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1980 Adams Lecture: Twenty Years of Pressure Vessel Steel Research

The steels and fabrication processes used today are more sophisticated than 20 years ago, and more complex alloy compositions, more high technology welding processes and more involved pre- and postweld heat treatments are predicted for the future

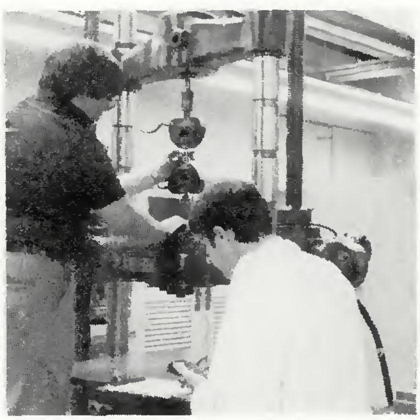
BY ALAN W. PENSE

ABSTRACT. The 20 year period between 1960 and 1980 has seen some significant advances in the pressure vessel industry and in pressure vessel steels. The 1960 Adams Lecture served as a review of the status of the metallurgy, properties and weldability of these steels to that date. Since that time, the Welding Research Council has carried on an active program of research on pressure vessel steels and Lehigh University has been a participant.

It is the purpose of this 1980 Adams Lecture paper to review this 20 year effort and indicate the status of pressure vessel steel research today. In the course of this review areas of research and practical development important to the future of the industry will be identified.

Introduction

The 20 year period from 1960 to 1980 has seen some significant changes in the pressure vessel industry. This period has produced new design procedures, new material requirements, new property evaluation techniques and, of course, some new pressure vessel steels. Change in any of these important areas requires a research effort by many organizations to provide the background information that permits innovation. Research is also required to provide the mechanical property data that code writing bodies need in order to be assured that advances in these areas are consistent with the production of safe and reliable structures.



Tension testing of low alloy steel specimens

For this reason, it is not surprising that the Pressure Vessel Research Committee and the Weldability Committee of the Welding Research Council have been active sponsors of programs over the last 20 years that were aimed at precisely these goals. Lehigh University has been a partner in Welding Research Council research since the inception of that organization. Because of this, it has been the author's privilege and opportunity to share in this interesting and rewarding endeavor over the 1960 to 1980 period. In the 1960 Adams Lecture, Dean R. D. Stout of Lehigh University took upon himself the responsibility of summing up the significant research in the area of pressure vessels steels performed at Lehigh University until that time. I will attempt to fill the same role . . . but in 1980.

To cover the changes in pressure vessel technology over 20 years, even as reflected in one University, it is necessary to focus on the broad issues rather than the details—that is, to see

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the forest rather than the trees. For this reason, the main thrust of this Adams Lecture will be on the trends and directions rather than the specific programs. Finally, it will be my opportunity as I make this review, to play the role of prophet and project directions for the future. In the end, it is intended that this assessment may serve to focus attention on the areas that can be the subject of productive research and practical development in the future.

Composition, Heat Treatment and Microstructure Development

A listing of some of the common pressure vessel steels that have been examined in the investigations over the past 20 years is seen in Table 1; their properties appear in Table 2. The chemical compositions listed are for typical heats on which we engaged in research and are by no means exhaustive of all the heats or even all of the materials that have been studied over that time period. They are arranged according to composition and application.

It should be pointed out that the mechanical properties listed on Table 2 are more or less a composite of the investigations performed over the 20 year period. Therefore, they do not represent properties of specific heats but are typical properties and must be treated as generalizations. Moreover, the purpose of the studies was trends and processes rather than specific statistical data; thus the results must be viewed from this vantage point.

Table 1—Compositions of Typical Pressure Vessel Steels, %

Steel	C	Mn	P	S	Si	Ni	Cr	Mo	Other
C-Mn: low and medium strength									
A201	.13	.49	.010	.020	.09				
A212	.27	.72	.007	.025	.24				
A515-70	.28	.79	.008	.026	.22				
A516-70	.27	1.09	.011	.025	.27				
A537	.18	1.21	.013	.016	.30				
Cr-Mo: elevated temperature									
A387B	.17	.59	.012	.024	.21		.91	.51	
A387D	.11	.37	.010	.010	.26		2.20	.96	
A542	.12	.47	.007	.020	.33		2.11	1.11	
Ni low temperature									
A203A	.14	.51	.014	.022	.18	2.35			
A203D	.11	.45	.010	.023	.23	3.35			
A645	.07	.48	.010	.010	.23	4.99		.31	
A553I	.07	.67	.009	.018	.26	8.96			
Combination high strength, light section									
A517A	.15	.80	.025	.015	.70		.60	.15	Zr
A517B	.17	.86	.011	.016	.23		.46	.19	V,Ti,B
A517E	.15	.65	.010	.012	.28		1.76	.50	Ti,B
A517F	.18	.85	.008	.017	.25	.85	.48	.50	V,Ti,Cu,B
A517J	.19	.68	.010	.021	.25			.57	B,Ti,N
Combination high strength, heavy section									
A302B	.19	1.28	.013	.010	.22			.45	
A533B	.20	1.28	.019	.030	.21	.53		.52	
A508-2	.20	.69	.011	.013	.25	.69	.37	.51	
A508-4	.18	.31	.007	.012	.18	3.53	1.73	.50	
A543	.16	.34	.014	.020	.27	3.60	1.89	.53	
Microalloyed									
A737B	.14	1.44	.009	.006	.19				Cb
A737C	.20	1.29	.010	.007	.30				V,N

The first group listed—carbon-manganese steels—has been investigated in a series of studies over this 20 year period as their mechanical properties, particularly their strength and toughness, have evolved. This group of steels is still one of the most widely used for all kinds of pressure vessel service, and will continue to be for many years, because of their low cost.

The first few materials—A201, A212 and A515-70—represent steels in which fine grain practice is not employed and normalizing heat treatment is done only occasionally (specifically, for heavy sections). The A516-70 and A537 represent materials in which the grain size is controlled and normalizing heat

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The Comfort A. Adams Lecture was created in 1927 to commemorate the contributions to welding by Harvard University's Professor Comfort A. Adams, founder and first president of the American Welding Society. Upon being invited to present the 1980 Adams Lecture at the 61st AWS Annual Meeting in Los Angeles, California, Dr. Alan W. Pense—a member of the Lehigh University Faculty since 1960—became only the 38th person to be so honored by the American Welding Society.

At Lehigh, Dr. Pense is Professor of Metallurgy and Materials Engineering as well as Chairman of the Department of Metallurgy and Materials Engineering, a post that he has held since 1977. A graduate of Cornell University in 1957 with a bachelor of metallurgical engineering degree, Dr. Pense received the master of science and doctor of philosophy degrees from Lehigh in 1959 and 1962, respectively. The reader will observe Dr. Pense's references to Dr. R. D. Stout in his 1980 Adams Lecture. It is to be noted that Dr. Pense studied under Dr. Stout at Lehigh, that Dr. Stout is a Past President of AWS and was the

A. W. Pense

Society's 1960 Adams Lecturer.

A specialist in physical and mechanical metallurgy, Dr. Pense has conducted sponsored research on the properties and welding of steels for the American Iron and Steel Institute, the National Science Foundation, the American Petroleum Institute, the Welding Research Council, the National Cooperative Highway Research Board, and the Department of Transportation.

Dr. Pense is co-author of about 45 articles published in technical and professional journals and the co-author of one book. He is a member of the American Welding Society, the International Institute of Welding, the American Society for Metals, the American Society for Testing, and Materials, the American Society for Engineering Education, the American Scientific Affiliation, Sigma Xi, Omicron Delta Kappa, and Tau Beta Pi. He has received awards for teaching, research and service from Lehigh University, the American Society for Metals, and the American Society for Engineering Education. He received the AWS Spraragen Award in 1963, the Adams Memorial Membership in 1966 and the Jennings Award in 1970. Dr. Pense has served on the AWS Technical Papers Committee and the Welding Handbook Committee where he recently completed a term as chairman.

Table 2—Typical Mechanical Properties of the Steels

Steel	Treatment	Yield strength ksi,	Tensile strength, ksi	Elongation, %	15 ft-lb transition temperature, °F
C-Mn: Low and Medium Strength					
A201	N	35	58	40	+25
A212B	Q&T	48	78	40	−20
A515-70	N	44	75	33	+25
A516-70	N	50	82	32	−50
A537-1	N	52	75	35	−75
A537-2	Q&T	65	85	25	−80
Cr-Mo: Elevated Temperature					
A387D	N	40	70	28	−25
A542	Q&T	95	110	19	−50
Ni: Low Temperature					
A203D	N	53	73	38	−160
A203D	Q&T	62	80	33	−210
A645	Q&T&T	70	100	32	< −320
A553-I	Q&T	100	110	23	< −320
Combination High Strength, Light Section					
A517A	Q&T	113	121	20	−50
A517F	Q&T	119	131	19	−70
Combination High Strength, Heavy Section					
A533-1	Q&T	70	90	27	−45
A508-2	Q&T	65	85	26	−50
A543	Q&T	95	110	23	−125
Microalloyed					
A737B	N	56	80	29	−120
A737B	Q&T	62	83	27	−120
A737C	N	60	85	26	−80

treatment more frequently applied. The A537 is also frequently marketed in the quenched and tempered condition, as is the A516-70 in some heavy sections. The primary effect of such treatments is an increase in the fineness of the microstructure which gives improved strength and substantially improved transition temperature.

Figure 1 shows the microstructures characteristic of A201 and A516 in the normalized condition. A516 is not substantially different from A212B, A201 or the A515 except for somewhat finer grain size. The microstructure is pearlite-ferrite aggregates with substantial amounts of coarse ferrite. Table 2 shows that the strength of the carbon-manganese materials generally increases with increasing manganese content. It should especially be noted that A537 Cl 2, which is quenched and tempered, is higher in strength than its normalized counterpart by about 10 ksi (68.9 MPa) for both yield and tensile strength. In terms of impact toughness, the transition temperature of the A537 Cl 2 is slightly but not substantially better than the normalized grade. The effectiveness of the use of fine grain practice and quenching and tempering is obvious.

The properties on Table 2 represent section thicknesses in the order of 1 in. (25.4 mm), although greater thicknesses are permitted for most of these grades. For heavier sections, the improvement is even more critical as toughness normally declines with section thickness.[1] The trend over the last 20 years is away from the semikilled to the killed steels. The improved toughness this provides is of such benefit that the slightly increased cost is more than offset by the reduced incidence of brittle fracture in fabrication and service.

The second group of steels listed in Table 1—the Cr-Mo steels—is designed primarily for elevated temperature service. At the beginning of the 20 year period under discussion it had already been understood that quenching and tempering would produce better mechanical properties in terms of ambient temperature strength in this type of material. However, since they are designed primarily for elevated temperature service, the advantages of such quenching and tempering were not apparent. Therefore, this group of materials was used in a series of investigations on elevated temperature behavior specifically to determine whether quenching and tempering would produce better mechanical properties in the low creep range or whether the materials in the annealed and normalized conditions would actually perform better in creep service.

The results of these investigations[2,3,4] which covered a period of some six years are discussed under creep properties. It is important to note here, however, that the application of quenching and tempering to improve strength and toughness microstructures (to produce the lower bainite rather than ferrite-pearlite aggregates) can be very effective, and this treatment has been widely adopted for the 2¼ Cr-1Mo grade. Examples of the mechanical properties that can be developed and the microstructures of one of these materials are seen in Table 2 and Fig. 2.

The development of optimum microstructures to produce superior mechanical properties is nowhere more evident than in the nickel alloyed low temperature steels. At the time of the Adams lecture in 1960, the primary low temperature steels were A203 Grades A and D; however, materials of higher nickel content were already under development at that time. From these developments have come three materials now in pressure vessel service—A645, A353 and A553I.

Most of these grades are now used in the quenched and tempered condition. The first of these was A353, a 9% nickel steel used in the normalized condition. The A553I specification is the same basic composition steel, except that it is heat treated by quenching and tempering. The expense of this material led to consideration of grades of leaner nickel content that would also perform well at

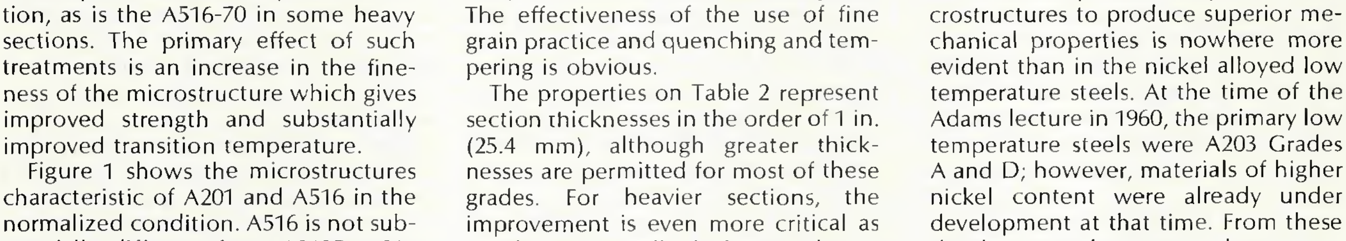

Fig. 1—Microstructures of C-Mn steels: A—A201(N); B—A516(N); C—A537-1(N); D—A537-2(Q&T). X250 (reduced 28% on reproduction)

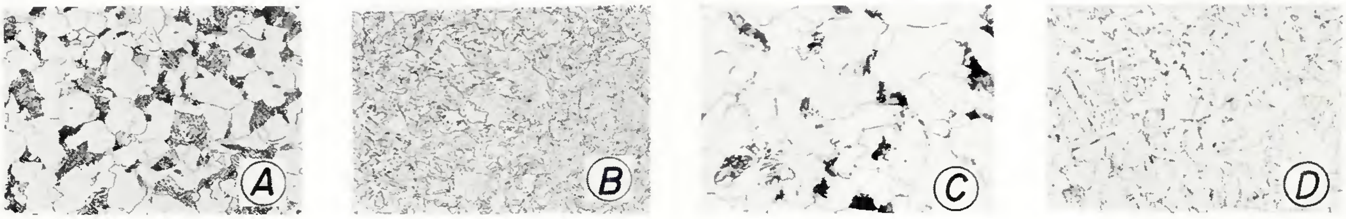

Fig. 2—Microstructures of Cr-Mo and Ni steels: A—A387B(N); B—A387B(Q&T); C—A203D(N); A203D(Q&T). A and B—X250; C and D—X500 (reduced 28% on reproduction)

low temperatures. The A645 specification was an outgrowth of these considerations. Examination of the mechanical properties in Table 2 shows that the strength and toughness of the nickel-bearing grades has increased steadily over this time period and, thus, materials with tensile strengths of 100 ksi (689.5 MPa) and transition temperatures sufficiently low to handle liquefied natural gas service have been developed. Indeed, as shown in Table 2, the A203 Grade D in the quenched and tempered condition will have a very low transition temperature.

The corresponding microstructural changes produced by this treatment are seen in Fig. 2.[5] The development of optimum mechanical properties—that is, toughness—in these materials requires control of not only ferrite, pearlite and/or tempered martensite but also retained austenite. Thus, as shown in Fig. 3, the retained austenite content in the range of 6 or 8% is critical in the development of the optimum toughness-strength characteristics.[6]

In the third and fourth groups we find a listing of combination alloy materials that are designed for high strength pressure vessel applications. The compositions listed are only a small sampling of the many now available. At the time of the 1960 Adams lecture, none of these ASTM specifications existed. Many of the steels had already been developed and were part of proprietary offerings of different steel companies. During the past 20 years they have been brought together in a few unified ASTM specifications.

Over that same time period materials for heavy sections also heat treatable to high strength levels had been developed, primarily from compositions of older materials such as A302-B, which were used prior to 1960 in normalized condition. Thus A533B is an outgrowth of the basic A302-B composition. Similarly, other heat treatable forging and plate grades represented by the A508 and A543 specifications have been developed. Examination of early studies[7] of the mechanical properties that can be developed from these grades of steel is seen in Figs. 4 and 5 which show good toughness-strength combinations in thicknesses up to and including as much as 12 in. (0.3 m). A comparison of the A302B and A533B microstructures is seen in Fig. 6.

The application of these quenched and tempered steels increased during 1960–1980. However, the weldability of these materials becomes increasingly difficult as the alloy content increases. Thus, their application has to some extent been limited by their weldability. Nonetheless, the application of quenched and tempered heavy section steels for nuclear pressure vessels, for hydrocracker vessels and for many other applications, as predicted by the 1960 Adams Lecture, has occurred and these have displaced many other grades of similar or leaner composition in normalized condition.

The last group of materials listed on Tables 1 and 2 are the microalloyed steels. These are the most recent development in steels for pressure vessel service, and they are unusual in that they appear, in a way, to be a retreat to the C-Mn steels at the top of Table 1. There is a sense in which this is true. Their mechanical properties are not at the strength levels that are achieved by the combination high strength groups. However, they have good toughness and offer a cheaper and more weldable alternative to the high strength materials. As such they are attractive to pressure vessel manufacturers when section sizes are not exceptionally large and good weldability combined with modest cost are an important consideration.

In the opinion of the author, the development of quenched and tempered steels, especially the combination high strength steels, has perhaps reached a peak during the 20 year period since the first Adams lecture on this topic. Development of new materials today has moved toward the microalloyed group. Comparative microstructures of the quenched and tempered and microalloyed groups are also shown in Fig. 6. It is apparent that the combination high strength groups represent mixtures of martensite and lower bainite even in heavy sections. The A737 microstructure seems to have much more in common with the A537 than it does the A533B.

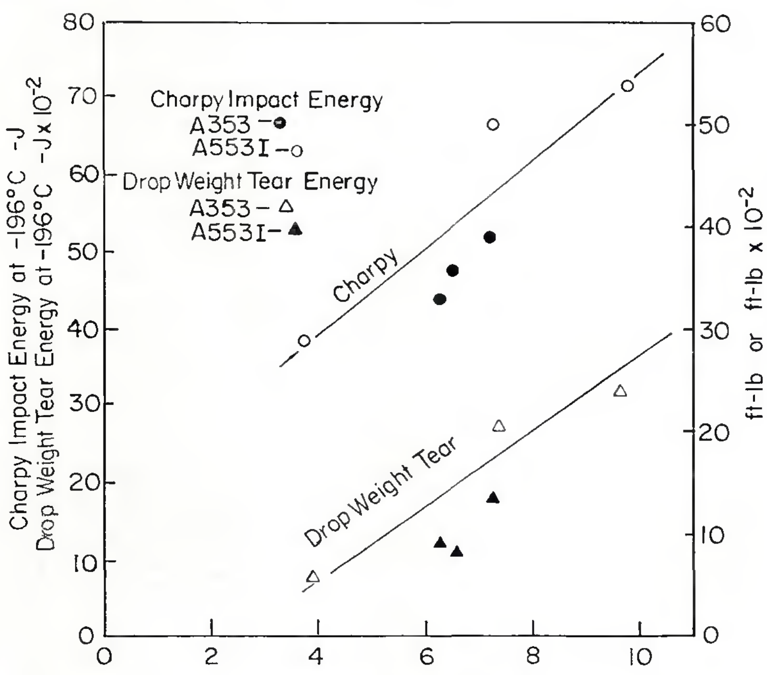

Fig. 3—Effect of retained austenite on toughness of Ni steels (courtesy of Armco Steel Corp.)

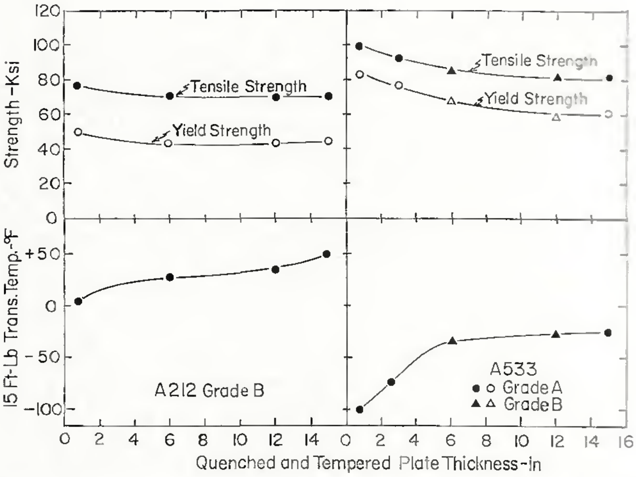

Fig. 4—Effect of heat treatment on heavy section A212 and A533 steels

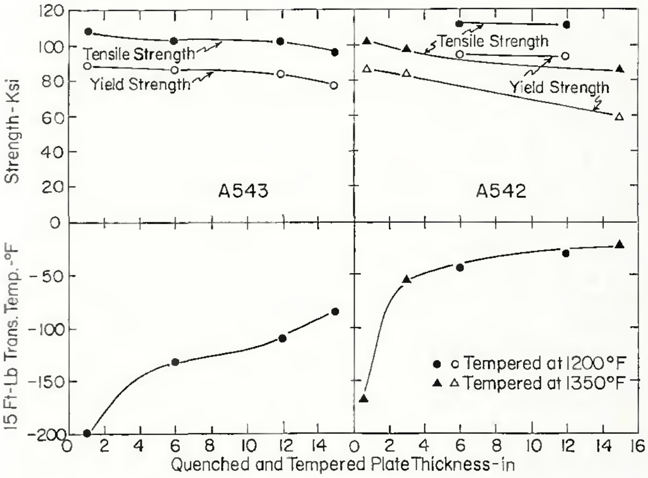

Fig. 5—Effect of heat treatment on heavy section A542 and A543 steels

The emphasis here on the microalloyed steels does not imply there will not continue to be applications for all of the steels listed on Tables 1 and 2. The intent is merely to point out there are many respects in which the expense and weldability of the quenched and tempered steels limit their application.

The inclusion of the microalloyed steels in ASTM designations is just starting. Ultimately they will probably fill an important role in pressure vessel service. The ideal microstructures for these steels (i.e., fine grained nearly pearlite-free ferrite with controlled precipitates of carbides and carbonitrides) can be achieved by several processing treatments. Normalizing, and quenching and tempering, are both utilized at the present time. Ultimately, controlled rolling will undoubtedly take its place as a viable and cost-effective means to achieve the same results in moderate section sizes.

Mechanical Property Development and Assessment

Although mechanical properties have improved over this 10 year period, as illustrated in Table 2, particularly the balance of strength and toughness obtainable, perhaps the most dramatic advance in this area is the rapidly changing way in which properties are assessed. The first 10 years of the era saw an increasing use of the Charpy impact and drop weight tests for toughness assessment. The use of these emphasized the reality of the toughness transition characteristic of most low carbon steels, and the transition temperature, above which ductile behavior was expected, was the controlling toughness parameter. This philosophy could be and was applied to steels for both ambient and low temperature service, including the cryogenic steels, where the traditional transition temperature could be below −325°F (−198°C).

During the last 10 years, emphasis has shifted to quantitative assessment of toughness behavior above and below the transition temperature, including the ductile regime where pressure vessel service may occur. This assessment is obtained using principles of fracture mechanics, an analysis procedure used only for essentially brittle materials 20 years ago. These concepts have been developed for use with pressure vessel steels to the point where some sections of the ASME code are written from this viewpoint. Thus the K_{IC} (or J_{IC}) of a steel is becoming an essential property characterization for both nuclear and nonnuclear service. Although many difficulties in this analysis technique still exist, it is here to stay and will probably grow in importance.

The use of fracture toughness concepts started slowly in the pressure vessel field. This was primarily because most of the materials used in pressure vessels have relatively high toughnesses and, therefore, classical fracture mechanics concepts were not strictly applicable. Perhaps the turning point came with the development of fracture toughness information and recommendations by the Pressure Vessel Research Committee for use in nuclear reactor vessel materials.

This document, published as Welding Research Council Bulletin No. 175,[8] was the product of a great deal of work by members of a Pressure Vessel Research Committee Task Group and represented the state-of-the-art in application of fracture toughness to pressure vessels at that time period. This recommendation has since become incorporated in sections of the ASME Boiler and Pressure Vessel Code and has influenced the application of fracture mechanics to all pressure vessel materials. The results of investigation (a combined data curve on these materials) is seen in Fig. 7. This is called the K_{IR} curve. It is a lower bound curve of fracture toughness data for 50 ksi (344.7 MPa) strength nuclear plate and forging materials over a broad range of temperatures. The NDT temperature is used to index a material to this curve.

The use of fracture mechanics provides assurance of adequate vessel

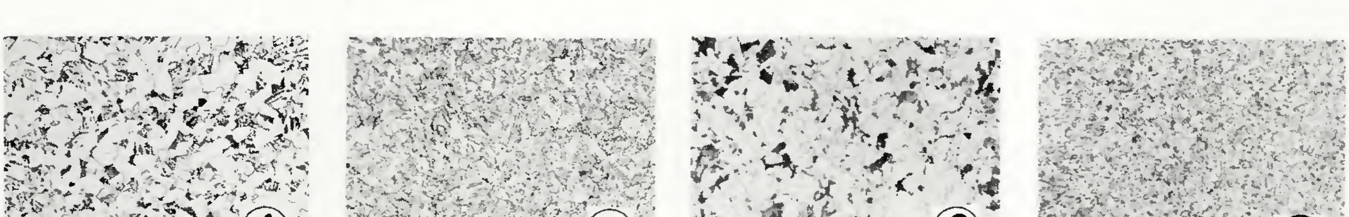

Fig. 6—Microstructures of combination alloy and microalloy steels: A—A302B(N); B—A533B(Q&T); C—A737C(N); D—A737C(Q&T). X250 (reduced 28% on reproduction)

toughness—not in terms of a transition temperature but in terms of the stress levels applied to the pressure vessel. Incorporated in this methodology is a recognition that all real structures contain discontinuities that can produce brittle fracture under the right conditions. Fracture mechanics analyses enable pressure vessel manufacturers and users to ascertain the conditions under which a given discontinuity, in combination with a stress field in which it resides, will lead to brittle fracture. This results in three positive benefits:

1. It allows a rational estimation of the safety of a pressure vessel in which a defect is found.
2. It predicts behavior when vessel operation conditions, such as temperature or pressure, are changed from those originally intended.
3. It allows a rational means by which to establish nondestructive testing criteria.

Since it allows prediction of which flaw sizes are likely to be critical and which are not, it is then possible to establish the standards that are required for a nondestructive examination. In the years ahead this methodology will continue to be useful in the establishment of more realistic and reasonable codes for the operation and inspection of pressure vessels.

The application of this methodology has, however, not been straight-forward. The quenched and tempered materials frequently used for pressure vessels today are relatively high in toughness. The measurement of the fracture toughness of these materials is presently at the limit of our established technology. A series of new techniques for the determination of the appropriate fracture toughness parameters has been developing over the last twenty years. This development period is not complete as toughness in the elastic-plastic or plastic regime where most pressure vessels operate is an area of current PVRC study.[9] Examples of some fracture toughness measurements taken from a PVRC investigation[5] are seen in Figs. 8 and 9.

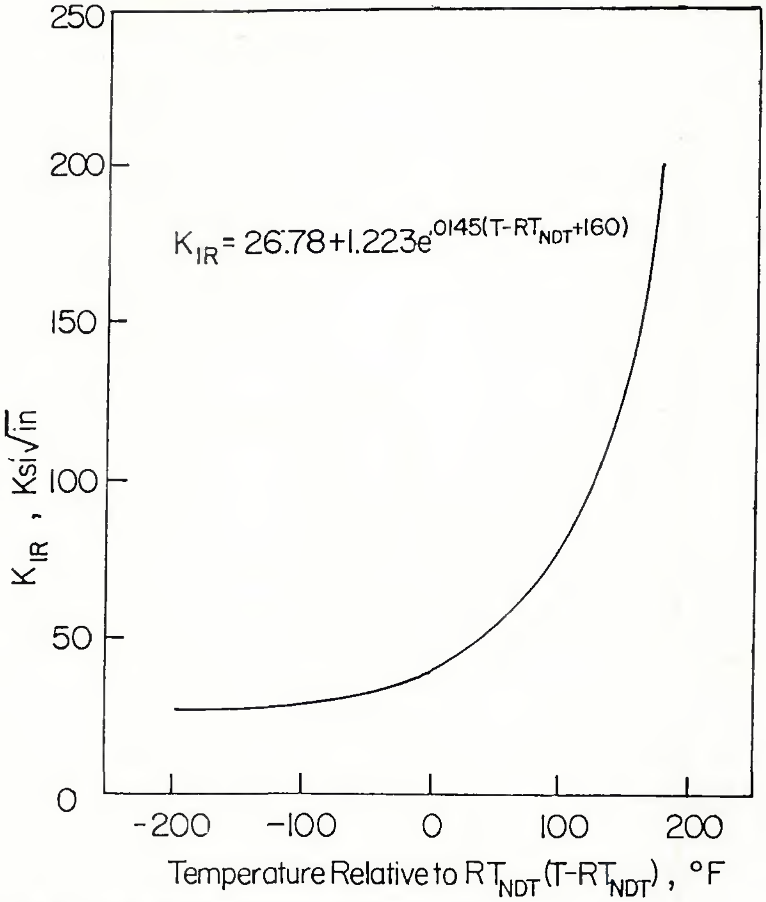

Fig. 7—The K_{IR} fracture toughness reference curve

Two comments about these tests are in order. The first relates to the difficulty in obtaining the information. Anyone who has attempted to perform fracture toughness tests realizes that specimen preparation is expensive. Specimen sizes are relatively large compared to the more common Charpy test, and preparation of the specimen is more difficult. Indeed, it is quite possible after preparing such a specimen and going through an extensive testing procedure to find that the data obtained are not useful.

Another comment is the substantial difference between static and dynamic fracture toughness. It has long been known that dynamic loading conditions enhance the capability of the material to fail in a brittle manner. The utilization of both static and dynamic tests, however, has demonstrated that defining of the conditions of service is important. If we utilize dynamic toughness curves for all conditions of service, we take a substantial penalty

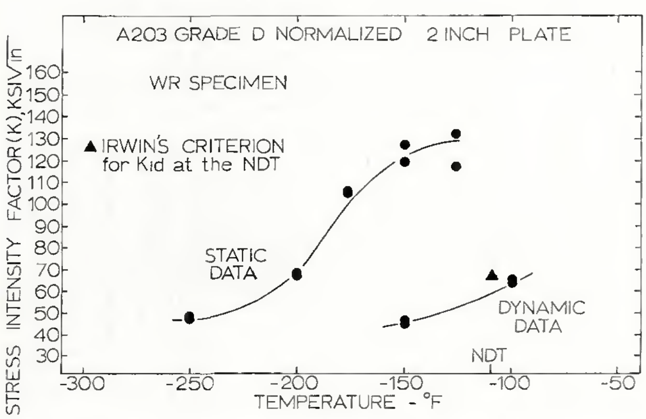

Fig. 9—Fracture toughness data for A203 Grade D quenched-and-tempered steel

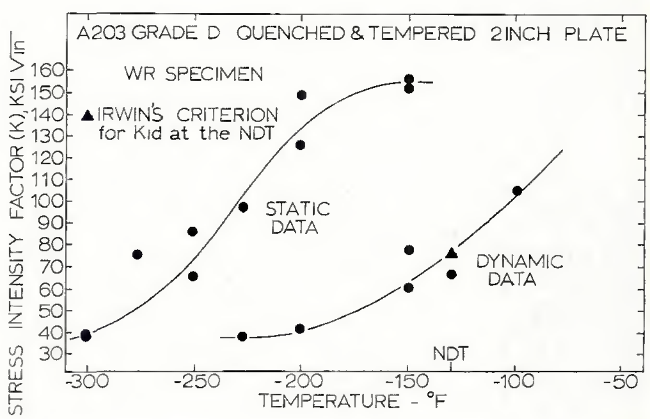

Fig. 8—Fracture toughness data for A203 Grade D normalized steel

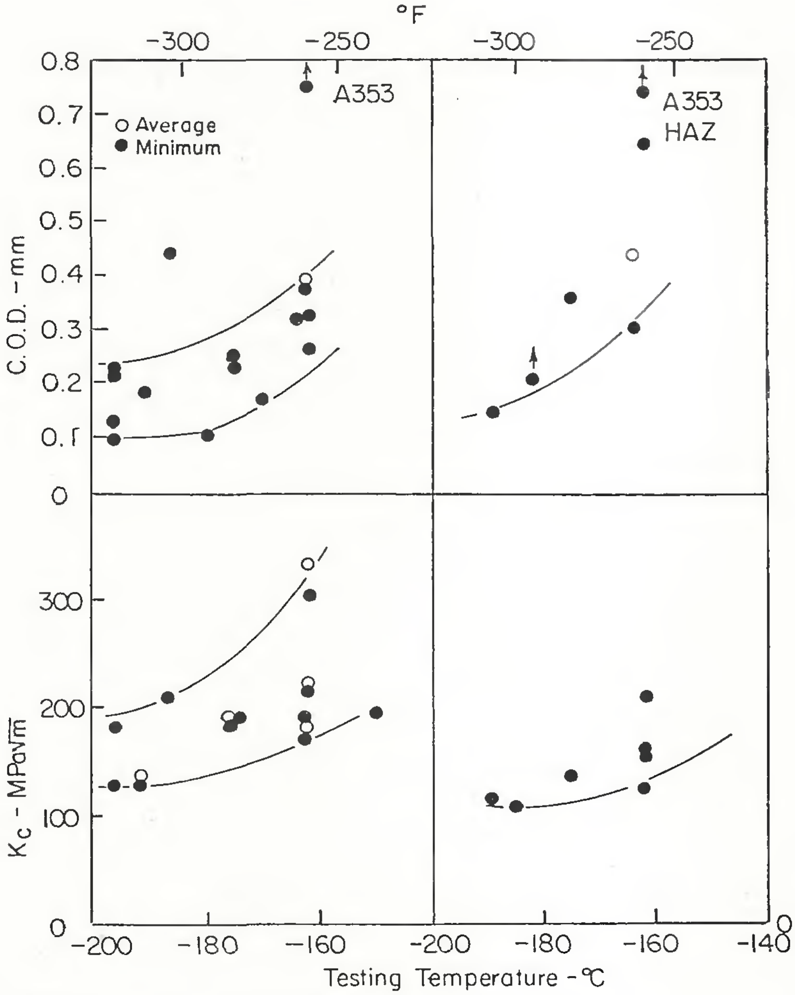

Fig. 10—Fracture toughness of A553 plate and weldments

as compared to using static curves in which the toughness is much higher at a given temperature. For example, at −200°F (−129°C) the difference between static and dynamic fracture toughness is very great—a factor of at least 3 or 4 for the quenched and tempered A203D. On the other hand, at temperatures above 100°F (37.8°C), static and dynamic fracture toughness are both quite high and the conditions of loading are relatively unimportant.

Recognition of this loading rate effect has been included in a number of codes designed to control fracture toughness in structures other than pressure vessels. It remains to be seen whether such considerations will be applied in the pressure vessel industry. Comparison of Figs. 8 and 9 show how heat treatment can influence fracture toughness.

The tests required to measure fracture toughness directly are relatively expensive. Because of this, there has been substantial interest in the last 10 years to utilize the large body of data obtained from other tests, such as the Charpy and drop weight test, to infer fracture toughness information. This procedure has been applied to pressure vessel and structural steels with some success and, thus, there is hope for the Charpy test yet. It may reasonably be expected that changes will continue to occur in this field.

An example of another difficulty which is experienced in obtaining fracture toughness data is illustrated in Fig. 10. Here fracture toughness data from a series of investigations on A553 cryogenic 9% nickel steel are plotted on a common axis. Two fracture toughness parameters, the K_c value which is familiar with American investigators and the COD value used extensively in Great Britain and parts of Europe, are both plotted as a function of testing temperature. The wide scatter of data obtained over a series of investigations and indeed, in the same investigation, illustrates the difficulties in making consistent fracture toughness measurements. Some of this scatter appears to be inherent in the testing method itself; thus, in our fracture

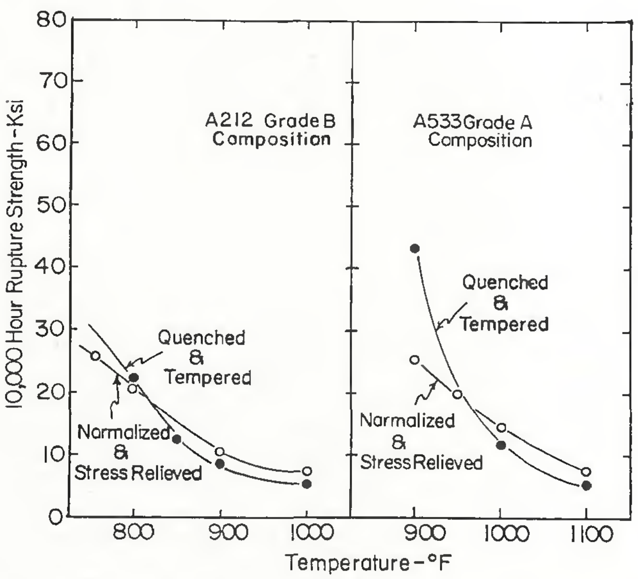

Fig. 11—Stress-rupture behavior of normalized and quenched-and-tempered A212B and A533A steels

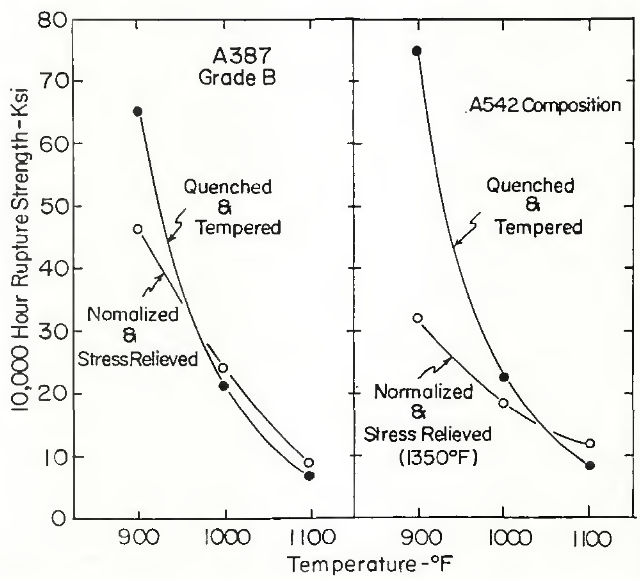

Fig 12—Stress-rupture behavior of normalized and quenched-and-tempered A383 Grade B and A387 Grade D steels

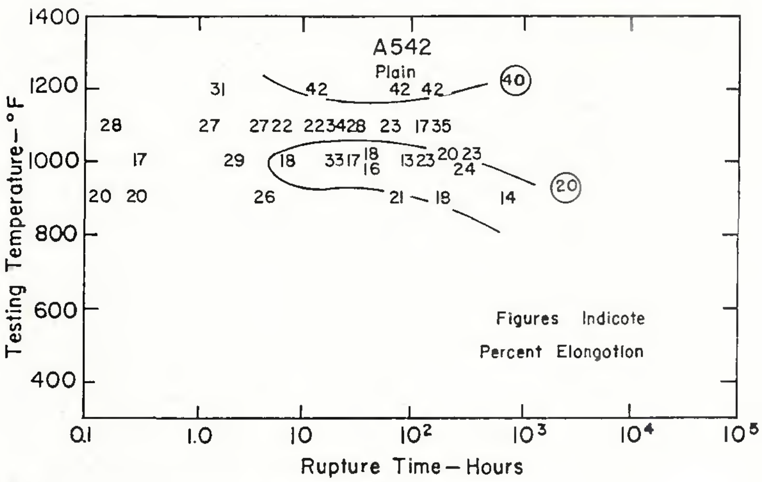

Fig. 13—Rupture ductility diagram for A542 steel

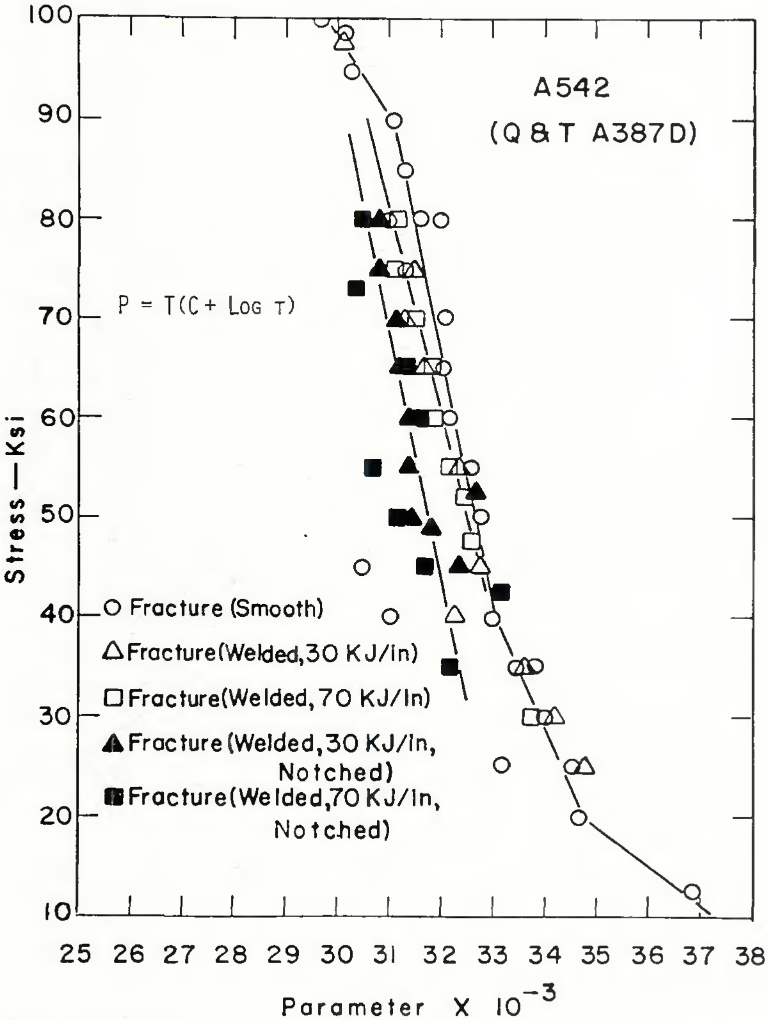

Fig. 14—Summary stress-rupture data for A542 steel

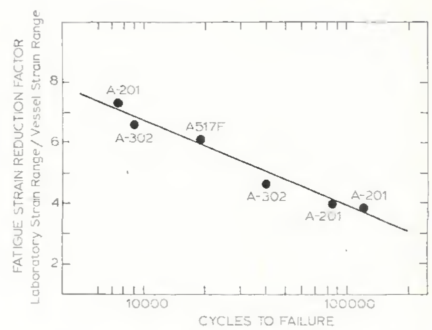

Fig. 15—Fatigue strain range reduction factors for pressure vessel steels

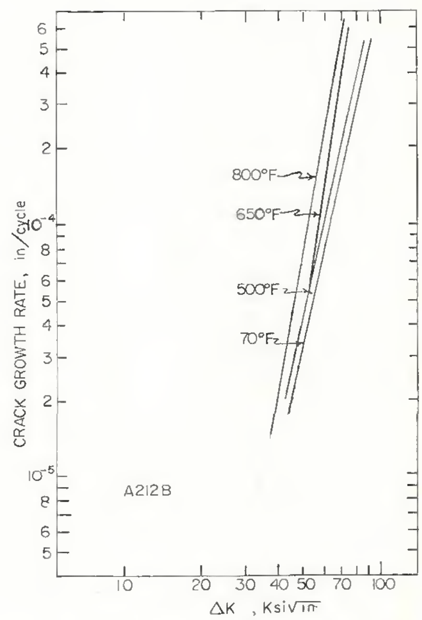

Fig. 16—Crack growth rate data for A212 Grade B steels

control programs we are forced in many instances to use minimum values of fracture toughness which are substantially less than the average values.

The creep rupture properties of pressure vessel steels were given much detailed study during the 1960's with special emphasis on the Cr-Mo grades. Creep properties of these materials were developed in terms of standard creep-rupture curves and rupture ductility curves over the range of service and, indeed, in many cases somewhat above the range of service intended for these materials. The results of some of these studies[2,3,4,10] are seen in Figs. 11 and 12.

For the materials that are used in this comparison, two materials are rather low in creep resistance—A212 Grade B (C-Mn steel) and A302 Grade B (C-Mn-Mo) steel. Included are two materials of higher creep resistance intended for intermediate and higher temperature service—A387 Grade B and A387 Grade D. In all four cases, these materials were heat treated in two conditions—the quenched and tempered, and the normalized and stress relieved.

It was learned that the traditional view that normalized and stress relieved or annealed materials will have better creep resistance is true to a limited extent, particularly in the temperature range where multiple creep processes can occur. But in the temperature range over which most of these materials are used, i.e., below 800°F (427°C) for A212 Grade B and

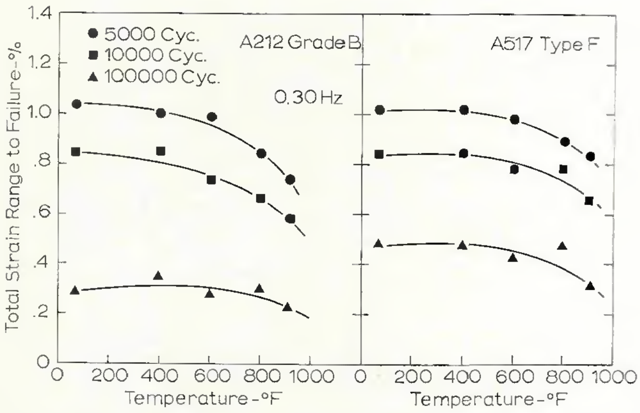

Fig. 17—Fatigue strain range to failure data for A212 Grade B and A517 Type F

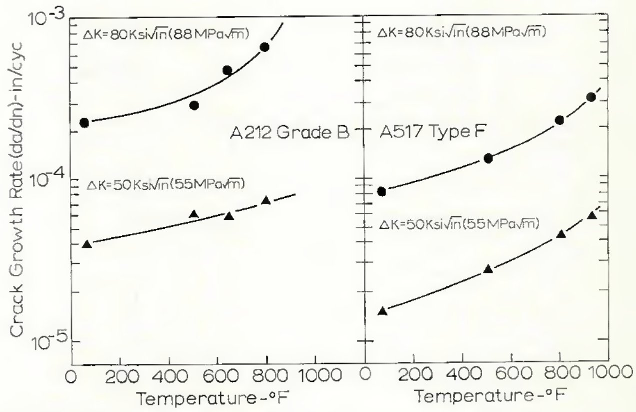

Fig. 18—Fatigue crack growth rate data for A212 Grade B and A517 Type F

950°F (510°C) for the other steels, it may be anticipated that the quenched and tempered condition will not be significantly lower in creep resistance. The advantage is that at lower temperatures the improvement in strength produced by quenching and tempering offers a positive benefit and, indeed, in the ambient temperature range produces better impact toughness as described before.

The decline in rupture ductility for these same materials during elevated temperature service is a matter that must be taken into account. Figure 13 shows rupture ductility as a function of testing temperature and time for A542 material. It will be observed that minimum ductilities can be experienced after 1000 hours (h), and failure in an essentially brittle manner could occur during creep service.

For some materials low ductility failure has been especially accentuated in weld heat-affected zones. This is a problem that appears not only during creep service but also to some extent during the stress relief of certain alloys. This is discussed in this paper under the topic of weldability.

To provide an example of how these investigations examined a whole myriad of factors in order to provide data that is reasonably useful to the pressure vessel designer, a summary of data obtained for quenched and tempered A387 Grade D (A542) is shown in Fig. 14. Here it will be seen that not only are the properties of the base material in smooth bar tests of importance in vessel design, but the behavior of notched specimens and of weldments is also important. The data are presented in the form of stress for rupture vs. a parameter combining the temperature and time at which failure occurs. This parametric form of data representation allows extrapolation of test data to other temperatures and times. The resistance of the notched and welded specimens approximates pressure vessel service more accurately than does smooth bar testing.

Perhaps the greatest change in mechanical property characterization has come in the area of fatigue design. At the beginning of the era, fatigue design was based on extensive cycle life (S-N) testing of plates, forgings and weldments. By the early 1960's a substantial amount of low cycle fatigue data obtained at Lehigh University and other laboratories under the sponsorship of the Pressure Vessel Research Committee was amassed.[11,12] However, the results of joint investigations at Lehigh University and the Southwest Research Institute as illustrated in Fig. 15 show that, although strain ranges could be measured in the laboratory to predict failure in vessels, the actual vessel life was reduced from the laboratory prediction by a factor of as great as 8 and no less than a factor of 4. It was clear that laboratory tests

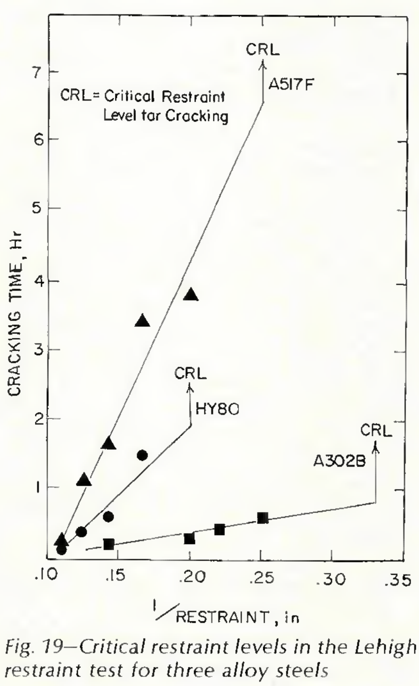

Fig. 19—Critical restraint levels in the Lehigh restraint test for three alloy steels

would not predict vessel failure unless some understanding of this discrepancy could be obtained.

In subsequent research programs[13] it was found that the answer was that pressure vessels, no matter how carefully manufactured, always contained a population of small discontinuities and that the fatigue life of the vessel was controlled by the growth of these discontinuities rather than by crack initiation and growth such as is measured in typical S-N curves. Thus, the laboratory data had to be reduced by a factor proportional to the crack initiation portion of the full life curve such that only the crack propagation portion would be applied to the pressure life. Consequently, emphasis in pressure vessel fatigue research shifted away from full life tests to crack growth tests. As a result, the Pressure Vessel Research Committee engaged in crack growth studies on pressure vessel steels both at ambient and elevated temperatures; this was done to complement full life curves that had been developed in the previous decade with relevant crack growth data.

An example of one such summary curve is seen in Fig. 16 where the crack growth rates for A212 Grade B are shown as a function of temperature. It should also be noted that emphasis has shifted to a fracture mechanics characterization of crack growth—that is to say the growth rate (da/dn in in./cycle) is plotted against ΔK, the stress intensity factor at the tip of the growing crack.

The increasing use of fracture mechanics not only to characterize the fracture toughness of materials but also to characterize crack growth has been one of the most remarkable changes in the characterization of materials that have taken place in the last 20 years. It will be noted that elevated temperature has a significant influence on crack growth and that the

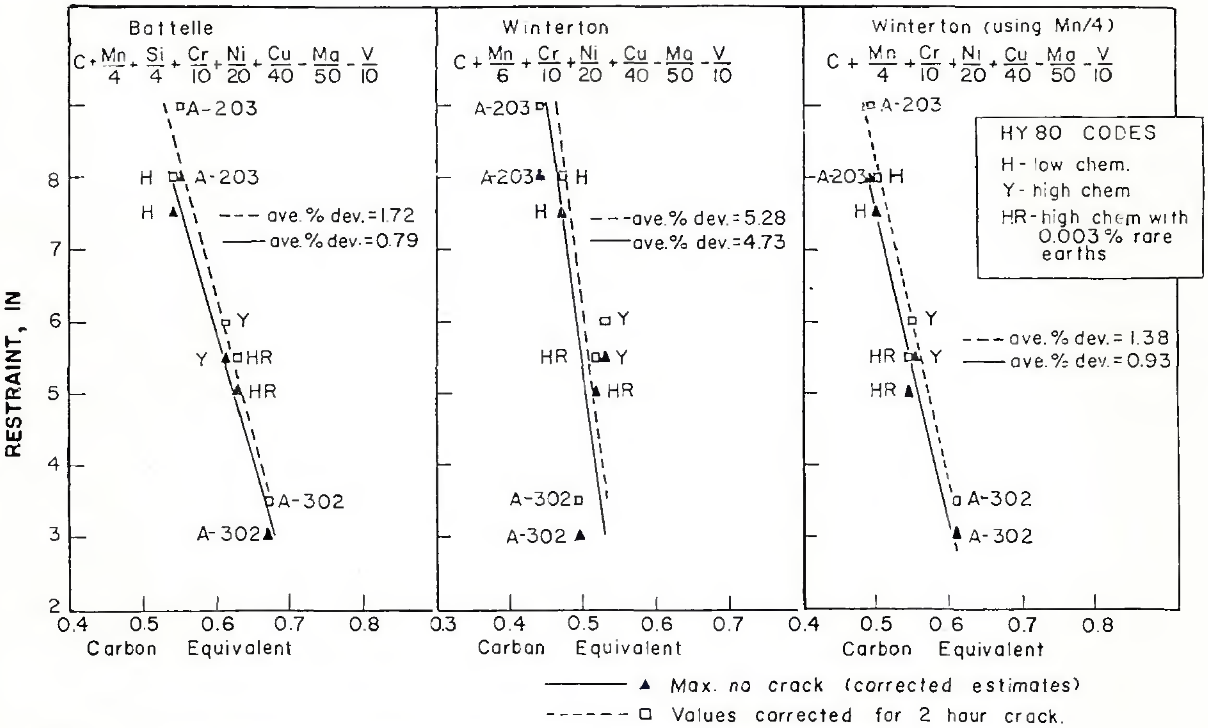

Fig. 20—Comparison of carbon equivalent formulas with Lehigh restraint test cracking data

growth rate may increase by a factor of 5 as temperature is increased from room temperature to 800°F (427°C).

Figures 17 and 18 show the decrease in total strain range to failure for two pressure vessel steels between room temperature and 900°F (482°C) based on full life testing can at least in part be attributed to increases in crack growth rate over that same temperature range. The summary of all these investigations[15] showed that temperature and to a lesser extent, cycle rate, had an influence on fatigue in pressure vessel steels but only over about 600°F. (316°C). Environmental effects do need to be taken into account when we consider the behavior of the steels in fatigue service at 800 or 900°F (or when in some aqueous solutions).

Although fatigue in many pressure vessels is only a small factor in their life, the increasing use of fracture mechanics to analyze fatigue provides a powerful tool in the control of fatigue failure. Because the growth of defects is directly related to the stress intensity factor at the crack tip, it is possible to estimate the size of the discontinuity that will begin to grow under fatigue conditions and to determine which defects are and which are not significant for fatigue service as well as brittle fracture resistance.

Fabrication and Service Weldability

It is not practical to propose a steel for pressure vessel service without giving careful consideration to the problems of weldability and fabricability that the steel may entail. For this reason, weldability may be just as important as properties in the unwelded state. Two aspects of weldability need to be considered: fabrication weldability and service weldability. The tendency toward highly alloyed and heat treatable steels will often result in decreased fabrication weldability. Thus the problem of delayed cracking, which was common in the higher carbon content pressure steels of the past, may reappear in these alloy steels unless adequate preheat, postweld heat treatment and care of welding consumables is observed.

The study of delayed cracking in pressure vessel steels was the subject of a series of Welding Research Council investigations in the 1960's. The results of these investigations highlighted again the role of many factors in the phenomenon of delayed cracking in steels. For example, they demonstrated the role of restraint in leading to cracking as is illustrated in Fig. 19. Restraint below a critical level was not able to induce cracking, and restraint above that level induced cracking in successively shorter time periods as the restraint level increased. Restraint in these investigations was created by the Lehigh restraint test, a development of the Welding Research Council programs at Lehigh University.[19,20]

Additional findings from the investigations showed that cracking susceptibility was proportional to the hydrogen content in the welding atmosphere and both hydrogen and moisture combined to act in a cumulative manner to produce the cracking phenomenon. Moisture contamination of a gas metal arc argon atmosphere was shown to be able to produce as deleterious an effect as cellulosic electrodes, and—not surprisingly—post-heating (if applied immediately) was shown to be effective in eliminating delayed cracking.

Studies with the carbon equivalent formulas available at that time indicated they could be an effective means for predicting cracking tendency. As shown by Fig. 20, some carbon equivalent formulas were more effective than others in predicting cracking. The investigations also showed that the prior microstructure of the steel to be welded and the composition of the welding electode (within the normal range used for a given steel) had a relatively minor influence on heat affected zone cracking compared to the major factors, i.e., the presence of hard martensite in the heat-affected zone, the presence of hydrogen or moisture in the welding atmosphere and the presence of restraint in the welded joint.

Less a recitation of these factors taken from investigations performed in the 1960's appear to be a message that needs not to be repeated in the 1980's, problems of delayed cracking in weldments continue today and many of the lessons learned in the past

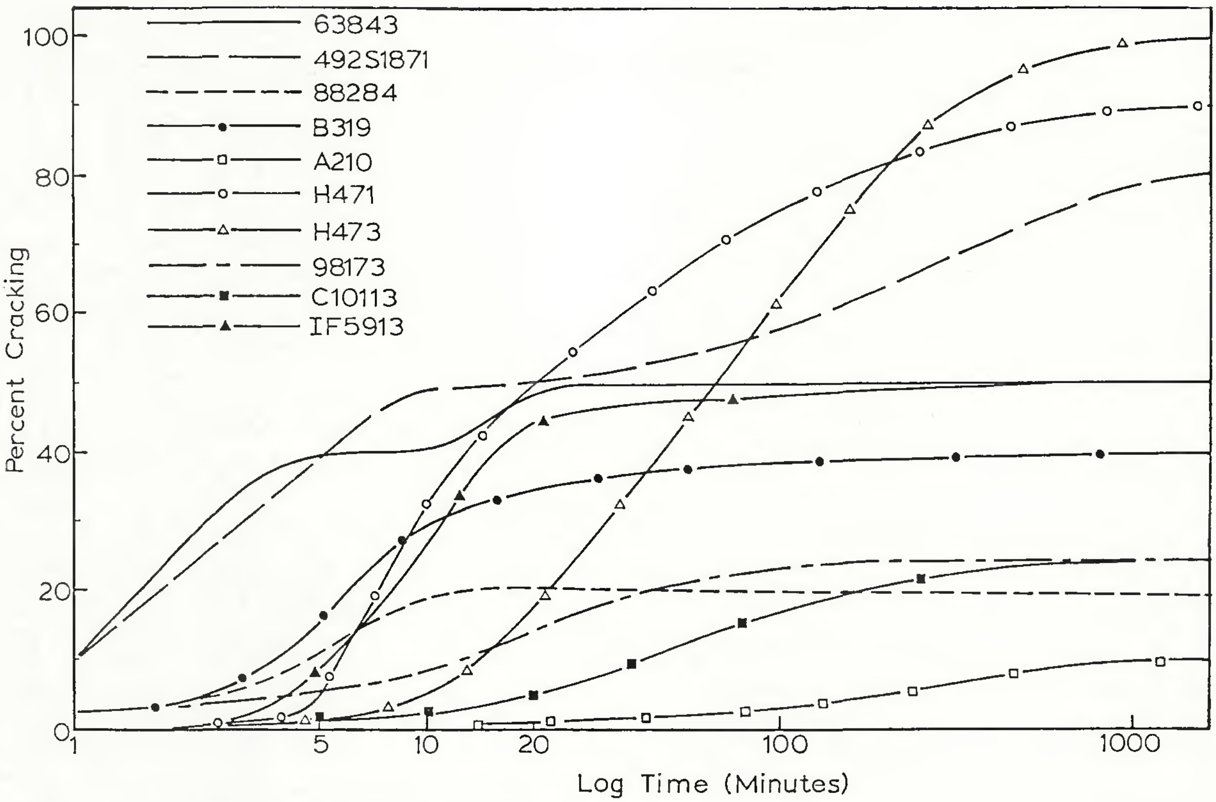

Fig. 21—Cracking of pipeline steels in field weldability test

20 years have apparently not been universally applied in the pressure vessel industry. Hydrogen induced cracking problems have been a special study of the Weldability Committee of the Welding Research Council during this 20 year period. Their more recent work has involved testing of materials used in pipelines and has resulted in the development of an on-site pipeline field weldability test.[21] Figure 21 shows some data obtained in this investigation for line pipe steels and demonstrates the ability of the test to measure the extent and time to weld cracking.

While much of our basic knowledge about hydrogen induced cracking had already been developed, at least in part, 20 years ago, our knowledge concerning solidification cracking has materially advanced during this last 20 year period. University research, especially at Rensselaer Polytechnic Institute, has established that segregation occurring during solidification is a controlling factor on the extent of hot cracking in welds and heat affected zones. This work has established the principles of weld consumable selection, base metal composition and process control, that will produce crack-free welds.

The critical role of manganese-to-sulfur ratio in plates to prevent heat affected zone hot cracking (microcracking) was the subject of a Welding Research Council study[22] and is illustrated in Fig. 22. Here it is demonstrated that manganese-to-sulfur ratios in excess of 60 are required if base plate microcracking is to be effectively controlled in alloy steels. The same principles apply to weld metal hot cracking; however, in this instance, the carbon content of the weld metal is lower and, as Fig. 22 indicates, lower carbon contents produce less cracking at the same manganese to sulfur ratio.

Some fabrication weldability problems are encountered after welding during postweld heat treatment for short times at 900 to 1200°F. (482 to 649°C). In this temperature range, as-welded materials undergo creep deformation in order that residual stresses produced during the welding process may be relieved. In some materials this stress relief process is accompanied by changes in the

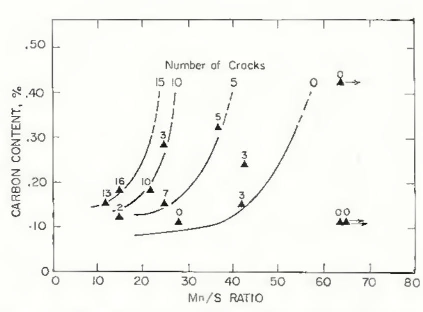

Fig. 22—Influence of Mn/S ratio on hot cracking of pressure vessel steels

microstructure, particularly in weld heat-affected zones. The result is a cracking phenomenon known as stress relief cracking.

This problem was first recognized as an adjunct to the creep rupture investigations performed on welded materials.[3,4,23] This is well illustrated in Figs. 23 and 24. Here Lehigh restraint specimens made of two different steels are welded and then given stress relief treatments. As Figs. 23 and 24 show, there is a tendency for these materials to undergo heat-affected zone cracking during this stress relief cycle. For A517-F, stress relief cracking can be severe and thus heat-affected zone cracking may occur during stress relief treatments or, if stress relief is not applied, cracking may occur during service at 800°F (427°C) or above. For A517-J, cracking is more moderate and stress relief treatments can be applied without extensive cracking phenomena. The nature of this cracking is intergranular within the heat-affected zone of the weldment.

Another fabrication weldability problem that has surfaced in the last 20 years is the occurrence of lamellar tearing in highly restrained joints. With the recognition of this type of welding defect as a significant problem in some structures, research programs were developed to determine its cause and allow appropriate solutions. The ultimate result of these studies has been a trend to plate materials of

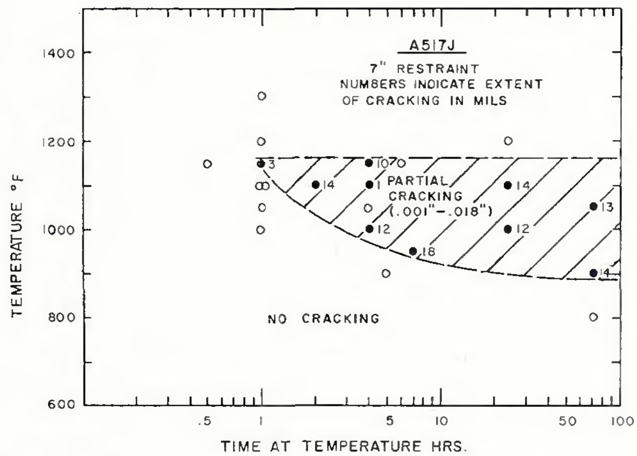

Fig. 23—Stress relief cracking in A517 Type J steel

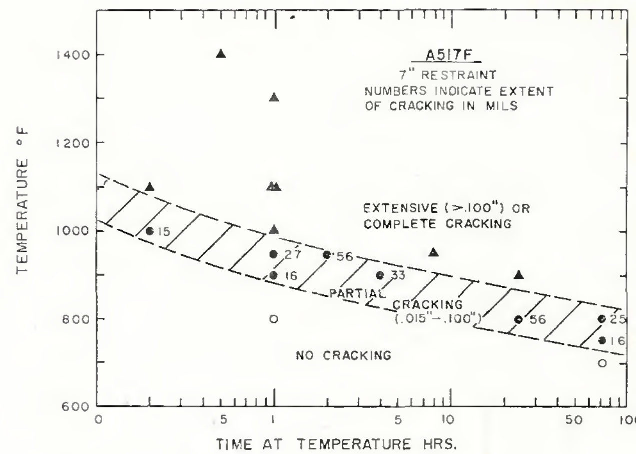

Fig. 24—Stress relief cracking in A517 Type F steel

higher cleanliness, lower inclusion counts and modified inclusion morphologies. From the weldment design standpoint, joint configurations have been modified to reduce through the thickness stresses that cause tearing. The role of process variables that promote tearing is currently under study.[24,25] Thus far, hydrogen and moisture contamination of the arc has again been identified as a contributing factor as well as joint restraint and material condition.

Another current area of fabrication weldability research is the study of cracking in electroslag welds. The impetus for this study has come from cracking observed in bridge steel weldments and, while these are not pressure vessel steels, there are some valuable precautions concerning electroslag welding of C-Mn structural and low alloy steels that can be learned from these investigations. It is clear that hydrogen introduced into weld metal during the welding process is a cause for some of the cracking observed.

Fabrication considerations do not consist only of problems in the weldability of steels. Hot and cold formability are also of importance. During the 1960-1980 era, the cold forming and aging characteristics of carbon and alloy steels have been the subject of several studies,[26,27] and have been examined from different viewpoints. It is clear from this work that strain aging during fabrication and service is a possibility in modern pressure vessel steels, and can result in measurable shifts in transition temperature.[26] On the other hand, recent research[27] indicates that, when the service temperature is well above the transition temperature (even in the embrittled state), strain aging will have a minimal effect on service behavior. As a sidelight to this phenomenon, studies have shown that warm overstressing can be an effective means to raise effective fracture toughness at and below the prestressing temperature; although subsequent aging of the prestressed material can reduce the effectiveness of prestressing, the overall effect is usually beneficial. Figurė 25 illustrates these effects in A516 and A533 Grade B steels.

Service weldability problems over the last 20 years have centered on the need for a balance of properties between the weld metal, heat-affected zone and base metal in a welded joint. Studies on the effect of welding variables such as heat input and postweld heat treatment and on the influence of welding consumables have been performed on weldments of all types.[28-31] Research programs have ranged from electroslag welded C-Mn (A537) steel[28] to high nickel cryogenic alloy gas-metal-arc welded with austenitic fillers.[29] These studies continue today on microalloyed steels for moderately low temperature service. In many of these studies, one particular region of the weldment had lower toughness than those that surround it. In some cases this region has been the heat-affected zone, but not always so. Moreover, postweld heat treatment may or may not be helpful with respect to improving the toughness of the poorer zone, or it may damage one zone while improving another.

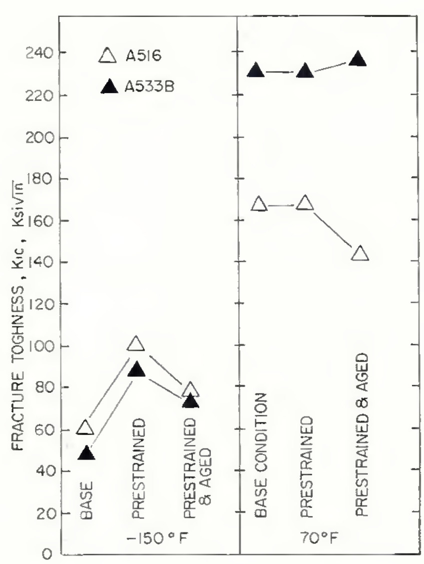

Fig. 25—Effect of prestressing and aging on the toughness of A517 Grade 70 and A533 Grade B Class 1 steel

This principle was clearly illustrated in the fracture toughness study of weldments of three of the materials studied during these PVRC investigations—A537, A542 and A517 Type F, all in 2 in. (51 mm) thickness. Figures 26 and 27 show the results of investigations of these materials. The weld metal, heat-affected zone and base metal were sampled in these studies. As may be seen, the relative toughness of these zones varies from material to material. The use of low welding heat input was not always helpful for the toughness of the weld metal nor was postweld heat treatment always beneficial to toughness.

For very high heat input processes like electroslag welding, heat treatment by normalizing after welding may offer improvement of badly damaged heat-affected zones. In these studies, admittedly limited, for the A542 material the weld metal was the weakest link—that is to say was the region of lowest fracture toughness, while for the A517-F material it was the base metal. For A537, when electroslag welded, it was the weld metal and heat-affected zones which were the regions of lowest toughness. A similar comparison is already illustrated in Fig. 7 where the base metals in heat-affected zones of cryogenic steels are compared with respect to fracture toughness in the low temperature regime of service.

Unfortunately, there is presently no substitute for actual weldment tests to determine how a given weldment will behave. In this regard, the lack of a

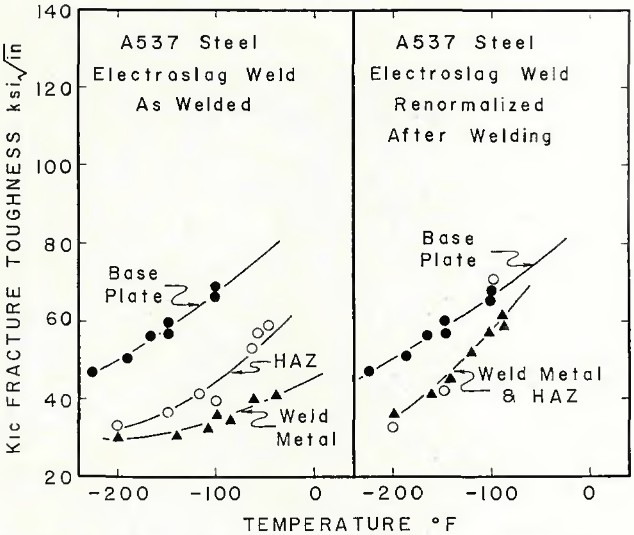

Fig. 26—Fracture toughness of weldment zones for A537 Type 1 electroslag welded steel

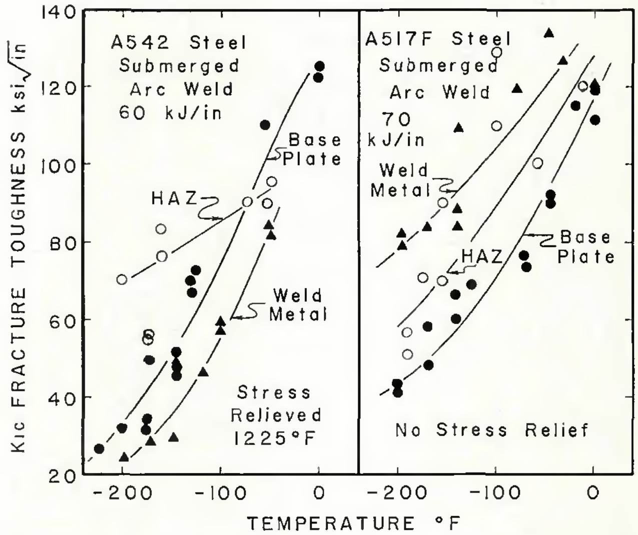

Fig. 27—Fracture toughness of weldment zones for A542 and A517 Type F steels

clear cut and widely accepted weldment test has been and continues to be a problem. Testing of individual small regions of a weldment—for example, Charpy impact tests or compact tension tests notched in, say, the heat affected zone—will not necessarily reveal the overall behavior of a weldment in service. Attempts to sample the composite weldment fracture behavior have not been entirely satisfactory, and the Pressure Vessel Research Committee has sponsored research in the area. More research is clearly needed.

Service Related Effects

In the period between 1950 and 1960 a substantial amount of work was done at Lehigh University on the effects of service environment exposure on the properties, particularly the toughness, of pressure vessel steels.[32] These investigations have continued, although the extensive study done between 1955 and 1960 has not, of course, been duplicated. One of the areas of special concern is the influence of temper embrittlement on the properties of steels designed for elevated temperature service. These studies[33,34] coordinated with the studies on the creep rupture characteristics of these steels. Although temper embrittlement is not a problem during actual service, the temperature range is such that temper embrittlement can and probably will occur, and the toughness of the steels in the ambient temperature range after extended service becomes a matter of concern.

Some results from studies on A542 steel are illustrated in Figures 28 and 29. Here it is shown that some pressure vessel steels used for elevated temperature service above 800°F (427°C) will increase in transition temperature in time periods beyond 1000 h. Thus, a material which initially has a transition temperature well below ambient can, after service exposure, suffer a significant loss in toughness in the ambient range.

Investigations on A542 steel with different heat treatments show that materials that have been cooled at slower cooling rates from the austenitizing temperature (for example, the simulated 6 in. (15.2 cm) plate in Fig. 29) appear not to have embrittled very substantially during the first 1000 h of their life; however, exposure beyond a 1000 h can produce substantial losses. As can be seen from Fig. 29, the 1 and 6 in. (2.54 to 15.2 cm) thick plates approach each other in toughness for a specific time period of about 500 h. Beyond this period, the finer microstructure (1 in.) plate begins to saturate in terms of embrittlement and further embrittlement is slight. However, the coarser microstructure (6 in.) plate continues to embrittle and, thus, will eventually have a transition temperature well above ambient after periods of service beyond 1000 h.

Another area of study has been the strain-aging characteristics of steels, already mentioned under fabrication effects, but which also relates to service. This was extensively studied for pressure vessel steels in the 1950-60 period[32] and to a limited extent between 1960-1980. With the advent of microalloyed steels, a new program of study of the strain-aging characteristics of these steels has also been undertaken.

Future Developments

If we look at the future, there are some areas of change and development that will likely influence the pressure vessel industry. As far as materials are concerned, the increasing demand for steels with assured fracture resistance in even moderate types of service will place a new emphasis on control of toughness for these applications. For this reason, materials with less controlled toughness such as A285, A515 and other typically coarse gained and as-rolled plate products will be displaced by materials with higher assured toughness, i.e., higher Mn/C ratios and grain size control.

As already mentioned, the microalloyed steels may be an effective alternative in providing both higher strength and higher toughness, although they will be somewhat more expensive. Quenched and tempered steels have, at the present time, reached a plateau where increased strength or hardenability will be obtained only at the cost of weldability.

Figure 30 illustrates the progress with these types of steels over the 1960-80 period. Using a pressure vessel 12 in. (30.5 cm) thick made of A533 Grade B Class 1 as a reference point (a thick walled nuclear vessel), the thickness of a similar vessel made of higher or lower strength steel using a typical code parameter (one-third of the tensile strength) is shown. Heat treatment of these typical heavy section steels from the class 1 strength level (as they are now for the most part) to the class 2 strength level results in only marginal decrease in thickness. Substantial increases can be achieved at this point only by using materials such as HY-130 (a navy hull steel) of much higher cost and lower weldability. Further devel-

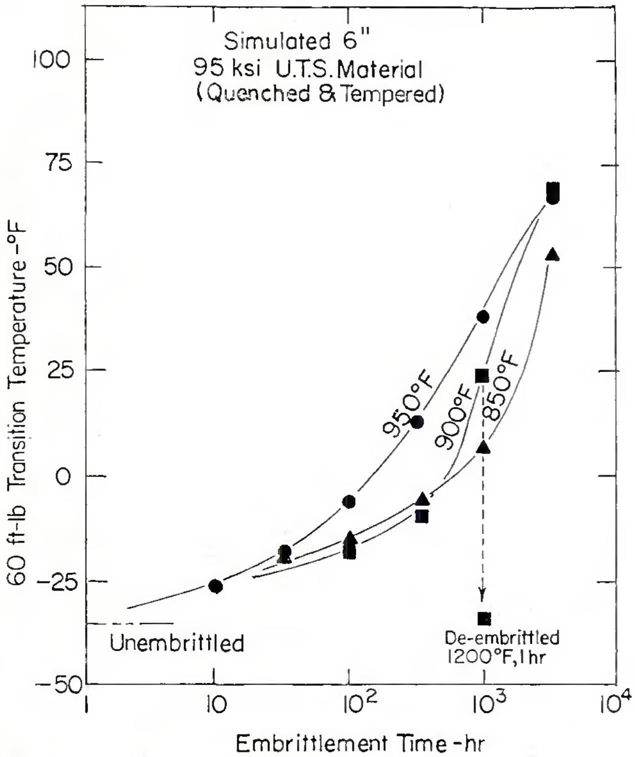

Fig. 28—Temper embrittlement characteristics of heavy section A542 steel

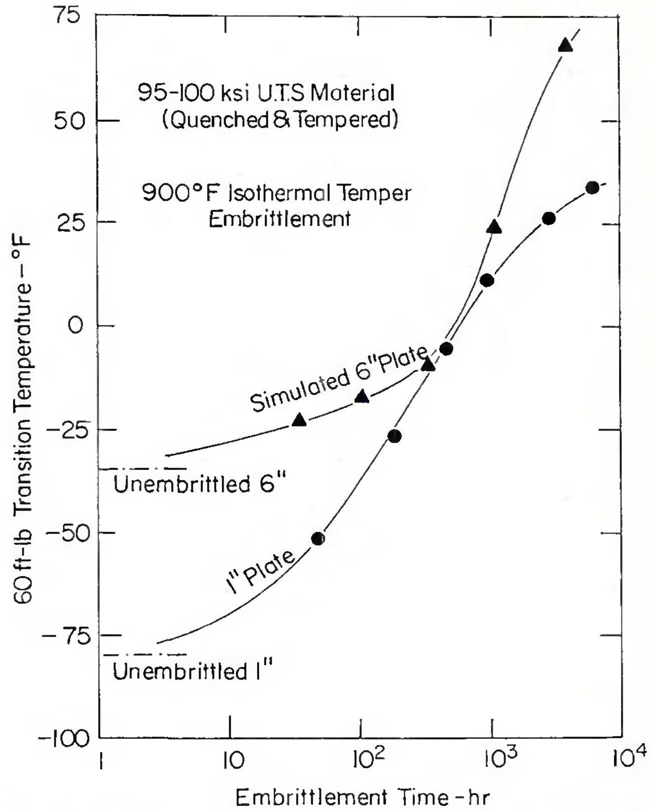

Fig. 29—Temper embrittlement comparison between light and heavy section A542 steel

opments in these steels will probably be limited in the immediate future.

If alternate energy sources, particularly coal gasification, attract substantial development funding in the next five years, the Cr-Mo steels will see use in new applications and very large pressure vessels of these steels—for example, A387 type 22—will be constructed. The economics of these processes is not so certain; development will probably be slow but steady. Similarly, the use of high nickel steels for liquified natural gas transport and storage is a possible growth area, but the instability of the political situation will make this development cautious also.

From the standpoint of mechanical property assessment, fracture mechanics methodology will continue to grow in application and will be the fracture control philosophy that predominates in codes of the 1980's. The procedures will change as far as the measurement of the fracture toughness parameters is concerned, but the basic philosophy will prevail pretty much as it is now for some time. Advances in methods for nondestructive testing and defect analysis, already seen in the 1970's, will make the fracture mechanics approach more effective. In terms of actual mechanical properties, it appears that changes will be incremental in the near term; the primary emphasis will be on cost effectiveness of a steel for a specific type of application rather than simply higher strength or better toughness as a general goal.

Finally, weldability and service environmental effects will probably require some careful study and research in the 1980's. Many new welding processes utilizing the concepts of depositing a minimum amount of weld metal using a narrow-weld-width automated system will be perfected. Some are already in use today. Weldability

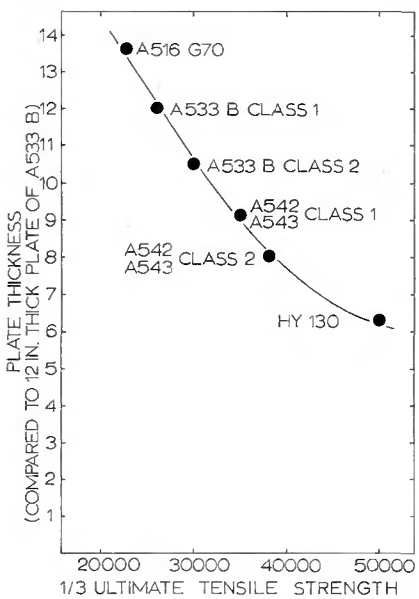

Fig. 30—Pressure vessel thickness as a function of strength for typical medium and high strength steels

problems such as hot and cold cracking, which are still with us today, will need to be evaluated and controlled with respect to these new processes. Even currently used processes are often not well understood, and cracking problems associated with these processes need further study and control.

Service environments that will require the greatest study are probably those associated with alternate energy sources, particularly utilization of high availability but lower quality fossel fuels. Recovery of these resources will place substantial temperature, corrosion, and erosion requirements on materials; these will present some real challenges to producers and fabricators of pressure vessels in the years ahead.

Summary

Twenty years of welding research, even in a restricted field and at one institution, is difficult to summarize in one paper. During this time period we have seen men travel to the moon and return, have seen nuclear power grow and in recent days come under fire, and we have seen a revitalization of alternate energy concepts. There is scarcely a field in which changes over the last 20 years have not impacted on pressure vessel technology. Research on pressure vessel steels has reflected these changes and, hopefully, provided the insights that allow advancement in pressure vessel technology.

Steels and fabrication processes used today are substantially more sophisticated than 20 years ago. More complex alloy compositions, more high technology welding processes and more involved pre- and postweld heat treatments will become the rule rather than the exception in the next 20 years. It is the hope of investigators at Lehigh University that the research investments of the past 20 years summarized here have provided the dividends of a sound technical basis for future developments in the pressure vessel field.

Acknowledgment

The author gratefully acknowledges the assistance of the Fabrication Divisions of PVRC and Effects and the Pressure Steels Subcommittees of these divisions. They have not only financially supported this work but have also participated in the research programs with suggestions, guidance and direct cooperation without which many programs could not have been completed. The research efforts reported here have truly been the joint effort of an industry-university team of which the author considers himself privileged to have been a part.

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