not involve cleavage but a form of low energy absorption tearing, to be discussed.

Fig. 5 illustrates the fracture of a rocket case fabricated of ultrahigh-strength (190 ksi yield strength) Q&T steel characterized by low energy absorption tearing in the sheet material. From a metallurgical and fracture point of view, the failure mode is the same as that of Fig. 4, the only difference is that the fracture path is entirely located in regions removed from welds or HAZ. Similar performance may be expected of thick-walled pressure vessels constructed of steels of equivalently low fracture toughness. An early discussion of the low energy tear problem was presented by the authors in 1956 [3], and later corroborated by service failures of heavy steel forgings at elevated temperatures [4].

The first sections of this paper have been structured to present a review of the principles of fracture-safe design for the transition temperature fracture mode, i.e., the classical "brittle fracture" problem. Following sections present discussions of limitations to the immediate extension of these principles based on Charpy V test criteria (correlations to the NDT) to all types of steels which are potentially subject to brittle fracture. These limitations arise from complications in the interpretation of Charpy V (C_V) data for a wide variety of steels. A summary of available NDT data for steels of pressure vessel interest is provided also for reference purposes. The final section presents discussions of the fracture problem of pressure vessels constructed

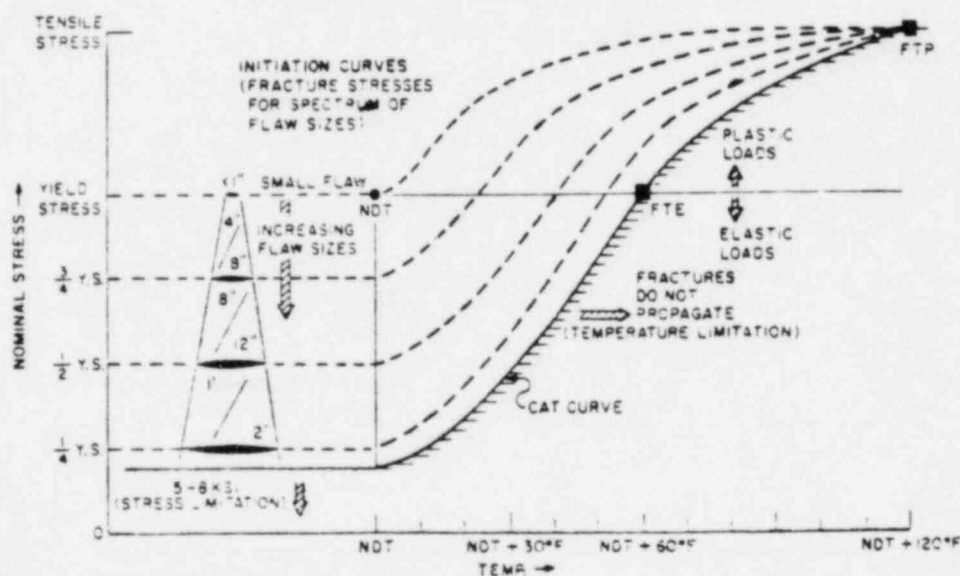


Fig. 6 Generalized fracture analysis diagram, as referenced by the NDT temperature

of steels which are potentially subject to both brittle fracture and to low energy tear fracture, depending on the temperature of service.

Significance and Practical Use of the Fracture Analysis Diagram

The origin of the "classical" brittle fracture problem is the decrease in fracture toughness resulting from a change in fracture mode from high energy absorption ductile tearing to low energy absorption cleavage fracture in a rather narrow range of temperatures. The full span of the transition is in the order of 120 deg F. However, the transition span from an "intermediate" to a "very low" level of fracture toughness, which is the transition range of engineering interest for conventionally loaded structures, occurs over a range of 10 to 60 deg F depending on the stress level. For example, the World War II ships (which represent examples of relatively low stress loading) fractured extensively and in large number at temperatures of 30 to 50 deg F, but not at all at temperatures above 70 deg F. The "sharpness" of the transition effect provided opportunities for devising laboratory test methods for identification of the transition temperature range of specific steels and thereby for application of this information to engineering design.

Fracture-safe design of steel structures may be based on preventing fracture initiation or on preventing fracture propagation. Until recently there was considerable contention as to the relative merits of the two approaches. The "fracture analysis diagram," Fig 6, serves to unify these approaches into a coherent analytical scheme² as illustrated by the generalized stress-temperature curves for both crack-arrest and fracture initiation. For temperatures above the crack-arrest-temperature (CAT) curve, brittle fractures are indicated to be prevented by the "crack-arrest" properties of the steel. This signifies a degree of fracture toughness sufficient to arrest the propagation of a brittle fracture translated through a brittle plate welded to a "test" steel. The CAT approach avoids questions of flaw-size and stress requirements for fracture initiation by restricting the use of steels to temperatures or to stress levels of "no propagation." This simplicity was particularly appealing prior to the development of adequate relationships of flaw size, stress, and temperature represented by the family of fracture initiation curves of the fracture analysis diagram.

The CAT approach has the advantage that it requires a minimum of engineering analysis and the disadvantage that it "forces" the selection of more expensive steels of 30 to 60 deg F lower transition temperature than actually required for many applications. The fracture analysis diagram makes it evident that the justification for the use of the CAT curve in design relates to conditions involving the presence of extremely large flaws. The flaw conditions to be expected in pressure vessels cover a wide gamut depending on fabrication quality and service conditions. Moreover, the influence of such factors as stress relief, fatigue crack growth, and wall thickness have important bearings which can be analyzed to economic advantage. With this point of view the CAT approach may be recognized more properly as a "backstop" procedure to be applied when engineering analysis dictates the use of nonpropagating steels

² A detailed discussion of the fracture analysis diagram and its validation has been presented previously by the authors (2) "Fracture Analysis Diagram Procedures for the Fracture-Safe Engineering Design of Steel Structures." Prior reading of this report is assumed for the purposes of the present discussion.

rather than as a uniquely separate method which avoids the need for engineering analysis of flaw-size conditions.

The use of the diagram depends on the accurate determination of the NDT temperature. This may be accomplished directly by the standard drop-weight test method (5, 6), or indirectly by the C_T test. The usefulness and limitations of the C_T test for such correlation purposes requires clarification. There is a considerable danger that misapplications will result for steels which have not been subjected to correlation studies.

The fracture analysis diagram provides a frame of reference for engineering analysis and for the application of intelligent engineering judgment. The family of fracture initiation curves direct attention to the "quality" aspects of the vessel with respect to possible flaw sizes and also to the effective levels of stress acting on the flaws. The increase in the stress level of the fracture initiation curves with increasing temperature above the NDT directs attention to the transition temperature of the steel. The following factors must be considered in the analysis and judgment process:

(a) What is the largest flaw size expected in the vessel as fabricated? The answer to this question involves judgment based on the quality of inspection and of the control procedures used during fabrication.

(b) What is the level of stress acting on the flaws considered to be present at various locations? For flaws located in the smooth cylindrical areas of the shell, the stress level is indicated by a suitable vector of the $PD/2t$ design stress. For flaws located at nozzle openings, the effective stress level will vary with design quality. In the absence of stress relief, small flaws located in the weld regions should be considered to be subjected to near yield level residual stresses.

(c) What contributions to flaw size enlargement are expected due to cyclic (fatigue) loading? The answer to this question requires consideration of prior deductions of items (a) and (b) for positions of potential fatigue action. A small flaw that is located, or developed, in the region of plastic stress loading should be expected to result in low cycle fatigue growth to a size that is established by the thickness of the section, if fracture does not terminate the growth process.

From a fracture analysis point of view the primary element of concern is not necessarily the $PD/2t$ design stress but the limiting (most severe) flaw size and stress combination of the vessel as a whole. In the absence of stress relief, small flaws located in zones of weld residual stress should be considered as the limiting combination of flaw size and stress for temperatures below the NDT. For this combination, fractures may develop for $PD/2t$ stresses of extremely low or zero value (as for the case of spontaneous fracture) because the additional general level of "locked in" stresses may provide for propagation, i.e., may exceed the 5 to 8 ksi minimum required by the CAT curve for propagation at temperatures below the NDT. This limiting combination may be eliminated from consideration by the use of stress relief. Creep flow relaxation at stress relief temperatures results in decreasing the level of stress in the weld zones to levels that are too low for fracture initiation due to the small flaws. If stress relief cannot be applied, as for large vessels and for field erection, recourse may be made to limiting the service to temperatures above the NDT. This approach is based on the fact that the initiation curve for small flaws rises sharply at temperatures above the NDT to levels of stress which correspond to the development of gross plastic deformation. This event cannot result from the addition of residual and load stress systems but re-

quires the application of plastic overloads, i.e., very high $PD/2t$ stress levels. There are two practical methods for assuring that the service temperature does not fall below the NDT: insulating and heating the vessel for the steels of high NDT, and the selection of a steel of NDT below the lowest anticipated service temperature.

It has been proposed that safe performance to temperatures below the NDT may be assured in the absence of thermal stress relief by the application of "hot" pretests, that is, by loading at temperatures near or above the CAT curve. Depending on the transition temperature of the steel this "hot" pretest could be cold by atmospheric temperature standards. The intent of "hot" pretests is to cause the small flaws to yield at the flaw tip, thereby resulting in a local relief of residual stresses and a blunting of the crack tips. This procedure effectively places the residual stress level acting on the flaws to some low level which requires the presence of relatively large flaws for fracture initiation. The method has been shown to be effective experimentally [7]. It is particularly applicable to pressure vessels because of the invariant direction of loading. It is not applicable to structures which involve a spectrum of service load directions which cannot be reproduced in "hot" pretests.

Safe operation below the NDT, based on stress relief or "hot" pretest protection requires that there be no flaw size and stress combinations of a critical nature other than the described small flaw and residual stress combination. For example, nozzles which may be stressed in the range of $1/3$ yield stress to yield stress levels (depending on design detail) must not contain or develop (fatigue) flaws which range from the 1-ft to the <1 -in. sizes, respectively. The dependence of fracture safety on the design and fabrication quality of nozzles is immediately apparent. A poorly designed nozzle which develops yielding at normal $PD/2t$ shell stresses is subject to fracture initiation due to the presence of a very small crack if the service temperature falls below the NDT of the nozzle. In fact the jeopardy is twofold, because such levels of stress may be expected to induce low cycle fatigue cracking. This process is then quickly terminated at first signs of fatigue cracking by catastrophic terminal fracture of the vessel, provided the shell CAT (NDT + 30 deg F) exceeds the service temperature, that is, provided the shell does not have crack-arrest properties at the temperature involved.

The most severe service conditions for a pressure vessel involve the combination of high $PD/2t$ stress levels and cyclic loading. Safety may be provided for such conditions by the "leak before fracture" approach. For purposes of discussion, let us assume that the stress level of the nozzle at high $PD/2t$ shell stresses is at or above yielding. This condition would be difficult to eliminate in practice even with the best of nozzle designs. Leak before fracture can be assured if the flaw size at the time of penetration to the exterior of the nozzle is less than the critical size for fracture initiation at stresses equal to or exceeding yielding. It is apparent that the flaw size at leakage is directly related to the wall thickness of the pressure vessel. Thick walled vessels with large nozzles could develop very large flaws, and conversely thin walled vessels could develop only small flaws. The fracture analysis diagram illustrates that the flaw size for fracture initiation at yield stress levels increases from approximately 8 in., to 1 ft, to over 2 ft for respective temperatures of 20, 40, and 60 deg F above the NDT temperature. As noted previously, the critical flaw size for the same level of stress at temperatures equal to and less than NDT is less than 1 in. Depending on the wall thickness of the pressure vessel,

leak before fracture performance may be designed by selection of steels with NDT temperatures of 20, 40, or 60 deg F below the lowest service temperature. Conversely, temperature control could be applied by heating of the vessel to accomplish similar results.

Pneumatically loaded pressure vessels containing very large flaws provide an interesting example of interpretation of the fracture analysis diagram. The previous report [2] presented a detailed discussion of burst tests of pressure vessels provided with external slit-flaws of sufficient length and depth to result in bulging of the flaw area. These tests demonstrated that fractures were obtained at temperatures above the CAT with pressurization equivalent to conventional $PD/2t$ shell stresses. The bulge which results from the loading of a pressure vessel containing a large flaw indicates that the significant level of stress is not that given by the $PD/2t$ calculation but that of the plastically deformed bulge region. Fracture initiation resulting from the development of bulging may be expected for pressurization by hydrostatic or pneumatic means; however, the propagation of the fracture at temperatures above the CAT is quickly arrested for hydrostatically loaded vessels. The reason for this is the almost instantaneous drop in pressure resulting from the loss of a relatively small amount of the highly incompressible fluid. Pneumatic loading results in complete fracture irrespective of the temperature above the CAT because the plastic instability is "self aggravating" after a rupture is started and the length of flaw is enlarged. This results from the fact that the highly compressible fluid exerts a continuing "bulging" pressure during the course of the propagation.

Additional examples of fracture propagation at temperatures above the CAT are provided by burst tests of two seamless tube air flasks. Details of the vessels and the sharply notched, machined slits are illustrated in Fig. 7. Both vessels were constructed of a Q&T chromium molybdenum steel of the ASTM A336 F22 type, as follows:

Flask no.	Chemical composition								
	C	Mn	Si	P	S	Ni	Cr	Mo	
E71	0.12	0.42	0.24	0.006	0.010	0.07	2.2	0.99	
E75	0.13	0.41	0.26	0.008	0.010	0.18	2.3	1.05	

Flask no.	Tension test data			
	0.505-in.—dia tension test data			
	0.2% Y.S. (ksi)	T.S. (ksi)	El. in 2 in. (%)	R.A. (%)
E71	84.3	100.9	22.5	75.8
E75	96.3	111.8	20.8	74.4

The burst tests were conducted at 48 deg F (E71) and 55 deg F (E75) and resulted in totally different fracture modes, as illustrated in Fig. 8. The first step in the fracture process for both vessels involved the development of an elliptical bulge over the length of the flaw and the tensile "neck" rupturing of the bottom of the slit. At this point, the bulge regions "opened" and the flaw ends were subjected to very high plastic load stresses. The rupture of flask E75 started and propagated in its entirety as a 45 deg shear tear. That of flask E71 started and propagated as a brittle fracture with $1/8$ -in. shear lips, as illustrated schematically in Fig. 7.

The fracture analysis diagram predictions for the modes of fracture, fracture surfaces, and effective level of burst stresses are indicated in Fig. 9. The $PD/2t$ stress levels at burst merely

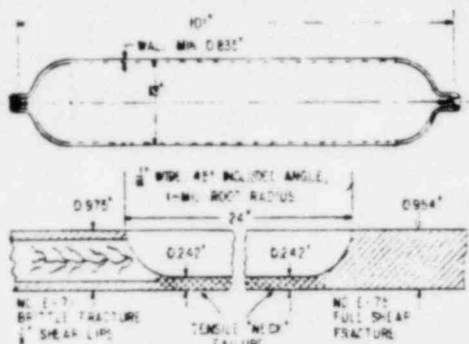


Fig. 7 Features of deliberately flawed pressure vessels used for pneumatic load burst tests



Fig. 8 Illustrating the general mode of failure of pneumatically loaded pressure vessels tested in the presence of large flaws. Top: brittle fracture with $1/2$ -in. shear lips developed by flask E71. Bottom: full 45-deg shear rupture developed by flask E75.

indicate the stresses required to develop the tensile neck ruptures and are of the same order for the two flasks. The NDT of the E75 flask was -70 deg F, which is 125 deg F below the burst test temperature. Accordingly, the fracture analysis diagram predicts that flask E75 should be expected to fracture in shear mode with an effective stress level in the order of the ultimate tensile strength of the steel. The NDT of the E71 flask was 0 deg F, which is approximately 50 deg F below its burst test temperature. Accordingly, this flask should be expected to fracture in a brittle manner with heavy shear lips and with an effective burst stress of a relatively high plastic level, but below the ultimate tensile strength of the steel. The fracture modes

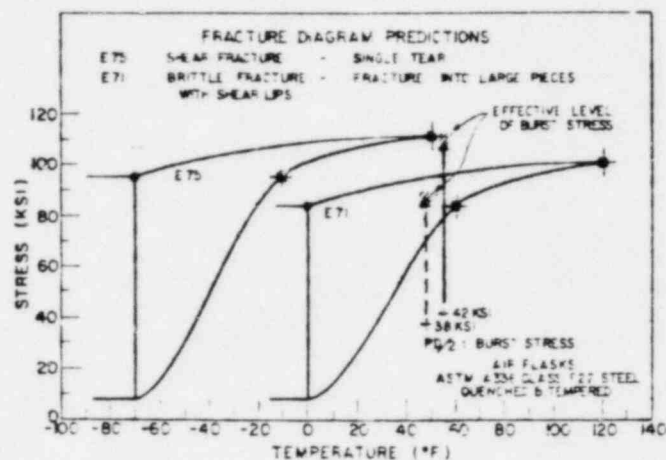


Fig. 9 Fracture analysis diagram predictions of failure modes and effective stress levels for the deliberately flawed and pneumatically burst air flasks E75 and E71

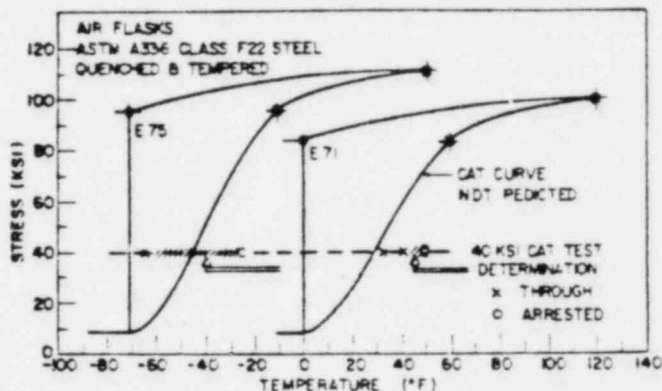


Fig. 10 Comparison of CAT test data (by RML) to predictions of CAT by the fracture analysis diagram

of both vessels were predicted exactly by the NDT location of the fracture analysis diagrams with respect to the burst test temperatures. The effective burst stresses were indicated indirectly by the degree of bulging in the flaw areas—the E75 vessel developed a considerably larger bulge than that developed by the E71 vessel.

CAT determinations for both of these air flasks were conducted [8] at the Culcheth Research Materials Laboratory in the United Kingdom. The curved sections of steels that were furnished from these flasks required flattening for the CAT tests, and a possible slight rise in transition temperature may have resulted (no flattening was required for the NDT tests). The CAT data for both air flask steels are presented in Fig. 10, in comparison to the CAT curves predictions of the respective fracture analysis diagrams. The CAT (40 ksi stress) predicted by the NDT-derived fracture analysis diagrams of the two steels were approximately -50 deg F (E75) and $+30$ deg F (E71); these are noted to be in excellent agreement with the CAT ranges of these steels determined experimentally (crosshatched regions).

The Charpy V data for the two air flask steels, presented in Fig. 11, represent a composite of NRL and RML data, which are noted to be in good agreement. Fig. 12 presents C_v energy, fracture appearance, and lateral expansion curves with notations

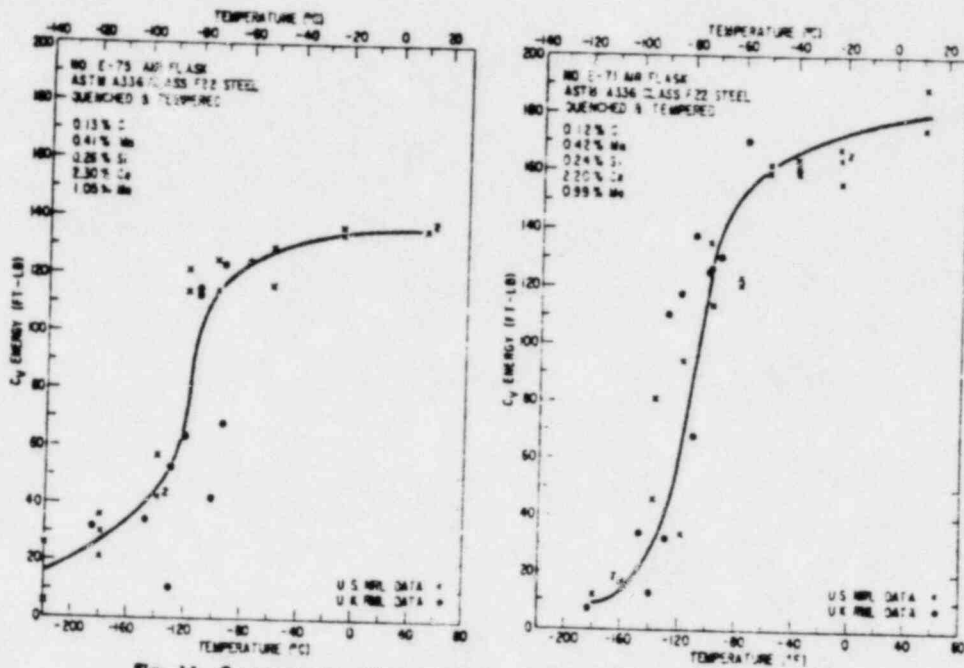


Fig. 11 C_v energy transition curves for air flasks E75 (left) and E71 (right)

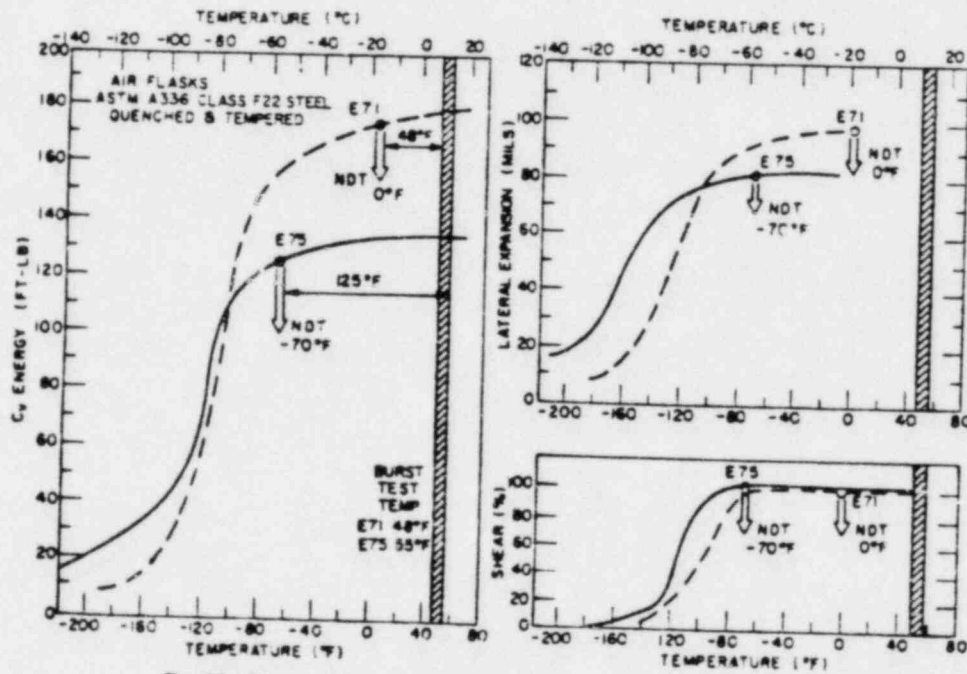


Fig. 12 C_v curves for energy absorption (left), lateral expansion (right top), and fracture appearance (right bottom) for the air flasks E75 and E71

indicating NDT and burst test temperature relationships to these curves. It is obvious that not only the C_T energy but also the other C_T criteria failed to provide a useful index for estimating the performance of these two steels. The "usual" interpretation of all three types of C_T transition curves would have resulted in predictions of similar fracture modes for the two steels. The engineering significance of this correlation failure by the C_T test will be analyzed in a later section in context of discussions of the potential use of the C_T test for indirect NDT or CAT determinations.

The foregoing discussions illustrate that the concepts of the fracture analysis diagram do not direct the design and fabrication of pressure vessels to novel or radical procedures. To the contrary, they provide means to make quantitative much of what has previously been either qualitative, intuitive, or instinctive in the design of pressure vessels. A recently published expression of similar views by F. S. G. Williams [1] is particularly noteworthy because of the breadth of practical experience that he brings to a review of the procedures, as well as his long-time association with the pressure vessel regulatory Code Bodies. He states that the fracture analysis diagram "gives demonstration to the concept that has been instinctively used, namely, that as the level of quality assurance by nondestructive tests is increased, the design stress can be increased up to the full capabilities of the material, and that conversely, as the level of quality assurance is decreased, safety can be protected by decreasing the allowable working stress used in the design. While this does not guard against careless workmanship, it does set a pattern for engineering thinking that is sound and pretty well supported by experience."

It should be noted that these observations relate to the allowable design stresses for temperatures within and below the transition range of the steel. Thoughtful consideration of the fracture analysis diagram will disclose that it does not specifically restrict the use of steels at temperatures above an arbitrarily designated temperature. This limitation is inherent to the CAT approach which inflexibly categorizes the limit of safety on a crack-arrest-temperature basis. The fracture analysis diagram permits the use of steels at temperatures far below crack-arrest and NDT temperatures provided that critical combinations of flaw size and stress level are not attained or exceeded. Thus, the designer is provided with a choice of limiting the working stress or of using steels of lower NDT temperatures. In arriving at a choice, it is essential that the fabrication and design quality and the stress level and transition temperature be considered as *interdependent* (not independent) elements of the design equation. These design tradeoffs are explicit to the intelligent use of the fracture analysis diagram.

Background of the Development of Service Performance Correlation Procedures for the Charpy V Test

Previous discussions have highlighted the amazing "sharpness" of the transition temperature range which separates "high" and "low" fracture toughness behavior of steels. It is this feature which makes it possible to consider the use of a simple notch-bend bar such as the C_T specimen to define the temperature scale location of the critical temperature range. A wide variety of simple specimen designs could be made to fit this purpose by correlation of results with service failures, or with a reference transition of direct design significance such as the NDT temperature or CAT. However, only two notch-bend bar specimens, the Charpy V and the Charpy Keyhole, have

been reduced to common usage. Of these two, the Charpy V has the desirable attribute of developing a smooth transition curve of the type required for correlation of an energy index point to a reference transition temperature. The Charpy Keyhole energy curve is discontinuous, involving high energy and low energy branches connected by a transition "scatter band"; such a curve does not lend itself to correlation usage [9].

During the World War II (WW-II) period, the catastrophic fracture of ships raised the relatively obscure problem of brittle fracture to a limelight of a national calamity. This problem provided stimulus, funds, and a backlog of service casualties which placed the C_T test in the temporary position of a primary "standard" for the definition of a significant transition temperature. Thus, the 1940's marked the emergence of the C_T test from the position of a research laboratory tool (the Keyhole test was then the engineering standard) to a position of unwarranted engineering usage for defining the transition temperatures of steels based on a ft-lb number which was considered to apply invariantly to all steels.

For the WW-II ship plate steels (which represented as-rolled, semikilled, or rimmed steels of relatively low Mn/C ratios) it was well established [10] that fracture "source," "through," and "end" plates of the fractured ships could be related to specific ranges of C_T energies at the temperatures of service failures. Briefly, the WW-II ship steels did not act as "source" (fracture initiators) at temperatures above the 10-ft-lb C_T transition. Similarly, the 20-ft-lb C_T transition temperature was found to represent the highest transition temperature of fracture "through" (propagation) behavior. A popular concept then derived that a desirable criterion of comparison of steels was the 15-ft-lb C_T temperature. Accordingly, estimates of decrease of the transition temperature resulting from alloy and heat-treatment variables were based on the C_T 15-ft-lb criterion on the erroneous assumption that it should be applicable generally to all steels.

In the early 1950's, the authors conducted extensive studies of the correlation of NDT temperatures with C_T transition curves of WW-II steels. These studies disclosed an excellent correspondence with the 5 to 10 ft-lb C_T transition [11-13]. This was an exact agreement with failure data for the ship fracture "source" plates. In other words, the drop-weight test NDT could be used to predict the fracture initiation temperatures of these steels. Extensions of these studies to fully killed steels, of higher Mn/C ratios and other heat-treatable steels disclosed that the correlation between NDT and C_T energy varied with the type of steel [3, 12, 14] and involved higher energy positions on the Charpy curve. In effect, the NDT test indicated that the "improved" steels could act as "initiators" at temperatures as high as their C_T 15, 20, and 35 ft-lb energy. Thus, the C_T test 15-ft-lb transition criterion was overestimating the decrease in transition temperature of the improved steels as compared to the WW-II steels. These early findings by the authors were first viewed with considerable suspicion but were validated [11-13] by new service failure data and CAT data developed in the middle 1950's. There is now general acceptance that all steels cannot be evaluated on an invariant C_T energy basis.

The 1960's opened with the C_T test securely entrenched as the national and international engineering standard of comparison of transition temperatures of steels. By this time, the application of NDT and CAT concepts had found considerable interest and use in engineering design. Also, a wide variety of new steels featuring distinctly novel alloying and heat-treatment practices appeared on the market [15]. This situation has led to a guessing game of the C_T ft-lb value which should be used for

estimating the NDT and CAT temperatures of these steels for the purpose of applying the new design principles to engineering practice. The dangers inherent to such free use of the C_V test is illustrated by the pressure vessel burst tests described in the previous section. The NDT temperatures for these two Q&T steels would have been predicted in serious error by a guessing estimate that a low position on the energy curve correlated with the NDT, as is the case for a wide variety of common grades of structural and pressure vessel steels. The actual position of correlation is near the top portion of the two curves, over 100 deg F above the low transition region. The alternate concepts of correlation based on percent "fibrosity" (shear fraction), and on lateral expansion are likewise shown to provide for similar error based on the popularly accepted criteria related to low levels of the respective transition curves. The only safe course, in considering characteristically new types of steel, is to establish reference correlations of NDT and C_V data. The same could be said of CAT versus C_V data.

Potentials and Limitations of the Charpy V Test for General Use in Relation to the Fracture Analysis Diagram

The C_V test may be considered as the only practical method for the indirect determination of NDT or CAT temperatures. This is presently the most realistic view of the position of the C_V test and it will remain so until other approaches are developed which provide better design guidance than presently afforded by the NDT or CAT indexed fracture analysis diagram. This latter approach required almost 15 years of development and "prove-out" time. It is unlikely that a radically new and different approach will be developed in a shorter time cycle in the future.

From this point of view, we may now analyze the potentialities of the direct NDT and CAT test methods for competition with the C_V test as tests for general usage. CAT tests are expensive, difficult to perform, and serve primarily as research laboratory tools. The drop-weight test method for the direct determination of the NDT has the potential of displacing the C_V test by virtue of being inexpensive and easily performed by any en-

gineering laboratory. The recent adoption [6] of the drop-weight test as an ASTM tentative standard (ASTM Designation E208-63T) is a major step in this direction. The advantages and disadvantages of the drop-weight test in comparison with the C_V test for purposes of general usage are as follows:

	Drop-weight test	Charpy V test
Nature of standard	Primary	Secondary
Amount of material for determination	Approximately 10 to 30X that of C_V , depending on specimen	—
Reproducibility	Uniformly high	Varies with steel
Effect of directionality	None	Marked
Cost for determination	Essentially equal	
Industry familiarity	Restricted but growing rapidly	General

The relative advantages and disadvantages are such as to place the two tests in competition for future general favor; presently the industrial familiarity with the C_V test gives it an undisputed lead.

The long range trends will reflect the degree of difficulty and complications developed in the use of the C_V correlation for determinations of the NDT temperature or the CAT. The experience to date in the correlation use of the C_V test has been favorable for common grades of structural steels. However, the preliminary experience with extension of C_V correlation use to other types of steels has disclosed serious limitations. Moreover, complications of interpretations of different C_V curve "shapes" and of specimen orientation effects provide for ready pitfalls to the unwary. The general view of the authors, based on extensive studies of the problem areas, is that the potentialities for the use of the C_V test will suffer irreparable harm if the deliberate development of proper procedures and education of

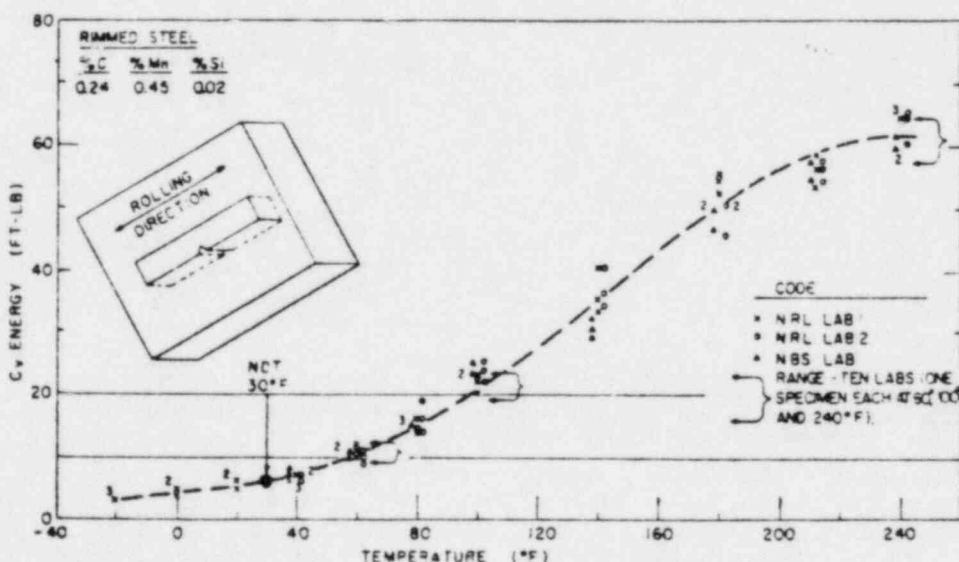


Fig. 13 Illustrating the high reproducibility of C_V data characteristics of the semikilled and rimmed World War II ship steels of low Mn/C ratios

the engineering field to these procedures is by-passed by guessing games as to the proper selection of C_T criteria for new steels. Thus, the discussions to follow will be directed to expositions of the present state of knowledge and to a call for caution in the expansion of the engineering use of the C_T test.

Semikilled and rimmed steels of low Mn/C ratios are characterized by surprisingly reproducible C_T curves with relatively little scatter of the data points at all test temperatures, as illustrated in Fig. 13. A typical correlation of the drop-weight NDT temperature with C_T energy is indicated by the large solid point. Tests of a large number of such steels provide a glamorous example of reproducibility of the C_T correlation energy (Fig. 14, left). In fact, it may be observed that the correlation range differs slightly between the rimmed and semikilled types.

The general effect of increasing the Mn/C ratio on the transition temperature, shape, and reproducibility of the C_T curve is illustrated in Fig. 15. The general trend is an increase in the upper shelf level, a decrease in transition temperature, and an increase in the C_T data scatter at any test temperature. The C_T energy value which correlates with the NDT temperature is also increased as shown in Fig. 14 center for nominally 0.18 percent C, 0.75 percent Mn compositions. The conservative "high end of correlation" C_T energy value of 10 ft-lb which applied to the steels of low Mn/C ratios now becomes a C_T value of approximately 15 ft-lb for the steels with higher Mn/C ratios. The correlation range for A302-B steels shown in Fig. 14 right indicates a similar "high end of correlation" at a C_T value of approximately 30 to 35 ft-lb.

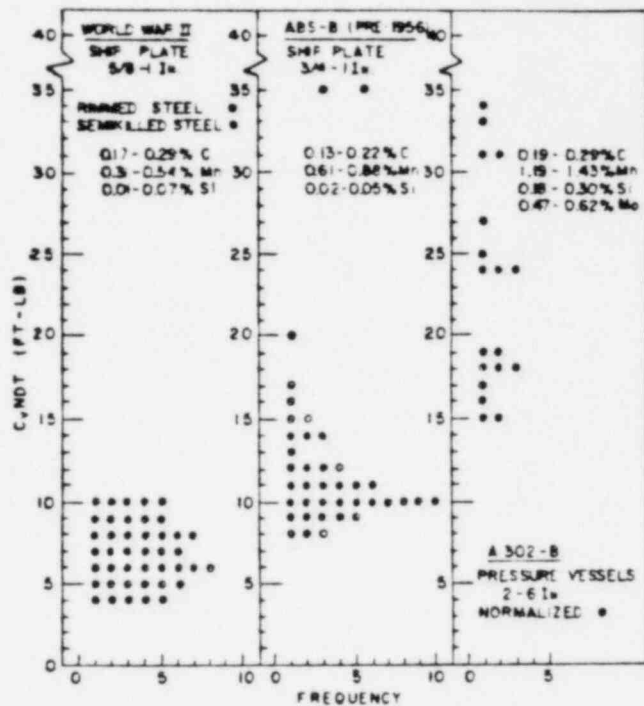


Fig. 14 Illustrating the range and variation of C_T correlations for different steel types

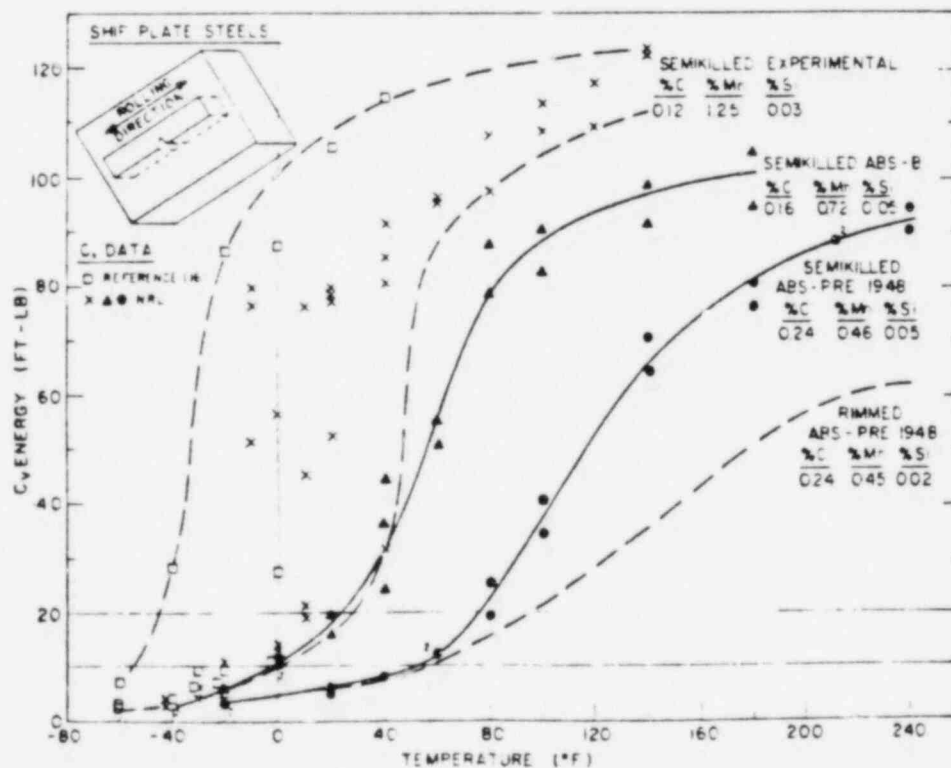


Fig. 15 Illustrating the effects of increasing Mn/C ratios on the transition temperature, shape, and reproducibility of C_T curves

Cursory reflection on these ranges would indicate that NDT determinations for the steels which correlate in the C_V 5 to 10 ft-lb range would be most accurate because the C_V correlation uncertainty is only 5 ft-lb. Conversely, the steels which show a C_V 15 to 35 ft-lb correlation range would be judged as exhibiting poor correlation. This analysis neglects consideration of the shape of the C_V curves. The very low slope of the curve in the 5 to 10 ft-lb range indicates a corresponding temperature range of 30 to 40 deg F, while the 15 to 35 ft-lb range for the steep C_V curve associated with steels of high Mn/C ratios may indicate a similar corresponding temperature range. We would conclude that, neglecting scatter effects which are usually related to steep curves (a separate problem), the correlation uncertainty for the described steels is approximately the same. The important aspect of the use of the C_V test for NDT correlation is the temperature uncertainty and not the C_V energy range uncertainty *per se*. Steels which have a well-established correlation basis and which do not exhibit excessive scatter generally provide for indirect determination of NDT within a ± 15 deg F range. For steels which show large scatter, such as the experimental low carbon, high manganese steel illustrated in Fig. 15, the accuracy of the NDT correlation depends on the number of C_V tests conducted at a given temperature. The excessive scatter developed by this steel is not uniquely related to the "low" (0.12 percent carbon content), because similar C_V scatter was reported to characterize all plate thicknesses (1/4 to 1 1/4 in.) and all heats (two 25-ton heats and seven production 300-ton heats) of the experimental, semiskilled steels ranging from 0.12 to 0.20 percent C and 1.00 to 1.35 percent Mn [16].

With the large amount of data available for the 0.12 percent C, 1.25 percent Mn steel shown in Fig. 15, we may draw a fitted curve through the data scatter indicating a C_V 30-ft-lb transition at 0 deg F. It will make relatively little difference if the NDT correlation is 20, 30, or 40 ft-lb; the steepness of the "average"

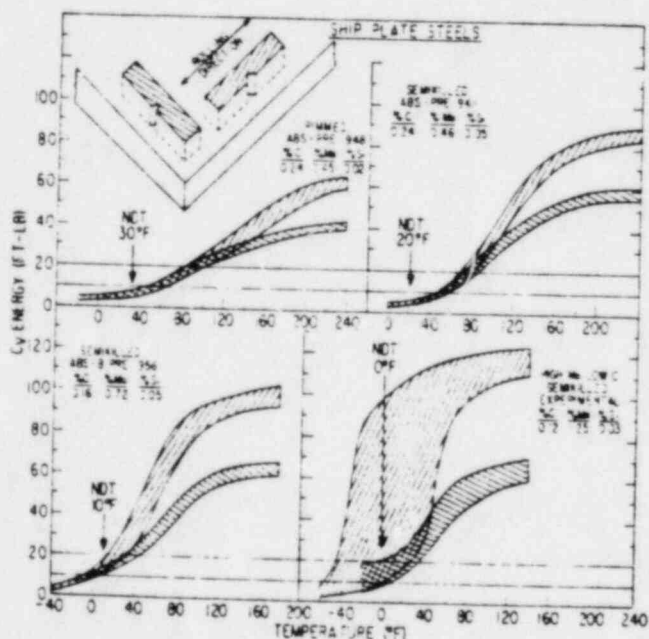


Fig. 16 General effects of C_V specimen orientation with respect to principal rolling direction for plate steels of increasing Mn/C ratios. The illustrated data bands encompass the C_V scatter observed for these steels at the respective test temperatures.

C_V curve will result in an estimate of NDT ranging from -10 to +10 deg F (drop-weight tests of this steel revealed an NDT temperature of 0 deg F). However, major uncertainties will result if existing ASTM procedures and retest provisions of A300 ("Specification for Steel Plates for Pressure Vessels for Service at Low Temperatures") are employed for C_V guarantee of the NDT for steels which show large C_V scatter. These procedures would provide for tests of three C_V specimens at a "single" specification test temperature (say 0 deg F) to determine if the steel has a minimum energy value of 30 ft-lb (assumed NDT equivalent) at the specification temperature. For the low carbon, high manganese steel shown in Fig. 15, the three C_V specimens may give average values of 30 ft-lb or 15 ft-lb, thus passing or rejecting this material on statistical happenstance. The foregoing is admittedly an extreme case which should not prejudice the applicability of C_V testing for steels having reasonable scatter. The intelligent attitude should be to recognize that the C_V scatter problem exists for compositions of very high Mn/C ratios and that safeguards of expanded testing (more specimens per temperature, and/or more test temperatures to develop a curve) should be applied for such steels. It does not follow that similar expanded testing should be applied needlessly for the steels which characteristically show small or moderate C_V data scatter. The benefits which accrue from C_V control of notch ductility of steels require additional understanding of the C_V test characteristics of the different steels.

Another area which requires additional understanding is that of the effects of directionality which the steels acquire as a result of rolling or forging [11, 17]. The general effect of C_V specimen orientation with respect to the direction of principal rolling for a plate is illustrated in Fig. 16. The principal feature illustrated by these curves is that the effect of specimen orientation is absent at the low end of the transition range and most pronounced at the upper shelf. At the low end, the C_V specimens fracture in a completely cleavage (brittle) mode. This mode of fracture is insensitive to orientation of the fracture path. Ductile (shearing) fracture is sensitive to orientation; thus, the directionality effect becomes more pronounced the larger the area of ductile fracture in the C_V specimen and most pronounced at the upper shelf, which involves completely ductile fractures.

The drop-weight specimen breaks only in the brittle mode; thus, the NDT determination is orientation insensitive similarly to the C_V specimen fractured at the low energy range of the curve. As described, the correlation of NDT to the C_V energy curve occurs at low positions for certain steels and at intermediate positions for others. This means that the steels which correlate at low points of the C_V curve are not involved with C_V specimen orientation effects. However, the steels which correlate at levels of C_V energy which relate to a partially ductile fracture will be involved with C_V specimen orientation effects. It is then necessary to specify the C_V specimen orientation for which the NDT correlation applies. For plate material, there is an established convention that the long axis of the C_V specimen is oriented in the direction of principal rolling with the notch oriented in the plate thickness direction. The problem then is simply one of correct identification of the rolling direction. For forgings, the selection of the principal forging direction requires a knowledge of the forging sequence. Complications may therefore arise, particularly for economic selection of steels which requires accurate NDT determinations.

The various discussions above point to certain areas which would benefit from expanded use of the direct NDT test method. Basically, these are the areas for which there is either uncer-

ainty or complications in the use of the C_1 specimen as a correlation tool. These are: (a) steels for which correlations have not yet been established; (b) cases involving specimen orientation uncertainty, particularly for forgings, and (c) steels which feature characteristically large C_1 data scatter. The above areas of uncertainty or complications still leave very large areas for the successful use of the C_1 test—specific problems should not be a cause of general prejudice.

Summary of NDT Data of Interest to Pressure Vessel Fabrication

The rapidly developing background of information on procedures for the prevention of brittle fractures has resulted in an increasing demand for fracture toughness data for the common grades of structural steels. In effect, the design engineer has awakened to the realization that large tonnage production steel mill procedures are sufficiently reproducible so that a particular grade of steel of a given thickness range may be categorized by an expected range of NDT or crack-arrest temperatures. However, there is no published compendium of such data to serve as guidance. In effect, the information on the use of these procedures to provide fracture-safe engineering design of steel structures has preceded the development of the necessary handbook data.

The primary specification requirements for steels involve controls on chemical composition, tensile test properties, plate thickness, deoxidation practice, and the use of normalizing heat-treatments. Detailed scrutiny of the various specifications reveals that closely similar counterparts to the old and the improved ship plate steels can be found in the common grades of pressure vessel steels. Thus information developed in the course of ship steel studies is directly applicable to pressure vessel steels. For example, except for deoxidation practice, many of the WW-II ship fracture steels would comply to the requirements applied for ASTM-A212 Grade B steels of plate thicknesses less than 2 in. The pre-1950 ABS-B ship plate steels would comply to the requirements for ASTM-A201 Grade B pressure vessel steels. The imposition of additional requirements involving normalizing heat-treatments and higher manganese contents specified by ASTM A300 results in a correspondence of A201B pressure vessel steels with the ABS-C ship plate steels.

In contrast to ship fabrication practices, thermal stress relief heat-treatments are generally required after weld fabrication of pressure vessels intended for low temperature service. Extensive studies reported previously [14] and confirmed in recent NRL tests have shown that thermal stress relief heat-treatments do not significantly affect the NDT temperatures of the structural steels. Recent practices of accelerated cooling (spray cooling) of heavy sections from the normalizing temperatures for steels intended for nuclear pressure vessel applications has promoted general interest in such treatments for nonnuclear pressure vessels and other applications. Such steels should be considered as a class distinct from the as-rolled and the normalized varieties.

It should be appreciated that steels produced by large scale commercial practices have a quality range which results from mill processing variables. If proper statistical random selection of steels is made, the NDT frequency distribution range of a particular grade of steel features a bell-shaped Gaussian distribution curve, as illustrated in Fig. 17. This curve represents the NDT frequency distribution established for the WW-II

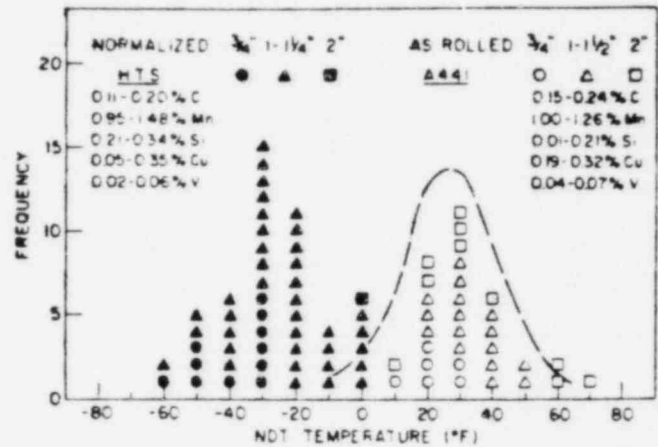


Fig. 17 Comparison of NDT frequency distribution plot of as-rolled ASTM A441 steels with that of normalized high tensile steels. The illustrated curve represents the NDT frequency distribution plot for World War II ship steels (points not shown).

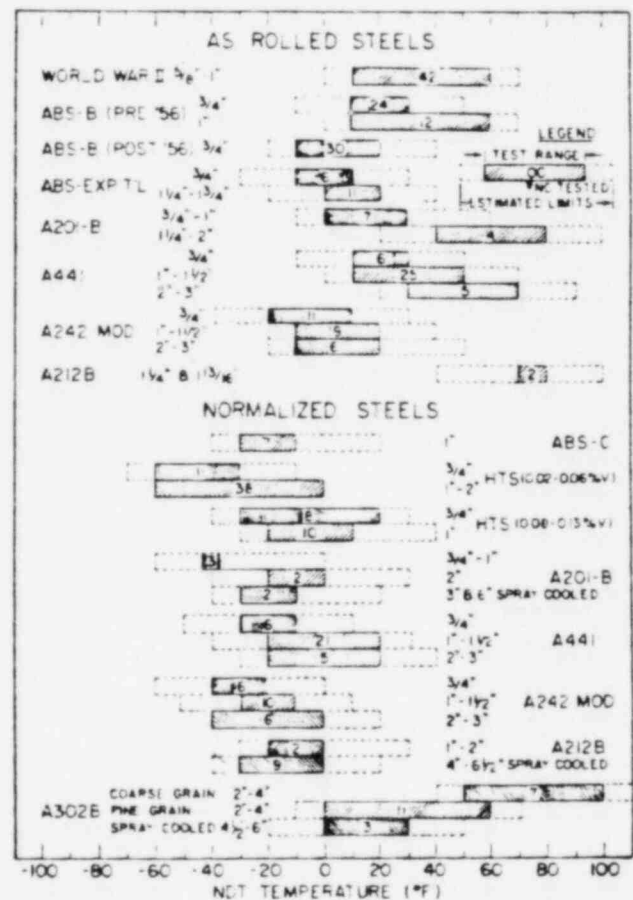


Fig. 18 Summary of NDT frequency distributions for ship plate and structural grade steels

ship fracture steels; the actual points were presented in Fig. 40 of reference [2]. The NDT frequencies of the Navy high tensile steel (HTS) and of ASTM A441 steels are given in this figure as additional illustrations of the generally expected Gaussian dis-

tributions. Both steels are essentially equivalent in respect to chemical specification requirements; the improved NDT of the HTS steel reflects the additional requirement for normalizing heat treatment.

A summary of NDT frequency distributions of the ship plate and structural steels is provided in Fig. 18. In summarizing the available NRL data, the authors have estimated the limits expected for the various thickness groups normally specified. In cases where the estimates are based on relatively few NDT test samples, C_v data available in the literature were studied for approximation of the indicated limits. In general, a moderate improvement in the NDT with decreasing thickness is observed for all steels. Normalizing heat-treatments result in a 40 to 60 deg F decrease in the NDT compared to the as-rolled condition. Accelerated cooling of heavy-section steels is noted to be beneficial, particularly for the heavy sections.

The "Low Energy Tear" Fracture Problem

The previous sections of this report were aimed at presenting a clarification of the "transition temperature," brittle fracture problem. Attention shall now be directed to a new problem area which derives from the use of high strength Q&T steels. Potential fracture in some Q&T steels and weldments is possible by a temperature-independent fracture mechanism which the authors have previously designated "low energy shear" fracture [3, 4]. This designation relates to fracture propagation not involving a cleavage mode but resulting from a very low energy absorption in tearing.

Electron microscope fractographic studies conducted at NRL by Beachem and Edwards [18-20] have shown that the microscopic appearance of "low energy tear" fracture surfaces of high strength steels involve either one or more of the following characteristic modes: (a) the opening up of a large number of voids of elliptical cross section with early rupture of the bridge volumes between the voids, termed "dimpled rupture," (b) the "decohesion" of grains by separation through intergranular paths; termed "grain-boundary rupture," and (c) transgranular fracture along planes of no presently provable orientation relationship with the crystallographic structure through which it passes; termed "quasi-cleavage rupture." This evidence suggests that "low energy tear" is the proper definition for noncleavage fractures of low energy absorption.

The chemical composition, strength level, and microstructural condition of the steel determines which one, or combination, of these modes predominate for a particular fracture. HAZ regions are particularly susceptible to the "grain boundary rupture" mode due to the development of unfavorable microstructural conditions. At high yield strength levels, Q&T steels of 0.40 percent or more carbon are also prone to the development of "grain boundary ruptures," and those with less than approximately 0.35 percent carbon are susceptible to the development of "quasi-cleavage ruptures."

The C_v test recognizes energy absorption differences in ductile tearing by the level of the upper shelf. Fig. 19 illustrates the improvements in transition temperature and in ductile tearing energy that may be obtained by shifting from the normalized (HTS) steels of approximately 50 ksi yield strength to a high alloy, Q&T steel of 80 to 90 ksi yield strength level, designated as HY-80. It is not implied that all Q&T steels of this strength level will have such excellent combinations of high shear energy and low transition temperature but that an optimized composition

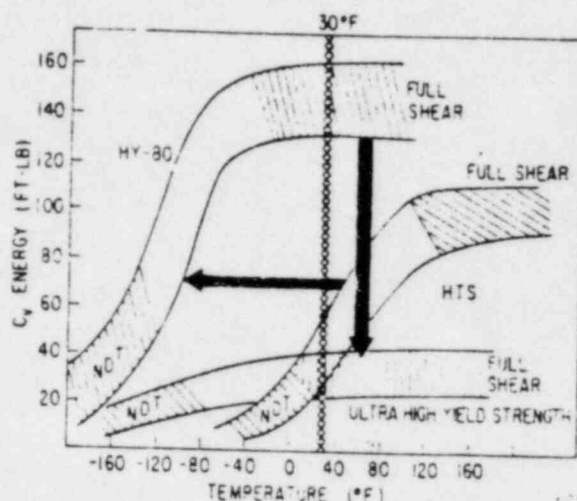


Fig. 19 Representative C_v test energy transition curves for HTS and HY-80 steels compared to those of ultrahigh strength (180-200 ksi yield) steels

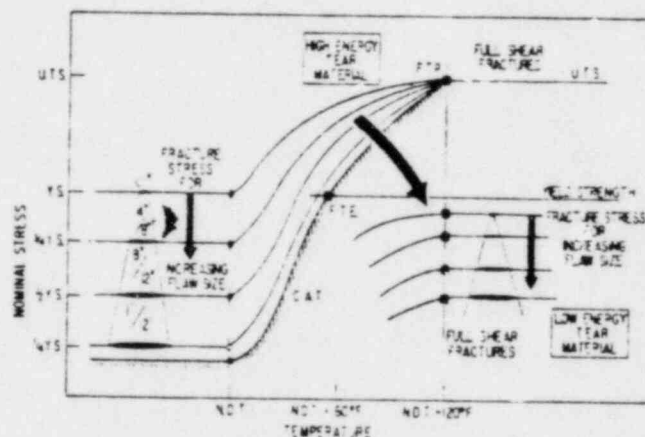


Fig. 20 Features of fracture analysis diagram modifications for steels of "low energy tear" characteristics

can be made to have such properties. The general effect of increasing the yield strength of Q&T steels to levels of 200 ksi and higher is illustrated by the lower curve of Fig. 19. At such strength levels, the very best steel compositions feature C_v shelf energies in the order of 15 to 20 ft-lb.

The effect of very low levels of energy absorption in the tearing mode is to provide for the initiation of fractures from small flaws at elastic stress levels close to yielding. Large flaws may then be expected to result in fracture initiation at lower levels of elastic stress. The rocket case fracture illustrated in Fig. 5 initiated from a flaw of approximately 1/8-in. dimensions when the hydrostatic proof stress approached the yield strength level of the steel. The fracture surfaces did not involve the presence of cleavage. A fracture analysis diagram interpretation of this fracture is illustrated in Fig. 20. This diagram represents the rocket case steel as fracturing due to the presence of small flaws when loaded to near yield levels of stress at temperatures above the FTP (full shear) transition. The effect of increasing flaw sizes is illustrated schematically by a family of fracture initiation curves. The modification of the fracture analysis diagram for

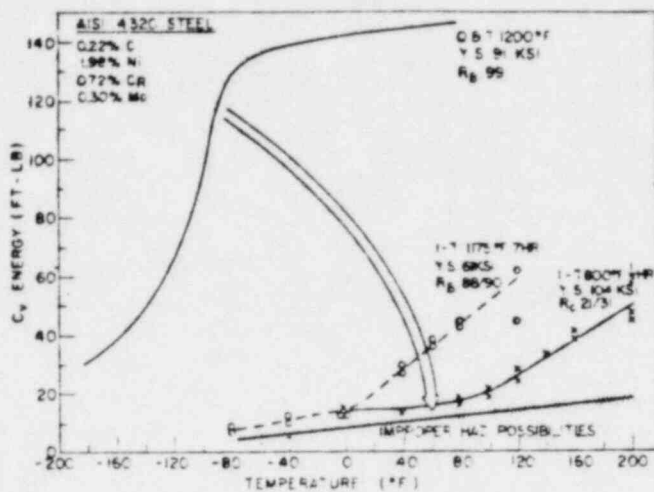


Fig. 21 Illustrating the progressive embrittlement and decrease in tear energy that results with increasing strength level of a Q&T steel which has had optimum heat-treatment

the case of low energy tear involves a general depression of the fracture initiation curves to lower stress levels, as indicated schematically by the large curved arrow. The problem of categorizing the level of tear energy which corresponds to particular combinations of flaw size and elastic stress for fracture initiation in the low energy tear mode is a matter of active current research by the authors.

We shall now consider various basic metallurgical factors involved in the development of low energy tearing for plate material and for weld heat-affected-zones (HAZ). The effect on C_T properties of increasing strength level in a given Q&T steel by tempering at progressively lower temperatures is shown by data for an AISI 4320 steel, Fig 21. The data represent values obtained for optimum Q&T heat-treatments with small specimen blanks aimed at the development of proper, fully tempered martensitic (optimum) microstructures. A progressive embrittlement is indicated by the shift in the curves and the higher NDT temperatures (large dots on each C_T curve). The decrease in tearing energy is emphasized by the large arrow indicating the drop in C_T shelf energy, with increasing strength level (from 91 to 166 ksi yield).

The drop in the shelf level of the C_T curve, and the increase in transition temperature resulting from increased strength for the Q&T steels featuring optimum microstructures presents one aspect of the low energy tear problem. Another aspect is the development of microstructural deficiencies resulting from inadequate hardenability of the steel, or improper heat-treatment of the plate. Similar conditions of unfavorable microstructural conditions of the HAZ may result from the use of improper welding procedures. Depending on the welding heat input and travel speed, HAZ areas may be quenched from too low, or too high temperatures or quenched with too slow or too fast cooling rates, resulting in hard zones or in mixed microstructures involving upper bainite, free ferrite, pearlite, or possibly retained austenite. Isothermal transformation (I-T) heat-treatments of the AISI 4320 steel, Fig 22, illustrate the possible embrittlement and drastic decrease in tearing energies at ambient temperatures that may be developed in the HAZ areas of Q&T steels. Inferior microstructures may reduce the C_T properties to the indicated low levels even for fully tempered HAZ of low strength level. In other words, stress relief may ameliorate high hardness HAZ

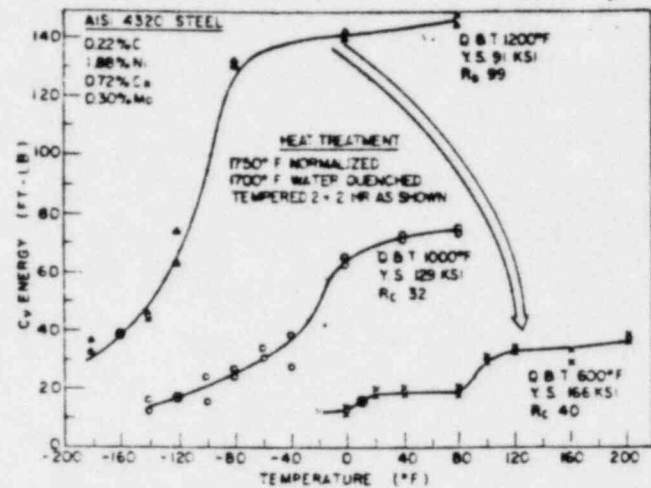


Fig. 22 Illustrating the very low tear energy characteristics that may be developed in HAZ of improperly welded Q&T steels. The possible deficiencies in HAZ microstructures was simulated by isothermal transformations (I-T) of the AISI 4320 steel.

conditions but serves little purpose for improving a HAZ which features inferior microstructures. It is not implied that all Q&T steels are prone to the development of such low HAZ fracture toughness properties. The use of proper welding procedures could result in HAZ of high fracture toughness.

The example of the rocket case rupture illustrates a low level of fracture toughness which is inherent to the best Q&T steels of very high strength (above 180 ksi yield strength) and cannot be overcome by alloying or heat-treatment improvements. Thus, the successful use of such ultrahigh strength steels must be predicated on fabrication, inspection, and design quality of the highest order. This translates to aerospace industry practices involving "thousand dollars per pound" cost for the fabricated product. In contrast, we shall now consider "dollars per pound" fabrication of conventional pressure vessels with the intent of demonstrating an intimate relationship between the gradations of quality within this cost regime and the fracture safety of Q&T steel pressure vessels. Competition, which results in depressing "quality" to below minimum tolerable levels, may have a serious effect on the reliability of such structures. The Q&T steels which provide welded fabrication capabilities in the "dollars per pound" regime are presently restricted to the range of 80 to 120 ksi yield strengths. Accordingly, the discussions that follow relate strictly to Q&T steels of this yield strength range (designated 80/120 Y.S.).

Fracture-safe analyses of the potential fracture mode of pressure vessels constructed of 80/120 Y.S. Q&T steels require separate considerations of the transition temperature and the low energy tear problems. The transition temperature problem may be analyzed in terms of the fracture analysis diagram provided the NDT temperature is known. Once the NDT has been determined, it is necessary to analyze for possibilities of low energy tear problems in either the plate or HAZ at temperatures above the transition range. In other words, the temperature range above the CAT transition curve cannot be considered invariantly safe, as it can for the case of low strength, mild steels of high tear energy properties. The possibilities of low energy tear fracture problems for these steels are highly dependent on the composition and heat treatment. Some steels may feature very high tearing fracture energy (no problem) while others may be of sufficiently



Fig. 23 Illustrating a low energy tear fracture in the HAZ of a Q&T steel weldment

low value to require flaw size analysis. The very low C_V shelf levels which characterize some of these steels indicate relatively poor plate tear energies, particularly in the "weak" direction of poorly cross-rolled or straight-away-rolled materials. Steels with such features could develop low energy tear fractures of the plates at ambient temperatures, similar to that of the rocket case example (Fig. 5) but for much larger flaw sizes.

Even if it is determined that low energy tear fracture is not a possibility in a given 50/120 Y.S. Q&T steel, there remains a question as to the characteristics of the HAZ. In fact, for these steels, the HAZ properties present the most crucial questions because the HAZ is the element of greatest potential variability. All Q&T steels are susceptible to HAZ degradation resulting from improper welding procedures, but some of these steels feature a comparatively high propensity for HAZ degradation even with the best of welding practices. Severe HAZ degradation may result in fracture initiation at ambient temperatures from relatively small flaws that are subjected to high elastic stress levels (located at points of unfavorable geometry). Fracture propagation may then proceed along the HAZ path into regions of relatively low elastic stress levels. In essence, such weldments fracture by HAZ separation, leaving the plate sections essentially fracture-free.

At present, there are no generally practical, laboratory means of conducting HAZ fracture toughness evaluations. The authors' approach to this problem for the past twelve years has been to conduct explosion bulge tests. Fig. 23 illustrates the behavior of a Q&T steel which featured low energy tear fracture properties in the HAZ. A Navship Standard procedure [21] derived from this approach relies not only on performance exhibited by as-welded samples but also on weldments to which a crack-starter bead is added for the introduction into the HAZ of a brittle-weld-crack flaw (approximately $1/2$ in.) to provide a realistic flaw condition. The fracture toughness of the HAZ is then judged on the extent of HAZ tearing which results from the application of low or high levels of bulge deformation. Qualification of a steel and its associated welding procedure by this method assures that no HAZ paths of low fracture energy will be present in ship structures. It is important to note that this is the last step in qualification—the steel and weld metal are previously investigated to ensure that high fracture toughness is inherent to these two potential fracture paths. Quality control is exercised by C_V specifications of plate and weld. Following the HAZ qualification, quality control is exercised by requirements for strict adherence to the welding procedures which qualified the HAZ fracture path.

The foregoing generalizations require some "case in point" examples to illustrate the realities of the problem. In this respect a comparison of high alloy (HY-80) and lean alloy (commercial 80/120 Y.S.) Q&T steels serves to highlight some aspects

of the influence of the cost element on the attainable levels of fracture safety assurance. The cost of the high alloy steel plate is approximately twice that of the best of the lean alloy varieties—competition within the lean alloy range influences the spread in cost. The cost of fabricating both types of steels is primarily related to the fabrication quality aim, and is relatively independent of the basic cost of the steel. The term "cost" is used herein to signify economic factors which affect the choice of the steel. From a metallurgical point of view the element of steel cost is related to alloy content and alloy content is related to heat treatment and welding response. As the alloy element content is decreased, it becomes more difficult to develop "fracture tough" microstructures, both in the heat treatment of the plate and in the welding transformations of the HAZ. Thus, a cost-element pressure for the use of alloy lean steels creates an increased requirement for exacting control of welding conditions because of the increased possibilities for degradation of HAZ properties to low levels of fracture toughness.

The fracture toughness level of the HAZ is related to the inspection quality that is required—the lower the fracture toughness the more exacting is the requirement for inspection. Similarly, the lower the fracture toughness, the lower the tolerance for design imperfections, fit-up deviations, and other factors which affect the level of stress applied to HAZ regions located at points of geometric discontinuity. In other words, the alloy content of the steel is a key element to the fracture-safe design analysis.

HY-80 steels serve as an example of a high alloy steel. This steel has been restricted primarily to military and other critical applications because of its comparatively high cost. The high amounts of Ni, Cr, and Mo used for HY-80 represent a balance designed to maximize toughness and to promote optimum microstructural response on welding thick sections. The average HY-80 plate has an NDT of approximately -160 deg F. This signifies that the material develops full shear fracture (FTP) to temperatures as low as -40 deg F. The CAT for stresses of yield stress level is approximately -100 deg F. Weld metal which approximately matches these characteristics is available for this (80 to 90 ksi) yield strength level. The use of proper welding procedures with the HY-80 steel results in HAZ of very high fracture toughness, as illustrated by the explosion bulge test sample of Fig. 24. Weldments of properly welded HY-80 may be used to very low temperatures without complications of either brittle fracture paths or low energy tear fracture paths. The same HY-80 may be welded improperly, resulting in a fracture toughness

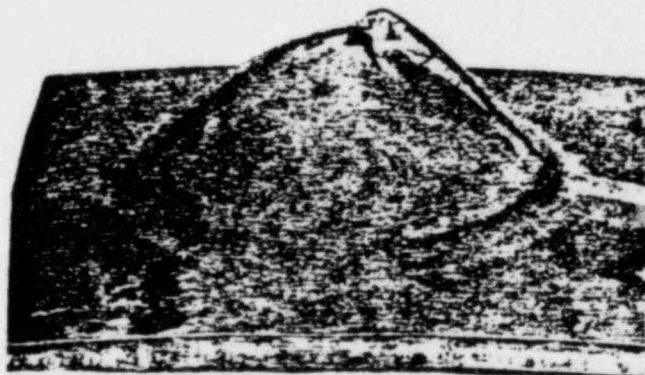


Fig. 24 Illustrating the high fracture toughness characteristics of properly welded HY-80 steel. All ruptures in the plate, weld, and HAZ represent high energy absorption tears.

degradation of the HAZ involving brittleness at low temperatures and/or low energy tear fracture at ambient temperatures. Thus, the reliability of such HAZ degraded weldments hinges entirely on the specific metallurgical condition of the HAZ, the flaw size that may be present in such zones, and the stress level normal to the line of the HAZ.

A wide variety of 80/120 Y.S. Q&T steels less expensive than HY-80 have been developed for welded fabrication. These steels have relatively low Ni, Cr, and Mo contents. The specific compositions depend on the thickness range of the plate material and on the desired balance between cost and other factors. Some of these steels are balanced in the direction of minimizing cost and others on the direction of maximizing toughness and weldability—within the limits of the balance that provides for economic competition. The only generalization that may be made regarding these steels is that they cannot be lumped into a single category as the yield strength range may suggest. The plate fracture toughness, the HAZ degradation to be expected within controlled welding limits, the effects of exceeding welding limits, etc., are highly specific to the metallurgy of these steels. Unfortunately, some of these steels are being used without sufficient laboratory information on fracture toughness or on required welding limits. Another general problem that these steels pose for a fracture-safe design is that the information on the necessary welding controls, inspection quality, and design quality requires intelligent application beyond that which is usually exercised in average shops.

We shall now consider the characteristics of the more highly alloyed grades of the commercially available steels having guaranteed yield strength properties in the 80 to 100 ksi yield strength range. It is emphasized that these discussions do not apply to the "cheaper" (cheaper) steels of this strength range. It is emphasized also that the discussions apply to plates of the thickness range covered by notch ductility guarantees. The expected NDT of these steels is in the range of -50 to -100 deg F, depending on thickness. The approximate CAT is in the range of -40 to 10 deg F, and the full shear temperature is in the range of 0 to 50 deg F. In effect, these steels should have good fracture toughness to low service temperatures if the fracture analysis diagram is applicable. Service experience indicates that this is the case insofar as plate material of free quality is concerned. However, a few failures have been reported in service and in hydrotests involving HAZ fracture paths. Failure analysis in such cases is almost an impossibility, because it involves assessing the fracture toughness of the HAZ and there are no methods for doing this. The general conclusions that are usually drawn in such failure analysis is that the fabrication resulted in the development of adverse conditions such as mismatch, large bead relief, and cracks or that fatigue action resulted in cracks. These findings implicitly emphasize the fact that the sensitivity of the degraded HAZ path to factors which involve flaws and high stresses is the general problem. The use of these steels at high stress levels (as presently limited by Codes) or to higher stress levels as is advocated demands close adherence to proper welding practices, good inspection, and quality design of details. If these are provided the weldments perform in a tough manner; if these are ignored the fracture safety of the vessel may be compromised.

The wide range of sensitivity to HAZ deterioration to which Q&T steels are potentially subject makes it imperative that HAZ qualification test procedures and stringent quality control practices be developed and adhered to. The welded fabrication of Q&T pressure vessels requires quality shop practices backed by a sufficient amount of well digested technical information. In effect, shop control must involve technical supervision. The

problems of fracture-safe welded fabrication of high strength Q&T steels have certain similarities to the problems encountered with the new precipitation hardened, carbon-free martensite steels (maraging type). These steels when heat-treated to very high yield strength levels are also subject to low energy tear plate fractures initiated from small flaws loaded to high elastic stress levels. Weld and HAZ characteristics of these steels are presently relatively undefined and no significant comment can be offered.

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