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THE NOTCH TOUGHNESS OF ULTRA-HIGH-STRENGTH STEELS IN RELATION TO DESIGN CONSIDERATIONS*

by

R. C. A. Thurston**

ABSTRACT

Data for the notch toughness of ultra-high-strength steels, as found in the literature, are presented and reviewed in terms of the type of steel, the strength level, and the appropriate test parameter. Some attention is given to the effects of the more important factors involved in the processing of the steel.

Design requirements are analyzed with reference to the demands of the particular application (e.g. pressure vessel, rocket motor case, hydrofoil), the fabrication procedures involved, the environment, and the applicability of non-destructive and proof testing. A number of possible design parameters or criteria, related to the more significant laboratory toughness tests, are examined with respect to their suitability and applicability in the light of present knowledge regarding ultra-high-strength steels. Where possible, the examination is complemented by a comparison with the results of service performance and/or medium or full-scale laboratory tests under simulated service conditions.

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RELATIONS ENTRE LA TÉNACITÉ À L'ENTAILLE EN ACIERS TRÈS RÉSISTANTS ET LES PRESCRIPTIONS TECHNIQUES*

par

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RÉSUMÉ

L'auteur présente ici les données relatives à la ténacité à l'entaille en aciers très résistants, telles que la documentation pertinente les fournit. Ces données sont examinées en fonction du type de l'acier, de son niveau de résistance et des paramètres convenables établis lors des essais. L'auteur apporte quelque attention aux effets des facteurs les plus importants intervenant dans l'élaboration de l'acier.

L'auteur analyse les prescriptions techniques en fonction des exigences d'utilisation dans la pratique (par ex. les autoclaves, les enveloppes de moteurs-fusée, les bâtiments à ailes portantes); des procédés d'élaboration utilisés; du milieu; des possibilités d'essais non destructifs et de 'proof testing'. Il examine, à la lumière des connaissances actuelles sur les aciers à très haute résistance, les avantages et les possibilités d'application d'un certain nombre de paramètres conceptuels, ou critères, se rapportant aux plus importants tests de ténacité des laboratoires. Partout où c'est possible, l'étude est complétée par une comparaison avec les résultats obtenus au cours de l'utilisation réelle et avec les résultats des tests de laboratoire à échelle moyenne ou à échelle normale sous des conditions d'utilisation simulant la réalité.

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1. INTRODUCTION

During the past ten years, rapid strides have been taken in the development and application of ultra-high-strength steels. For the purposes of the present report, ultra-high-strength steels will be arbitrarily defined as those steels having a yield strength (0.2%) in excess of 200 ksi. Such steels are generally available in the form of sheet, plate, bar and forgings, and their major applications are aircraft undercarriages, pressure vessels, solid propellent rocket motor cases, and machine parts, with perhaps many additional less-known uses. At the present time, the use of ultra-high-strength steels is restricted almost entirely to those applications where the strength/weight ratio of the component or product is of prime importance, and their success or failure in these specialized fields will determine the extent of their contribution to the manufacture of engineering items of a more general nature.

The properties of the ultra-high-strength steels which are of particular interest to the user include tensile strength at various temperatures, ductility and toughness, fatigue strength, corrosion resistance, and weldability. While these are all probably of equal importance, and to some degree interrelated, the following discussion will be restricted to the notch toughness characteristics, by which is meant the reaction of the steels to the presence of stress concentrations, whether these be design discontinuities, surface cracks, or internal flaws. It is an established but unfortunate fact that as the tensile strength increases, the ductility and toughness tend to diminish. Consequently, any ultra-high-strength steel development program involves the generation not only of conventional smooth tensile test data, but also of data from notched tensile tests and V-notch Charpy impact tests. The uniaxial elongation given by the standard tensile test may be a satisfactory criterion for a medium-strength steel, but may show no correlation with the performance of an ultra-high-strength steel. Cottrell[1] has reported the results obtained from burst tests on two welded rocket motor cases of a

3% Cr-Mo-V steel. Tensile specimens heat-treated with the cases gave tensile strengths of 231 ksi and 237 ksi respectively, and the same elongation, 9-1/2%. When hydraulically pressurized to failure, the first case burst in a ductile manner at a hoop stress of over 235 ksi, whereas the second case burst in a brittle manner at 159 ksi. Hence, for the ultra-high-strength steels, it is essential to supplement the normal tensile data with the results of some type of notched test.

This necessity was recognized at an early stage, and numerous tensile and impact tests were carried out on specimens containing a relatively mild stress raiser, until it was realized that any meaningful material evaluation must include specimens with high stress concentrations, preferably a natural crack. The development of a suitable specimen was assisted by the notable work of Irwin and his collaborators[2-5], who extended the Griffith theory and derived expressions for K, the stress intensity factor, and G, the crack extension force or strain energy release rate, based on the principles of linear elastic fracture mechanics. A knowledge of the fracture toughness of a material, obtained from suitable tests, should enable a designer to determine the size of crack the material will tolerate without fracture, when loaded to a level approaching that at which it would fail by excessive plastic deformation.

In a series of reports[6-10] the ASTM Special Committee on Fracture Testing of High-Strength Materials has detailed the requirements for suitable tests on both sheet and rounds, and, more recently, a very clear exposition of the present state of the art was presented by Srawley and Brown[11]. Basically, there are two types of flat specimen used for fracture toughness tests in tension: the through-crack type, either centre-notched or edge-notched, and the partial or surface-cracked type. Satisfactory tests have also been made using a single-edge-cracked specimen in tension or in bending, but the amount of data available is still relatively small. For round bars, a circumferentially notched and cracked specimen is recommended with a minor diameter/major diameter ratio of 0.707.

Prior to the development of fracture mechanics, the majority of notched tensile tests were carried out on round specimens with a machined notch, and an indication of the notch toughness was given by the ratio of the notched tensile strength to the smooth tensile strength or the yield strength. With a very sharply notched specimen (root radius less than 0.001 in.), this test still has many adherents and is often used for screening purposes. The recommended fracture toughness tests, although intended for the estimation of plane stress or plane strain values of K and G, will also give values of the notched tensile strength.

In addition to the foregoing, there are several arbitrary empirical procedures for evaluating notch toughness which have been proved by correlation with service failure studies. Prominent among these are the explosion bulge and drop weight tests developed at the U. S. Naval Research Laboratory, the pre-cracked Charpy impact test, and the Allison instrumented bend test. Reference will be made to their particular merits in the course of the text.

2. REVIEW OF NOTCH TOUGHNESS DATA

The ultra-high-strength steels presently available can be broadly classified in the following categories:

1) Low alloy structural steels
2) Hot die and tool steels
3) Nickel alloy steels
4) Precipitation-hardening steels

Table 1 gives typical chemical compositions of those steels for which sufficient data of a satisfactory nature were available to the writer.

TABLE 1

Chemical Composition of Selected Ultra-High-Strength Steels

Steel	Composition - %							
	C	Mn	Si	Ni	Cr	Mo	V	Other
Class (1)								
AISI 4340	0.4	0.75	0.3	1.8	0.8	0.25	-	-
300 M	0.4	0.75	1.6	1.85	0.85	0.4	0.08	-
Airsteel X-200	0.4	0.85	1.5	-	2.0	0.5	0.05	-
AMS 6434	0.36	0.7	0.3	1.8	0.8	0.35	0.2	-
4137 Co	0.4	0.7	1.0	-	1.1	0.25	0.15	1.0 Co
MBMC #1	0.4	0.8	1.7	-	0.8	-	0.05	-
Class (2)								
H-11) Vascojet 1000)	0.4	0.35	1.0	-	5.0	1.3	0.5	-
D6Ac	0.44	0.8	0.2	0.55	1.0	1.0	0.05	-
Class (3)								
18% Ni Marag- ing	0.02	0.08	0.08	18.0	-	4.8	-	7.5/9.0 Co
9 Ni - 4 Co	0.44	0.3	0.1	8.5	0.3	0.3	0.1	4.0 Co
Class (4)								
PH 15 - 7 Mo	0.07	0.5	0.35	7.0	15.0	3.0	-	1.0 Al
AM 355	0.14	0.7	0.3	4.3	15.5	2.75	-	-

In an attempt to obtain a realistic appraisal of the notch toughness characteristics of the available alloys, the data contained in over 100 publications were critically examined and analyzed. Much of the data was regrettably incomplete or failed to comply with the necessary requirements for a satisfactory test, e.g., the specimen width was too small or the notch root radius too large. This criticism is not meant to imply that the tests were not perfectly adequate for the purpose for which they were intended. The remaining data were averaged for the particular steel and strength level, and are presented in Table 2.

It will be observed that the notch toughness level is expressed by two values:

(a) The notch strength ratio (notched tensile strength/ultimate tensile strength), derived from tensile tests on notched or cracked sheet or rounds, the notch root radius being less than 0.001 in.

(b) The critical stress intensity factor (K_c or K_{Ic}), derived from tensile or bend tests on pre-cracked sheet or rounds.

No data from V-notch Charpy impact tests are included since, in the ultra-high-strength range, the test is insufficiently discriminatory. Furthermore, the results obtained from small experimental heats were neglected.

TABLE 2

Summary of Notch Toughness Data

Steel	Notch Strength Ratio					Critical Stress Intensity Factor, K_c - ksi $\sqrt{in.}$				
	UTS - ksi	300	280	260	240	UTS - ksi	300	280	260	240
AISI 4340	sheet	-	0.65 (1)	0.54	0.63	sheet	-	86 (1)	200	180 (1)
	rounds	-	1.06	1.12	1.27	plate	-	35 (1)	42 (1)	57 (1)
	cracked	-	0.34	0.66	-	rounds	-	[92] (1)	[61] (1)	-
300 M	sheet	0.37	0.50 (1)	0.70	0.62	sheet	-	133	194	-
	cracked	0.29	0.40	0.58	0.55	rounds	[73] (1)	[72]	[69] (1)	[70] (1)
Airsteel X-200	sheet	0.27	-	0.47 (1)	-	sheet	85*	105	90*	125*
	rounds	-	0.61	-	0.98					
	cracked	-	0.55	-	-	rounds	[70] *	[55] *	-	[80] *
AMS 6434	sheet	-	-	0.68	0.84	sheet	>90 (1)	>120 (1)	>120 (1)	200
	cracked	0.47 (1)	0.64 (1)	0.66 (1)	0.69		[32] (1)		[54]	
4137 Co		-	-	-	-			[42]		
MBMC #1	sheet	-	0.35	0.38	-	sheet		74*	83* (1)	
H-11	sheet	0.35	0.29	0.49	-	sheet	20/100	47	90	108 (1)
Vascojet 1000	cracked	0.32	0.34	0.35	0.86 (1)		[34] (1)			
D6 AC	sheet	0.30 (1)	0.49 (1)	0.44 (1)	0.64 (1)	sheet	77* (1)	112* (1)	98* (1)	130* (1)
	rounds	-	0.74	0.85 (1)	1.05					
	cracked‡	0.56 rd.(1)	0.75 rd.(1)	-	1.00 rd.(1)	rounds	[49]	[57]	[42] (1)	[80]
18% Ni Maraging	sheet	0.78	0.87	0.90	0.92	sheet	175	217* (1)	160 (1)	203* (1)
	plate			1.40	1.48 (1)	plate	20/190	213 (1)	109	-
	cracked	0.53	0.54 (1)	0.77 (1)	0.81 (1)		[97]	[84]	[85]	
9 Ni - 4 Co						sheet		[47] (1)	[56] (1)	[88] (1)
						plate			[110] (1)	[112] (1)
PH 15 - 7 Mo	sheet	-	0.80 (1)	0.94 (1)	0.68	sheet	-	-	-	77
	cracked	-	-	0.95 (1)	.25/.71					
AM 355	sheet	0.91 (1) (CR 40%)	-	0.86 (1) (CR 15%)						

Notes on Table 2

* indicates that the fracture toughness values were obtained from notched specimens which had not been pre-cracked.

(1) indicates that the value tabulated is believed to be the result of a single test - or a small number of tests made in one laboratory on a single heat.

[ ] indicates a K_{Ic} value.

‡ indicates that the cracked NSR values were obtained from round specimens; the remainder were derived from tests on sheet specimens.

The results of the survey of available data were neither very satisfactory nor conclusive. They showed the expected trend of decreasing notch toughness with increasing tensile strength (Figure 1), though even here there were some peculiar inconsistencies, presumably due to composition or processing variables or to an insufficient number of tests. The 18% Ni maraging steel, which has been the subject of intensive investigation, is seen to possess a definite superiority over the earlier steels (Figures 1 and 2) and should obviously be of particular interest to the designer of high-strength, minimum-weight hardware. The more recently developed 9% Ni-4% Co alloy also appears to have desirable characteristics but, unfortunately, insufficient reliable information was available to the writer for a more accurate assessment. The behaviour of the remaining ultra-high-strength steels is far from consistent and varies with the strength level and parameter considered. The best of the group would appear to be 300 M alloy, and the poorest, H-11 (Vascojet 1000) and MBMC No. 1 alloys.

A comparison of the steels is best obtained by reference to those investigations in which similar tests have been made on specimens of the same form cut from sheets of the same, or approximately the same, thickness. Espey[12] has stated that the alloy having the lowest notch sensitivity varies somewhat with the strength level considered, and has reported the superiority of 300 M and D6AC steels over Vascojet 1000 at a yield strength of about 230 ksi. Davis[13] has tested a number of ultra-high-strength steels in plate form (0.3 in.) and, in addition to confirming the higher fracture toughness of the 18% Ni (250) steel, has shown D6AC alloy to be superior to H-11 alloy at a strength level of 280 ksi.

Matas, Hill and Munger[14] compared several of the alloys in the form of 0.080 in. sheet, and tentatively classified them into three groups in terms of notch strength and fracture toughness. The most desirable qualities were shown by the 18% Ni alloy (group III); the 9% Ni-4% Co alloy (group II) was somewhat inferior; and the older alloys, D6AC, 4340, 300 M and H-11 (group I), gave the lowest values. Jones[15] carried out a similar investigation on 0.180 in. specimens, and reported his results in terms of the notch strength. At a strength level of 280 ksi, both 18% Ni and 9% Ni-4% Co steels were found to be superior to 4340 (air-melt) steel, which in turn was superior to H-11 steel. An additional observation of particular interest was that vacuum-melted 4340 steel was comparable to both of the high-nickel steels, which raises the question of the importance of processing variables.

3. EFFECT OF PROCESSING VARIABLES

As the design requirements of the various types of hardware are raised and the alloys have to be fabricated to meet the higher strength levels, the beneficial or deleterious effects of processing variables tend to become of greater significance. Those variables which have a degrading influence on the toughness of the steel must be more closely controlled, while those which appear to be advantageous must be utilized to their fullest extent. In the present context, "processing" covers all stages of the manufacture of the steel from the melt to the finished stock, sheet, plate or bar, and it includes, in particular, melting practice, composition, cold or warm reduction, decarburization, and banding. The respective effects of these variables are outlined below.

(a) Melting Practice

While there is a general belief that melting under vacuum should produce a superior grade of steel because of a reduction in gas content and non-metallic inclusions, this belief is not consistently substantiated by the data from notch toughness tests of a number of alloys. The effect of vacuum melting will be seen to vary with the alloy composition and with the tensile strength level, and is not always beneficial. The results reported by Gilbert and Brown[16] for AMS 6434 alloy are typical, in their trend, of those obtained in several investigations in which vacuum melting produced an improvement. The net fracture strength of transverse centre-notched specimens was increased by about 100%, whereas that of longitudinal specimens was increased by about 50%. The marked directionality of the air-melted sheet was almost completely removed by vacuum melting.

Cottrell[1] has investigated the effect of consumable electrode vacuum melting on the surface strain to failure in a wide bend test, using a 3% Cr-Mo-V steel in the ultra-high-strength range. He reported that vacuum melting of this steel increased the tensile strength for a given surface strain to failure by about 20 ksi.

On the other hand, reference[1] reports somewhat different results obtained with 9% Ni-4% Co steel. Vacuum re-melting reduced the directionality, but gave no increase in the nominal notch strength of longitudinal, centre-cracked, sheet specimens (0.08 and 0.180 in.) from a Si/Al deoxidized heat. Vacuum carbon deoxidation practice, however, resulted in a significant improvement for 0.180 in. sheet in the lower part of the ultra-high-strength range.

Additional evidence concerning the effect of vacuum melting on the notch strength ratio and the fracture toughness of several steels has been compiled in Table 3.

TABLE 3

Effect of Vacuum Melting on Notch Toughness

Steel	Form of Material	Type of Test-piece	Ultimate Tensile Strength, ksi	Notch Strength Ratio		K_c - ksi$\sqrt{in.}$		Reference
				Air Melt	Vacuum Melt	Air Melt	Vacuum Melt	
H-11	0.063 in. sheet	Edge-notch	250	0.65	0.70	120	135	17
			300	0.24	0.26	45	45	
300 M	"	"	250	0.40	0.50	80	95	
			300	0.26	0.31	65	78	
4340	0.180 in. sheet	Centre-cracked	280	0.31 (0.37)	0.64* (0.37)			15
300 M	0.07 in. sheet	Centre-cracked	250	0.37	0.33			18
			300	0.37	0.36			
18% Ni	0.3 in. dia.	$K_t > 15$	260	1.46	1.52			19
			290	1.06	1.46			
18% Ni	0.3 in. dia.	$K_t > 10$	270	≈ 1.39	≈ 1.45			20
4340	0.067 in. sheet	Centre-cracked	285	0.35 (0.24)		100 (60)		21
			300		0.36		108	
			270		0.73 †		220 †	
18% Ni	0.625 in. plate	$K_t > 10$	260	(1.44)	(1.44)			22
	0.5/1.0 in. plate	"	265	(1.25)	(1.43)			
	0.063 in. sheet	$K_t > 15$	275	0.92	0.96			
AMS 6434	0.063 in. sheet	Centre-cracked	250	0.89 (0.85)	0.97 (0.96)			23

Notes on Table 3

* The individual values showed considerable scatter.

† These values apply to vacuum induction melted and vacuum induction re-melted heats: the remainder apply to CEVM heats.

( ) Values in parenthesis are for transverse specimens; the remainder are essentially for longitudinal specimens.

Referring to Table 3, it will be seen that alloys H-11 and 300 M showed little improvement with vacuum melting. 4340 steel gave some improvement, particularly with the vacuum induction treatment. The 18% Ni maraging alloy and AMS 6434 also showed some improvement, this being more marked at the 290 ksi level for the maraging alloy.

It would appear that the case for vacuum melting is by no means resolved. While it can do no harm, its general effect is a reduction in directionality, with possibly some upgrading of notch toughness. Since the latter varies with the alloy and its strength level, any specific application would have to be considered on its merits, and the controlling factor may well be the economic aspect.

(b) Composition

No attempt will be made to discuss the effect of alloying elements in detail, but some comments may be of interest on the particular effects of variation between heats, carbon content, sulphur content, decarburization and purity. Campbell, Barone and Moon[24] have reported the results of notch toughness tests on two heats of 18% Ni steel (300 grade), one being a low chemistry heat and the other a high chemistry heat. In the case of bar stock, the former gave a notch strength ratio ($K_t = 12$) of 1.49, and the latter gave 1.26. In the case of 0.115 in. sheet, the former gave a K_{Ic} value of 230 ksi $\sqrt{in.}$, whereas the latter gave 119 ksi $\sqrt{in.}$ Melville[25] carried out tests on surface-cracked sheet specimens from three heats of the 300 grade material, and his results are presented in Figure 3 in terms of net strength vs. crack length. Though the chemistry was similar, the difference in fracture toughness behaviour is readily apparent. An inspection of the individual analyses revealed some correlation with nickel only, the respective contents being 18.63%, 18.43%, and 17.80%.

As regards the effect of carbon content, it is generally accepted that the toughness increases as the carbon percentage is reduced, and the lower limit of carbon is usually determined by the yield strength requirement

Cottrell, Langstone and Rendall[26] have investigated the effect of carbon content on the toughness of a 1% Cr-Mo steel of ultra-high purity. As the carbon content was increased from 0.30% through 0.37% to 0.44%, both the Charpy energy absorbed and the biaxial ductility in a wide bend decreased. Klier's[27] findings from edge-notch tensile tests on a series of 43xx (V-modified) steels were similar. Espey and his co-workers[17, 28] made an extensive study of the sharp-edge-notch characteristics of H-11 and 300 M sheet steel (0.063 in.). The notch strength ratio for the 300 M alloy in the ultra-high-strength range decreased steadily for a given tensile strength with increasing levels of carbon (0.28%, 0.34%, 0.40%, and 0.46%), in agreement with the results quoted above. The results obtained with the H-11 alloy, however, were quite the reverse, the ratio increasing with increasing carbon (0.23%, 0.26%, 0.29%, 0.39%, and 0.43%). The effect was less pronounced and tended to fade out at 0.39% C. This behaviour was confirmed by Hamaker and Vater[29] with Charpy impact tests on an H-11 type alloy, and it appears that the alloy is an exception to the general rule.

High sulphur and phosphorus, as in other steels, are detrimental to the properties of the 18% Ni maraging alloys[1]. The Charpy value for the 250 grade was reduced from 20 ft-lb at a level of 0.002% S to 10 ft-lb at the 0.014% S level. Wei[30] recently reported the results of plane-strain fracture toughness tests on a series of AISI 4345 steels containing four levels of sulphur and prepared by carefully controlled melting procedures. The K_{Ic} value (Figure 4) increased steadily as the sulphur content was reduced from 0.049% to 0.008%, at all ultra-high-strength levels. It is of interest to note that the same reference confirmed that silicon, though increasing the tempering resistance of these steels, does not yield improved fracture toughness at a given strength level.

Surface decarburization, though generally regarded as a deleterious influence, has been found to give a striking improvement in the fracture toughness behaviour of certain steels. Nevertheless, it must be remembered that decarburization can be quite harmful to the fatigue properties of ultra-high-strength steels, and its usefulness in any particular application will depend upon the extent to which the hardware is subject to cyclic or repeated loading in service. Warke and Elsea[31] have prepared a comprehensive review of the subject, to which reference should be made for detailed information. Figure 5, based on an investigation by Manning, Murphy, Nichols and Caine[32], shows the marked increase in burst strength of 12 in. diameter pressure vessels, of X-200, 300 M and MBMC-1 steels with an increase in depth of decarburization from 0.005 to 0.015 in. The increase was accompanied by a reduction in the tensile strength of about 30 ksi. The reviewers also report that Pratt and Whitney recommend decarburization for solid-propellent rocket-motor cases of H-11, D6AC and 300 M alloys.

Similar results have been reported by Cottrell and Turner[33], and Langstone[34], from burst tests on 17 in. diameter tubes and motor cases of RS 140 (3% Cr-Mo-V) steel. The strength level of the material was in the lower part of the ultra-high-strength range and the depth of decarburization varied from 0.001 in. to 0.008 in. The results were expressed in terms of the burst hoop stress/tensile strength ratio, and the consistently beneficial effect of the decarburization was evident in all tests, but more particularly for the motor cases, where biaxial ductility is important. According to Langstone, the surfaces of motor cases made by Bristol Aerojet Ltd. are partially decarburized during heat treatment.

Sheehan and Manning[35] have measured the fracture toughness of X-200 sheets, from experimental heats containing four levels of carbon, by means of the centre-notch tension test, and have studied the effect of surface decarburization. They concluded for this material that decarburization was beneficial only insofar as it decreased the yield strength, and that above 240 ksi, the fracture toughness was poor even though the sheet was decarburized. Nevertheless, it might still be better than that of the undecarburized material.

In general, therefore, it may be said that decarburization should be advantageous for such items as rocket-motor cases, but should be applied with caution to items such as landing gear until further information as to its effect on the fatigue properties is available.

(c) Thermo-Mechanical Treatment

Thermo-mechanical treatment in the present context is intended to cover those processes which result in metal reduction at low, room, or elevated temperatures, e.g. cold-rolling, marforming, ausforming. The greater part of the information available on the effects of cold-rolling was developed in connection with liquid-propellent rocket tanks or the skin for a supersonic transport, and relates to a variety of stainless steels. In general, however, the strength level at room temperature of the materials investigated is below the lower limit of the ultra-high-strength range; in several cases the notch was insufficiently sharp. Appropriate data for several stainless steels are presented in Table 4.

Notes on Table 4

1. L = longitudinal; T = transverse.
2. Figures in parenthesis refer to transverse tests.

* Aged at 750°F for 8 hr.

† Aged at 825°F for 3 hr.

‡ Aged at 700°F for 3 hr.

++ Aged at 800°F for 3 hr.

TABLE 4

Notch Toughness of Cold-Rolled Stainless Steel Sheet

Steel	Thickness, in.	Cold Reduction, %	Direction	0.2% Yield Strength, ksi	Ultimate Tensile Strength, ksi	Notch Strength Ratio	Type of Test	Reference
AM 355	0.025	15 #	L	255	256	0.86	Edge-notch	36
			T	255	265	0.64		
	0.024	20 #	L	218	239	1.01		
			T	186	233	0.95		
	0.024	30 #	L	257	262	0.98		
			T	243	281	0.86		
	0.026	35 #	L	286	289	0.96		
			T	258	286	0.80		
	0.024	40 #	L	289	294	0.91		
			T	270	297	0.71		
AISI 301 (CEVM)	0.025	50	L	215	220	1.00(.73)	Edge-notch	37
		60	"	230	235	0.87(.64)		
		70	"	245	255	0.84(.59(		
		80	"	270	275	0.57(.33)		
AISI 301	0.063	70	L	220	250	0.80	Edge-notch	38
			T	220	265	0.64		
AISI 301	0.063	70	L	218	248	0.81	Edge-notch	36
			T	220	261	0.45		
	0.031	70	L	260	>263	<0.72		
			T	-	287	0.37		
AISI 301	0.043	67	L	249	264	0.56	Centre-crack	21
			T	264	288	0.26		
		67*	L	268	280	0.65		
			T	299	315	0.22		
AM 350 (CEVM)	0.025	30 †	L	239	245	1.04	Edge-notch	39
			T	222	244	0.94		
		30 ‡	L	241	243	1.07		
			T	228	247	0.94		
		45 †	L	274	280	0.98		
			T	274	280	0.82		

It will be noted that the notch strength ratios, based on edge-notch tests, are remarkably high, even at the 280 ksi level. Alloys AM 350 and 355 appear to be closely comparable and somewhat superior to AISI 301, in which directionality is more pronounced as shown by the transverse notched tests. More recently, test results were reported by Alper[40] for cryogenically stretch-formed AISI 301. 14 in. diameter spherical pressure vessels, stretched at -320°F with or without ageing, gave an increase in burst strength at -320°F of more than 25%. The technique obviously shows promise for the lightweight cryogenic pressure vessel field.

The effect of cold rolling has also been studied on 18% Ni maraging steel, although here it is called marforming and is carried out between the annealing and ageing treatments. The results, however, are contradictory. Decker, Eash and Goldman[19] made tests on small experimental heats with 50% marforming, and reported an increase in yield strength, tensile strength and notch strength with a slight increase in the notch strength ratio. The figures they reported for the K_c value of 50% marformed sheet, 0.039 in. to 0.079 in., were also relatively high, ranging from 170 to >244 ksi $\sqrt{in.}$ Data given in reference 22, on the other hand, indicate a steady reduction in the K_c value for 0.115 in. sheet as the degree of marforming increases from 0 to 50%, for both the 250 and 300 grade.

Reference 22 also reports the effect of 50% hot-working in the austenitic range, followed by quenching, on the notch strength (centre-crack specimens) of 9% Ni-4% Co steel sheet. When tempered at 400°F, the yield and tensile strengths in both directions and the longitudinal notch strength increased, but the transverse notch strength decreased. Matas, Hill and Munger[14] report a similar effect, though no details are given. After the above treatment, tensile strengths as high as 370 ksi were obtained with a K_c value of 150 ksi $\sqrt{in.}$ The transverse fracture toughness was stated to be only 70-80% of the longitudinal value, but it would appear that the process should have some specialized applications. Kula and

Dhosi[41] applied the treatment to SAE 4340 steel plate, tempered at 450°F, and found a corresponding improvement. Reductions up to 50% at 1550°F had no effect on the tensile strength, but tended to raise the Charpy energy absorbed vs. temperature curve and translate it in the direction of lower temperatures. Unfortunately, no data on the notched or cracked tensile strength are available for comparison.

The effect of mar-straining, in which the quenched and tempered steel is plastically deformed and subsequently re-tempered or aged, was investigated by R. E. Yount[42] with regard to its possible use for solid-propellent rocket-motor cases. Centre-notch tension tests were made on two alloys, D6AC and modified S-5 (0.5 C, 1.8 Si, 0.5 Mo, 0.25 V) in the form of sheet. In all cases, the specimen blanks were pre-strained up to 1.0% and aged before notching. Results for both steels showed that 0.2% mar-strain lowered the K_c value to about 90 ksi $\sqrt{\text{in.}}$, but that this value remained substantially unchanged up to 1.0% mar-strain. Yount states that this value is still higher than that of H-11 steel, which has been successfully used in pressure vessels. Furthermore, the pre-straining process would be expected to reduce the effects of sub-critical defects already present in the material, by the addition of compressive stresses and/or notch blunting. Tests were also made on 6 in. diameter cylinders from ring forgings, pre-strained by pressurizing, and then aged. The burst strengths for both alloys were equivalent to the tensile strength of the mar-strained material, the highest value obtained being 362 ksi.

Similar results were reported by Steigerwald[43] from edge-notch and centre-crack tests on H-11 sheet (UTS 290 ksi) after warm pre-stressing. In this treatment, the specimen blanks were pre-stressed at various levels at 80°F or 600°F before notching and testing at room temperature. The results showed, as above, a general decrease in the notch tensile strength, independent of the pre-stressing temperature and more marked as the pre-stress level was raised. Similar tests, however, made on specimens pre-stressed after notching showed a beneficial effect which tended to increase with the pre-stress level. The increase appeared to be limited only by the notch strength at the pre-stressing temperature. A much smaller improvement was observed in tests on another steel, 300 M alloy, presumably due to its low notch strength at the pre-stressing temperature (550°F). The investigator pointed out that the treatment was most effective when applied to materials of fairly high notch sensitivity; H-11 steel at a lower strength level was less improved. Its principal application would appear to be to hardware containing local potential trouble-spots, such as welds with microcracks.

Before leaving this section, some reference should be made to the effect of banding which may be found in rolled products and has been particularly prevalent in the 18% Ni steel. Fracture toughness tests on the 250 grade have generally given higher values (about 30%) from surface-crack specimens than from edge or centre-crack specimens. To investigate this effect further, Pellissier[44] carried out single-edge-crack tests on 0.14 in. thick specimens cut from 1-1/8 in. plate in four principal orientations. The orientations, with respect to the rolling direction, and the average G_{Ic} values obtained are shown in Figure 6. The G_{Ic} values in the longitudinal$^{(A)}$ and transverse$^{(B)}$ directions are closely similar, but the G_{Ic} value for the C orientation, simulating the surface-crack specimen, is about 25% greater. The much lower toughness observed in the D orientation demonstrates the harmful effect of banding in those conditions under which a crack can develop in the plane of the bands. This deleterious effect must obviously be taken into account in any application involving material which is known to be subject to banding. Pellissier also reported that the banding could be essentially eliminated by annealing at 2300°F for 16 hours prior to heat treatment. Unfortunately, homogenization reduced the longitudinal notch-tensile strength of round specimens from 319 ksi to 253 ksi.

4. EFFECT OF ENVIRONMENT

Fracture mechanics concepts provide the designer with a relationship between defect or crack size and the stress at the moment of catastrophic failure. The slow growth of the original defect to its critical value is obviously a matter of some importance, and several investigations have been carried out to shed some light on the effect of the environment on this problem. The time-dependent delayed failure of rocket motor cases at constant pressure when exposed to aqueous environments led Steigerwald[45] to examine the effect of such environments on centre-crack specimens of 300 M and H-11 sheet (UTS 290 ksi). At about 85% of the notch tensile strength, failure did not occur in 100 hours with 300 M steel and no liquid environment. The same steel in the presence of aqueous solutions of different pH values (4.8 to 9.0) gave failures in a matter of minutes; non-aqueous solvents and lubricating oil extended the time required considerably. H-11 tool steel gave similar results in distilled water. One other observation of interest was that the K_c value remained constant over the delayed failure range.

Saperstein and Whiteson[46] and Bennett[47] reported the results of fracture toughness tests on cracked sheet specimens of 4340 steel in distilled water. Delayed failure was again observed, due to slow crack growth, within 30 minutes at stresses as low as 40% of the tensile strength. The latter author, however, reported satisfactory behaviour in oil saturated with water. Saperstein and Whiteson made comparison tests on 18% Ni maraging steel and demonstrated its clear superiority, on the basis of 30 minutes exposure to stresses over 90% of the net fracture stress in air. Similar results for 4340 steel sheet (UTS 265 ksi) in distilled water have been presented by Yen and Pendleberry[48], who showed that the gross strength was proportional to the logarithm of the holding time for a given initial shallow-crack length.

Tiffany and Masters[49] recently reported the results of sustained-load tests on welded shallow-crack specimens of 18% Ni steel plate in a water environment. Little effect was observed with the base metal up to 100 hours, but the initial-to-critical stress intensity ratio decreased rapidly with holding time in the weld metal. Tests on notched bar specimens of 17-7 PH also showed only a small effect of a wet environment on slow crack growth, but data for surface-crack specimens of 4330M steel showed a significant effect and indicated a threshold stress intensity level of about 30% of the critical value.

Slow crack growth is also effected by cyclic loading, and the combined effect of water vapour plus repeated loading on the fracture toughness of 4340 steel (UTS 260 ksi) has been studied by Van der Sluys[50]. Tests were made on pre-cracked round specimens in an argon atmosphere containing various amounts of water vapour. The data indicated that, though the presence of moisture only reduced G_{Ic} slightly, there was an increase in slow crack growth with increasing humidity. The addition of cyclic loading produced a further increase. When the results were compared on the basis of the stress required to cause either slow crack growth or failure in less than 100 cycles, it was found that a condition of 100% relative humidity reduced the "dry" stress level by nearly 50%.

Additional confirmation of the effect of moisture is provided by the work of Tiffany and Lorenz[51] on D6AC steel plate. The endurance of pre-cracked round specimens under cyclic loading was reduced by a factor of more than ten in an atmosphere of high humidity. As above, they too reported no significant effect of moisture on fracture toughness.

Although the data are still relatively sparse, it is evident that the environment, in particular water or water vapour, can have a profound effect on slow crack growth, more so with some alloys than with others. The designer must therefore pay due attention to this factor, whether his hardware is subject to sustained or to cyclic loading conditions.

5. EFFECT OF LOADING RATE

Even less experimental information is available on the effect of loading rate on the toughness of ultra-high-strength steels, which are not generally regarded as being strain-rate sensitive. Since the hardware may be exposed to high loading rates due to shock or impact, this is a matter of considerable importance and merits further investigation. Srawley and Beachem[52] carried out centre-crack tests on a martensitic stainless steel, 422 M (UTS 250 ksi), over a range of temperatures. With rapid loading, from 500 to 1000 X normal rate, the net-fracture-stress transition temperature (NFSTT) was hardly affected. Below the NFSTT (e.g., at room temperature), however, the net fracture stress was lowered slightly, while above the NFSTT, it was raised (Figure 7). Additional work by the same investigators, reported by Marschall[53], showed a greater decrease (about 43%) in the net fracture stress at room temperature for a higher-strength steel (UTS 290 ksi). The rapid loading rate was from 200 to 300 X normal rate; other details were not available.

A few test results were presented by Raring et al[39] for AM 355 stainless steel sheet (UTS 230/240 ksi). The specimens were large, centre-notch panels, 24 in. wide with an 8 in. notch, and the loading time varied from 0.2 to 200 sec. For tests lasting between 2 and 200 sec, the net fracture stress and G_c showed little effect of loading rate, but at the higher rates both values decreased, about 25% and 40% respectively.

Some tests reported by Yen and Pendleberry[48] on shallow-crack specimens of 4340 steel sheet (UTS 265 ksi) indicated a similar effect. Two series of tests were carried out with three crack lengths at stress rates of 100 and 12 ksi/min. Tests at the slower rate gave consistently higher values of the fracture strength, though the effect was slight (maximum, 3%).

Evidence for an improvement in notch toughness as a result of rapid loading was given in reference 22, in which the results of G_c determinations were presented for 4340, D6AC and 18% Ni (250 and 300 grade) alloys. The tests were made at strain rates of 0.00005, 0.05 and 0.15 in./in./sec. At the two lower rates, the G_c values for each steel were closely comparable. At the highest rate, the 4340 and D6A C values showed no change, whereas the 18% Ni values showed a marked increase.

In the present state of the art, it would appear that materials for hardware for which high strain rates are a service condition or a potential hazard should preferably be evaluated under comparable loading rates. Failing this, some additional factor of safety might well be incorporated.

6. TESTS ON TYPICAL HARDWARE

Having reviewed the available notch toughness data and discussed the effects of some of the more important parameters, it is desirable to pay some attention to the results of tests on actual or sub-scale hardware before considering the question of design requirements. The great majority of the data in the literature relate to internal pressure tests on tubes, cylinders or spheres, and a summary of the results for ultra-high-strength steels is presented in Table 5.

TABLE 5

Summary of Results of Burst Tests on Pressure Vessels

Steel	Form of Specimen	Thickness, in.	0.2% Yield Strength, ksi	Ultimate Tensile Strength, ksi	K_c ksi $\sqrt{in.}$	Hoop Burst Stress, ksi	Depth of Decarburization, in.	Type of Fracture	Ref.	Remarks
AISI 4130	12 in. dia cylinder	.08	186	226	> 290	247(3)	0.019	Full Shear	32	Longitudinal weld: NSR > .88
	$3\frac{1}{2}$ in. dia cylinder	.08	207	250	60.6*	258	-	" "	54	Draw and Spin manufacture: pre-cracked vessel gave burst stress of 208 ksi ($\frac{1}{4}$ in. crack) K_{Ic} after test = 56.0 ksi $\sqrt{in.}$
MBMC No. 1	$11\frac{1}{2}$ in. dia cylinder	.10	216	253	-	281(4)	0.011	-	55	Drawn vessel
	16 in. dia cylinder	.06	237	280	-	318(32)	0.011	-	"	Forged and spun vessel
	" "	.05	240	271	-	289(23)	0.004	-	"	Rolled and welded vessel
	12 in. dia cylinder	.08	204	243	> 209	261(3)	0.019	Mainly shear	31	" " " " : NSR = .70
	" "	.08	234	276	80	167(2)	0.008	Mixed		" " " " : NSR = .43
Mod. S-5	6 in. dia cylinder	.06	≈ 300	-	≈ 87	354	-	Full shear	42	Non-welded manufacture: .26% pre-strain
	" "	.06	≈ 310	-	-	360	-	" "	"	" " " : .30% " "
	" "	.06	≈ 310	-	-	362	-	-	"	" " " : .30% " " and pre-cycled 9 times.
4137 Co	6 in. dia cylinder	.05	238	273	-	268(2)	-	-	56	Non-welded manufacture
	" "	.05	235	270	-	250	-	-	"	Girth weld
	12 in. dia cylinder	.10	240	275	-	286	-	-	"	Rolled and welded vessel
	54 in. dia vessel	.08	240	275	-	280	-	-	"	Deep drawn, longitudinal weld: pre-cycled 3 times.
D6AC	6 in. dia cylinder	.08	≈ 280	-	≈ 92	313(2)	-	Full shear	42	Non-welded manufacture: .34% pre-strain and pre-cycled 9 times.
	$9\frac{1}{2}$ in. dia cylinder	.04	243	277	-	326(2)	0.005	-	55	Girth weld
	10 in. dia cylinder	-	220	270	-	322(2)	-	-	"	Forge-extruded
	" "	-	218	235	-	257(2)	-	-	"	" "
	24 in. dia cylinder	-	229	264	-	235(2)	-	-	"	" " , and ring-rolled.
	" "	.08	190	215	-	256(4)	-	-	57	Forged and girth welded: Allison parameter 150 ksi.
	" "	.08	245	286	-	157	-	-	"	" " " " " " 45 ksi
	40 in. dia cylinder	.08	199	219	150†	Service test satisfactory	-	-	58	Girth weld: K_{Ic} = 66 ksi $\sqrt{in.}$
	$3\frac{1}{2}$ in. dia cylinder	.08	234	264	40.5*	> 168	-	Flat	54	Draw and spin manufacture: K_{Ic} after test = 28.9 ksi $\sqrt{in.}$ Pre-cracked cylinder gave burst stress of 102 ksi ($\frac{1}{4}$ in. crack).
	17 in. dia cylinder	.25	247	280	50*	169	-	-	51	Roll and weld manufacture: pre-cracked ($\frac{1}{4}$ in.)
	" "	"	"	"	"	73	-	-	"	" " " " " " ($\frac{1}{2}$ in.)

TABLE 5 (Cont'd.)

Summary of Results of Burst Tests on Pressure Vessels

Steel	Form of Specimen	Thickness, in.	0.2% Yield Strength, ksi	Ultimate Tensile Strength, ksi	K_c ksi$\sqrt{in.}$	Hoop Burst Stress, ksi	Depth of Decarburization, in.	Type of Fracture	Ref.	Remarks
3Cr-Mo-V	17 in. dia tube	.07	208	260	-	245(4)	(None)	-	33	Helical welding
	" "	.07	199	248	-	259(6)	0.005	-	"	" "
	" "	.07	200	244	-	212(2)	(None)	-	"	" "
	" "	.07	191	231	-	226(4)	0.004	-	"	" "
	17 in. dia cylinder	.07	190	230	-	178	(None)	-	"	" "
	" "	.07	189	231	-	235(3)	0.003	-	"	" "
	" "	.07	185	238	-	159	(None)	Brittle	"	" " : Allison parameter = 0 ksi
	" "	.07	183	231	-	236	0.004	Full shear	"	" " : " " = 33.6 ksi
AMS 6434	$3\frac{1}{2}$ in. dia cylinder	.07	225	270	58.5*	295	-	Full shear	54	Draw and spin manufacture: K_{Ic} after test = 44.4 ksi$\sqrt{in.}$ Pre-cracked cylinder gave burst stress of 167 ksi ($\frac{1}{4}$ in. crack).
AISI 4340	$3\frac{1}{2}$ in. dia cylinder	.07	214	254	62.0*	290	-	Full shear	54	Draw and spin manufacture: K_{Ic} after test = 49.7 ksi$\sqrt{in.}$ Pre-cracked cylinder gave burst stress of 195 ksi ($\frac{1}{4}$ in. crack).
X-200	$3\frac{1}{2}$ in. dia cylinder	.08	246	286	31.6*	>205	-	Mixed	54	Draw and spin manufacture: K_{Ic} after test = 26.1 ksi$\sqrt{in.}$ Pre-cracked cylinder gave burst stress of about 105 ksi ($\frac{1}{4}$ in. crack).
	12 in. dia cylinder	.07	215	259	>197	278	0.018	Mainly shear	31	Roll and weld manufacture: NSR = .60
	" "	.07	215	259	>171	246(2)	0.024	" "	"	" " " " : NSR = .71
	" "	.07	251	291	102	100(2)	0.006	-		" " " " : NSR = .37
Rocoloy 270	$3\frac{1}{2}$ in. dia cylinder	.08	256	313	30.1*	204	--	Mixed	54	Draw and spin manufacture: K_{Ic} after test = 30.7 ksi$\sqrt{in.}$ Pre-cracked cylinder gave burst stress of 113 ksi ($\frac{1}{4}$ in. crack).
AISI 4330V (Mod. + Si)	3 in. dia cylinder	-	206	248	115*	171(2)	-	-	58	Ring forgings with external notch (.15 in.) pre-cycled to leak.
	3 in. dia tube	-	206	248	115*	165(2)	-	-	"	

TABLE 5 (Cont'd.)

Summary of Results of Burst Tests on Pressure Vessels

Steel	Form of Specimen	Thickness, in.	0.2% Yield Strength, ksi	Ultimate Tensile Strength, ksi	K_c ksi$\sqrt{in.}$	Hoop Burst Stress, ksi	Depth of Decarburization, in.	Type of Fracture	Ref.	Remarks
300-M	9 in. dia cylinder	.06	244	290	-	329	0.007	-	55	Girth weld
	12 in. dia cylinder	.08	216	261	>177	272(3)	0.016	Mainly shear	31	Roll and weld manufacture: NSR = .63
	" "	.08	237	280	144	170(2)	0.010	-	"	" " " " : NSR = .53
	40 in. dia cylinder	.08	232	282	150†	232	-	Full shear	58	Hot-cupped and cold drawn; pre-cycled: K_{Ic} = 86 ksi$\sqrt{in.}$
	$3\frac{1}{2}$ in. dia cylinder	.08	242	286	53.3*	324	-	Full shear	54	Draw and spin manufacture: K_{Ic} after test = 43.2 ksi$\sqrt{in.}$ Pre-cracked cylinder gave estimated burst stress of 200 ksi ($\frac{1}{4}$ in. crack).
H-11	$11\frac{1}{2}$ in. dia cylinder	.10	211	255	-	279(4)	0.008	-	55	Drawn
	$9\frac{1}{2}$ in. dia cylinder	.07	241	291	-	354	0.0005	-	"	Flow-turned
	6 in. dia cylinder	.05	226	255	-	284(78)	-	-	"	Roll and weld manufacture
	" "	.05	≈ 230	286	-	312(3)	-	-	"	Drawn
	" "	.05	≈ 220	263	-	308(3)	-	-	"	"
	" "	.05	≈ 210	243	-	287(3)	-	-	"	"
	12 in. dia cylinder	.08	226	256	80	219(3)	0.009	5% shear	32	Roll and weld manufacture: NSR = .41
	" "	.08	244	297	97	233(2)	0.007	-	"	" " " " : NSR = .42
	$3\frac{1}{2}$ in. dia cylinder	.09	265	303	34.6*	235	-	Flat (fragmented)	54	Draw and spin manufacture: K_{Ic} after test = 31.5 ksi$\sqrt{in.}$ Pre-cracked cylinder gave estimated burst stress of 185 ksi ($\frac{1}{4}$ in. crack).
301	$13\frac{1}{2}$ in. dia sphere	.03	≈ 195	≈ 220	-	207	-	-	40	Hydroformed and welded: pre-strain at -32°F, .06
	" "	.03	≈ 275	≈ 280	-	258**	-	-	"	" " " : " " -32°F = .06
	($14\frac{1}{2}$ in. dia after stretching)	.08	≈ 275	≈ 280	-	287**	-	-	"	" " " : " " " = .075
		.08	≈ 260	≈ 270	-	248**	-	-	"	" " " : " " " = .06
					**Aged after pre-straining					
18% Ni	6 in. dia cylinder	-	-	≈ 290	-	324(2)	-	-	55	Forged and welded
	" "	.14	310	315	-	338(3)	-	Full shear (fragmented)	25	Draw and spin manufacture
	24 in. dia cylinder	.13	288	291	-	328(2)	-	" "	59	Roll-formed and welded; pre-cycled.
	$3\frac{1}{2}$ in. dia cylinder	.09	257	267	59.1*	268	-	Full shear	54	Draw and spin manufacture: K_{Ic} after test = 59.2 ksi$\sqrt{in.}$ Pre-cracked cylinder gave estimated burst stress of 250 ksi ($\frac{1}{4}$ in. crack).
	" "	.09	307	310	63.8*	322	-	" "	"	Draw and spin manufacture: K_{Ic} after test = 55.8 ksi$\sqrt{in.}$ Pre-cracked cylinder gave estimated burst stress of 300 ksi ($\frac{1}{4}$ in. crack).

Notes on Table 5

1. K values:

   † indicates K_c value obtained from a centre-crack tension test.

   * indicates K_{Ic} value obtained from a centre-crack tension test.

   The remaining data represent K_c values obtained from centre-notch tension tests.

2. Figures in parenthesis indicate the number of specimens tested when greater than one.

3. The chemical compositions of the following alloys were not included in Table 1:

Mod. S-5:	0.50 C, 0.80 Mn, 1.8 Si, 0.50 Mo, 0.25 V.
3Cr-Mo-V:	0.40 C, 0. 60 Mn, 0.25 Si, 3 Cr, 1 Mo, 0.20 V.
Rocoloy 270:	0.45 C, 0.55 Mn, 1.2 Si, 1.7 Cr, 1.3 Ni, 0.5 Mo, 0.20 V, 0.30 W.

Considering first the hoop burst stress data, it is apparent that values in excess of 300 ksi can be obtained with a number of ultra-high-strength alloys (MBMC No. 1, Mod. S-5, D6AC, 300 M, H-11 and 18% Ni) in vessels ranging from $3\frac{1}{2}$ in. to 24 in. diameter. The highest values reported are 354 ksi for a $9\frac{1}{2}$ in. diameter vessel of H-11 steel with slight decarburization, and 362 ksi for a 6 in. diameter vessel of Mod. S-5 steel, pre-strained 0.30% and pre-cycled. At the other end of the scale, the lowest value (100 ksi) was obtained with a welded 12 in. diameter vessel of X-200 steel. Unfortunately, the literature is not always too clear regarding the method of manufacture, but at least three alloys, 300 M, D6AC and 18% Ni, gave burst stresses exceeding 300 ksi when welding was involved.

The majority of the results are shown plotted in Figure 8 in terms of hoop burst stress/yield strength ratio versus yield strength. The plot is very similar to that presented by Manning et al[32], and confirms their conclusion that satisfactory performance, burst stress exceeding yield strength, can be expected for pressure vessels fabricated from steels with yield strengths up to 220 ksi. The only exceptions to this conclusions are two results for the 3Cr-Mo-V alloy. Above a yield strength of 220 ksi, the scatter becomes most pronounced, and the only alloy giving consistent satisfactory performance is the 18% Ni maraging steel. It is in this range that methods, previously discussed, for improving the notch toughness might most efficiently be employed. For example, the pre-straining of vessels of Mod. S-5 and D6AC steels in all cases gave burst stress/yield strength ratios greater than one at yield strengths well above 220 ksi. Similar experiments with 301 vessels were not quite as effective.

It is interesting to note that the beneficial effects reported for a small amount of surface decarburization on notch tensile specimens are confirmed by burst tests on actual vessels. It has been suggested that decarburization is beneficial only insofar as it reduces the yield strength, and there is considerable evidence to support this argument. The results of the burst tests on 3Cr-Mo-V steel vessels[33], however, are directly contradictory. The small amount of decarburization (0.003-0.004 in.) had little effect on the yield strength, admittedly somewhat low, but a significant effect on the burst stress of 17 in. diameter cylinders.

Table 5 also gives results obtained from burst tests on pre-cracked vessels of a number of alloys. These tests were essentially carried out in one laboratory[54] on $3\frac{1}{2}$ in. diameter cylinders of draw-and-spin manufacture, and were supplemented by K_{Ic} determinations from centre-cracked tensile tests. The results for both uncracked and cracked cylinders are shown plotted in Figure 9 in terms of burst hoop stress/yield strength versus K_{Ic} value. The data for the uncracked vessels suggest that a K_{Ic} value of 45-50 ksi $\sqrt{in.}$ is required to give a burst stress exceeding the uniaxial yield strength. In the presence of a 1/4 in. crack, which presumably would not escape detection but might develop during service, a K_{Ic} value of about 60 ksi$\sqrt{in.}$ would probably give a burst stress within 10% of the yield strength.

Additional notch toughness data included in Table 5 are values of the Allison parameter, obtained from instrumented bend tests; of K_c, obtained from centre-notch tensile tests; and of the notch strength ratio, obtained from edge-notch tensile tests. Data for the former are rather meagre and are available only for vessels of D6AC and 3 Cr-Mo-V alloys. In both cases, a lower value of the parameter is associated with a burst stress below the yield strength, and a higher value with a burst stress above the yield strength. Apart from showing the discriminatory nature of the test, the results do not warrant any further conclusions.

The K_c and NSR data were obtained from a series of tests on specimens taken from burst vessels of AISI 4130, MBMC No. 1, X-200, 300M and H-11 steels, carried out in one laboratory[32]. The results are shown plotted in terms of hoop burst stress/yield strength ratio versus K_c and NSR in Figures 10 and 11, respectively, taken from the author's publication. Manning et al suggested that minimum values for K_c of 150 ksi $\sqrt{\text{in.}}$, and NSR of about 0.57, are required for satisfactory performance, but pointed out that their data were derived from specimens which had undergone a certain amount of plastic deformation.

Apart from the information available and discussed above on pressure vessels, a few results have been reported for recoilless rifles[60] and aircraft landing gear[61]. In the case of the recoilless rifles, the requirement was for a steel with a minimum yield strength of 220 ksi and good notch toughness. The alloy selected was 4330 V (Mod. + Si) steel, and at the specified yield strength, the NSR value was about 0.9 and the K_{Ic} value about 85 ksi $\sqrt{\text{in.}}$ The data were obtained from notched round specimens (no details of notch given). Centre-notch sheet specimens were used to determine K_c, the value obtained being 290 ksi $\sqrt{\text{in.}}$ Hydrostatic tests were carried out on $3\frac{1}{2}$ in. diameter cylinders, and gave full shear failures with the combined yield stress very close to the uniaxial yield strength. Firing tests on actual rifles were also satisfactory. It will be apparent that these results are in agreement with the conclusions arrived at earlier in this section.

The information regarding landing gear is contained in an article analyzing the service failure of three aircraft parts made of AISI 4340 steel, quenched and tempered to a tensile strength of about 270 ksi with a yield strength of about 235 ksi. Although no data pertinent to the present discussion were given, the investigation was of some importance since it stressed the dangers of hydrogen embrittlement in ultra-high-strength steel. All three failures were attributed basically to this cause, and the necessity of extreme care in processing and fabrication and in the design and maintenance of the hardware was emphasized. It is not proposed to review this particular aspect of the notch toughness problem here, but reference may be made to a related study of solid-fuel rocket chambers carried out by Shank et al[62]. The authors stressed the beneficial effects of surface decarburization, and concluded that it was necessary with present alloys to design to notch strength ratios of less than 1.0 to secure minimum feasible weight.

7. DESIGN REQUIREMENTS

Examination of structures and structural components of ultra-high-strength steel which have failed in service generally reveals that the origin of failure was a small crack or crack-like flaw. Presumably the initial flaw size was insufficient to cause fracture in the proof test or upon initial loading, and required a number of load cycles and/or time under sustained load to attain the critical size for failure. The flaws normally encountered can be classified as surface flaws, embedded flaws, and through-the-thickness cracks[10,49]. For surface and embedded flaws, the conditions are generally those of plane strain. The initial flaws may or may not reach the critical size before growing through the thickness, depending on the K_{Ic} value, the applied stress level and the material thickness. For through-the-thickness cracks in relatively thin material, plane stress conditions normally predominate and K_c is the important parameter. As the thickness increases, the fracture appearance changes from full shear to flat, and the K_{Ic} value should be used.

It will be apparent from the foregoing that three of the principal factors controlling the performance of ultra-high-strength hardware are the initial flaw size, flaw growth, and critical flaw size. These factors in turn are dependent upon a number of others, notably the material, its processing and heat treatment, the fabrication procedure, the service temperature, the type of loading, the environment, and the accuracy and extent of the non-destructive inspection. The effect of the majority of the factors on the notch toughness of various materials has already been discussed, but some additional comments are desirable at this point.

Firstly, with regard to the inherent toughness of the material, a number of investigators[25, 63] have reported significant variations from heat to heat and from vendor to vendor with no obvious connection with the chemistry. Such a situation is to be regretted, but also accepted, in the present state of the art, and is an additional reason for the incorporation of some form of fracture toughness test, other than the standard Charpy test which is insufficiently sensitive, into material specifications.

Secondly, the effect of temperature has so far been largely ignored. The operating temperature is undoubtedly an important factor, and must be taken into consideration for such applications as cryogenic tankage and those involving aerodynamic heating. It is not intended in the present review, however, to deal with this aspect in detail, but references are provided for those who are specifically interested. In general, the notch toughness parameters (NSR, K_c, and K_{Ic}) of the ultra-high-strength steels decrease as the test temperature is lowered from room temperature to cryogenic temperatures. References 60, 64, 65, 66 and 67 present data for 4330 V (Mod. + Si), Vascojet 1000 and 300M, 4340, X200, and AMS 6434 alloys, respectively. In contrast, limited tests on both grades of 18% Ni alloy[24] indicate little effect, if anything a slight increase, down to -45°F. The information available on the effect of elevated temperatures (up to 400°F) on the toughness is less consistent. References 16, 24 and 68 indicate a decrease for AMS 6434, 18% Ni, and 4340 alloys respectively, when the temperature is raised, whereas references 64 and 66 show an increase for Vascojet 1000 and 300M and X200 alloys, respectively.

Thirdly, there is the question of the degree and extent of the non-destructive inspection. The adequacy of the inspection procedure, since it controls the magnitude of the initial flaw size, is a major factor in determining the performance of the hardware. The difficulties involved in the adequate inspection of certain items such as large rocket-motor cases, however, must be appreciated. It has been reported that the smallest flaw known to cause a failure was 1/32 in. long and 1/32 in. deep[69]. An additional control, in the case of pressure vessels, is supplied by the conventional proof test which, if successful, actually defines the maximum possible initial flaw size that exists in the vessel.

The three primary factors controlling service performance, as mentioned earlier, are initial flaw size, flaw growth, and critical flaw size. With the aid of fracture mechanics and a knowledge of the fracture toughness, the critical flaw size can be derived for the operating conditions. In order, then, to determine the limiting initial flaw size, and hence the level of inspection, data must be obtained on the rate of flaw growth, whether this be due to cyclic or sustained loading, or both. Tiffany et al[49, 51] have utilized the reverse approach to predict with reasonable accuracy the life of a 17 in. diameter cylinder of D6AC, subjected to low cycle fatigue. The cyclic flaw growth was determined, using notched bars and surface-cracked specimens and cycling them to failure at various percentages of the critical stress intensity. The curve obtained was then used to predict the life of pre-flawed cylinders. Data were also presented on the cyclic flaw growth of 17-7 PH and 18% Ni steels and the flaw growth under sustained loading, by a similar procedure, of 17-7 PH, 4330M and 18% Ni steels. Under sustained loading, crack growth does not normally occur when the stress intensity is less than about 80% of the critical value, but both types of growth may have to be taken into account in any given application. An exception to this behaviour has been noted for certain alloys in the presence of water, when crack growth may take place at considerably lower stress intensities.

One test on a surface-cracked specimen of 18% Ni steel is of particular interest, since it supports Irwin's so-called "second line defence" suggestion[70]. The plane stress fracture toughness of an ultra-high-strength steel is much higher than the plane strain value, which governs the onset of instability, and may be sufficient to arrest the running crack. In the test on a 1/4 in. thick specimen of 18% Ni steel, the surface crack was observed to pop through the thickness at 173 ksi and was arrested until the stress was raised to 178 ksi, at which level the specimen fractured.

It will be apparent that this aspect of the problem, cyclic flaw growth, has overlapped into the field of fatigue, and some consideration may usefully be given to a tentative approach from this field[71]. The method, proposed by Kuhn, is designed for the prediction of the effect of flaws on both the static and the fatigue strength. Briefly, the static strength of a cracked component is predicted from the theoretical stress concentration factor corrected for size effect by the Neuber constant and for the effect of plasticity. For fatigue loading near the fatigue limit, the plasticity correction is omitted. The method appears to have been applied with reasonable accuracy to the prediction of the fracture stress of cracked aluminum and titanium alloy sheet, and of the notch fatigue factor of low-alloy steel shafts, but no corresponding applications have been presented for ultra-high-strength steels. Some data are given for H-11 steel in which the method is used satisfactorily to predict the notch strength ratio for a limited range of root radii on the basis of a value of Neuber's constant derived from tests of cracked specimens. Further examination of this approach is warranted and might be quite rewarding.

Numerous other design procedures and design criteria, both theoretical and experimental, have been proposed during the past few years. One of the earlier criteria was suggested by Srawley[72], and involved the fracture appearance transition temperature (FATT) and the net fracture stress transition temperature (NFSTT). The former was defined as the lowest temperature at which a centre-crack specimen would exhibit full shear, and the latter as the temperature at which the net fracture stress is equal to the yield strength. If, for the particular thickness and condition of the material, either the FATT or the NFSTT were above the lowest operating temperature, the author contended, the steel should not be used. Of the alloys studied at that time, only two gave an FATT below room temperature (75°F) with a yield strength above 200 ksi, as can be seen in Figure 12. Unfortunately, no service data were available to check the operational validity of this approach.

A design criterion, based on fracture mechanics and associated with the foregoing, was proposed by Irwin[3] for pressure vessels, and is known as the leak-before-burst criterion. It is based upon the parameter β_c, which is proportional to the ratio between the plastic zone size ahead of the crack tip and the plate thickness (B).

$$\beta_c = \frac{K_c^2}{B\,\sigma_y^2} \qquad (\sigma_y = 0.2\%\ \text{yield strength})$$

Irwin suggested that β_c should be equal to or greater than 2π, which is equivalent to stating that the material should be able to arrest a through-crack of length equal to 2B when the stress equals the yield strength. This relationship corresponds experimentally[73] to test-pieces containing more than 80% shear on the fracture surface, hence the connection with Srawley's criterion. Experimental justification is afforded by burst tests on pressure vessels carried out by Carman et al[58]. From the results for several alloys, the authors concluded that a value of 2π for β_c was, at least, desirable, and that extreme care in fabrication and inspection was necessary below this value. They also pointed out that under certain conditions, K_c/K_{Ic} ratio of less than two, the fracture behaviour is governed by the plane strain fracture toughness. Additional verification is provided by the data for pressure vessels summarized in Table 5. For all those tests for which the necessary information is available, the behaviour of the pressure vessel could essentially be predicted by the β_c criterion. Furthermore, the condition proposed by Manning et al[32] of a minimum K_c value of 150 ksi$\sqrt{in.}$ at yield strengths up to 220 ksi for vessels of about 0.08 in. wall thickness corresponds to a minimum β_c value of about 2π [4].

Other design criteria have been suggested which are not based directly on fracture mechanics; in particular, a critical value of the notch-strength ratio (NSR). A value of unity has been proposed and, as a less conservative criterion, a notch tensile strength exceeding the yield strength[12]. Unfortunately, the notch strength is dependent upon the specimen size, notch depth and root radius, and much of the available data does not meet the current requirements for a sharp notch. Experimental data from pressure vessel tests by Manning et al[32] led to the conclusion that a NSR value of about 0.57 was sufficient for satisfactory performance, as mentioned earlier. Very little additional information is available in Table 5 to check this conclusion. Klier[65] has carried out tests on fatigue-cracked, round specimens of 4340 steel and has concluded that the temperature at which NSR = 1 corresponds to the nil ductility transition (NDT) temperature as measured in the drop-weight test. This point will be returned to later in the discussion.

Another possible criterion, which has been shown to correlate well with the behaviour of pressure vessels, is based on the instrumented bend test. In this test, developed by Hanink and Sippel[74], the specimen dimensions are designed to produce a biaxial stress field similar to that in a cylindrical pressure vessel, and no artificial notch is required. The bend test parameter is the difference between the maximum fibre stress and the stress at which rapid crack propagation begins, and has been shown to correlate with the per cent shear and the G_c value for centre-crack tests. The results of tests by Cottrell and Turner[33] would suggest a critical parameter value of more than 30 ksi, since their bend tests were made on material from burst vessels. The only other data available[57] indicate a value in excess of 45 ksi. The important point is that this simple test does reproduce pressure vessel conditions, and can discriminate between satisfactory and unsatisfactory material. Cottrell et al[26] have also used a wide bend test to evaluate the biaxial ductility of low-alloy steel for rocket-case construction. If the specimen can be bent through 180° without failure around a former of radius equal to four times the sheet thickness, the ductility is considered to be satisfactory.

The consideration of design criteria would not be complete without a reference to the extensive work of Pellini and his collaborators. In order to study the response of thicker material, suitable for submarine hulls and hydrofoils, to a crack, these investigators developed the drop-weight test and the explosion bulge test, and more recently the tear versions of these tests. The drop-weight test was designed to evaluate the resistance of the material to crack propagation in a stress field of yield strength level, and leads to the determination of the NDT. The drop-weight tear test differs essentially in the direction of crack propagation relative to the motion of the impacting load, and in the magnitude of the initial crack velocity which is higher when it reaches the test material. The explosion bulge and explosion bulge tear tests are similar insofar as the loading method is concerned, but in the latter the crack propagates from a 2 in. flaw in a direction of essentially uniform loading. The explosion bulge tear test may be regarded as establishing an extreme upper limit to the severity of loading conditions in a structure.

Pellini and Puzak have drawn attention to the "low energy shear" characteristics of the ultra-high-strength steels, and have discussed the implications[75,76]. At yield strength levels above 200 ksi, the Charpy V upper shelf or plateau energy may drop to very low values (15 to 20 ft lb), as indicated schematically in Figure 13. In effect, the ductile energy absorption of such steels is comparable to the energy absorption for brittle fracture of lower strength steels. Hence, fracture with no sign of cleavage may initiate from small flaws at elastic stress levels close to the yield strength, and large flaws may be disastrous. Typical failures from service were quoted to illustrate this point. Explosion bulge tear tests of a number of steels demonstrated that a high level of fracture toughness was associated with a shelf energy in excess of 50 ft lb. A maraging steel of 220 ksi yield strength, however, gave a "flat break" of full width at 30°F. Similarly, in the drop-weight tear test, the energy absorbed by the maraging steel was very low. Presumably, the other ultra-high-strength steels would have shown much the same behaviour. The authors conclude that maraging steels of the 260 ksi strength level, which are not subject to quench-and-temper treatments, are acceptable for large booster casings, despite the fact that the flaw size for fracture initiation at yield strength levels is of the order of 3/8 in. for a 3/4 in. thick plate. This conclusion is based upon inspectability and the nature of the service, and obviously the same considerations would not apply to submarine hulls.

The difference in the requirements for motor cases and submarine hulls has been stressed by Manning and Martin[77]. The latter, in particular, must be capable of withstanding shock wave loading, and it is desirable that the material should absorb the applied energy by distributing the plastic deformation through as large a volume of metal as possible. If it is accepted that the explosion bulge tests evaluate this characteristic, and if the correlation with the Charpy V shelf energy is maintained, then it is apparent that the application of ultra-high-strength steels to hulls may be limited entirely by the extent and level of the inspection procedure.

To summarize the foregoing discussion, a number of criteria are now available to the designer of ultra-high-strength steel hardware. For conditions involving essentially static, "one-shot" loading, the β_c criterion has been found to correlate well with service performance. When the conditions are such that a fair degree of cyclic loading is involved, allowance must be made for possible slow crack growth, and here the environment may be an important factor. For those applications in which shock loading is likely or even possible, and in the present state of the art, it would appear that more severe tests are necessary, such as those developed at NRL, particularly if failure of the hardware involves a hazard to personnel. Nevertheless, it is suggested that for certain applications more use might profitably be made of simpler tests, such as the instrumented bend test, which have also shown good correlation with performance.

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Legend:

ASM	-	American Society for Metals
ASTM	-	American Society for Testing and Materials; prior to May 1961, American Society for Testing Materials
NASA	-	National Aeronautics and Space Administration, Washington, D.C.
DMIC	-	Defence Metals Information Center, Battelle Memorial Institute, Columbus, Ohio
NRL	-	U.S. Naval Research Laboratory, Washington, D.C.
ASME	-	American Society of Mechanical Engineers
SAE	-	Society of Automotive Engineers

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(Figures 1-13 follow,)
(on pages 43-49.)

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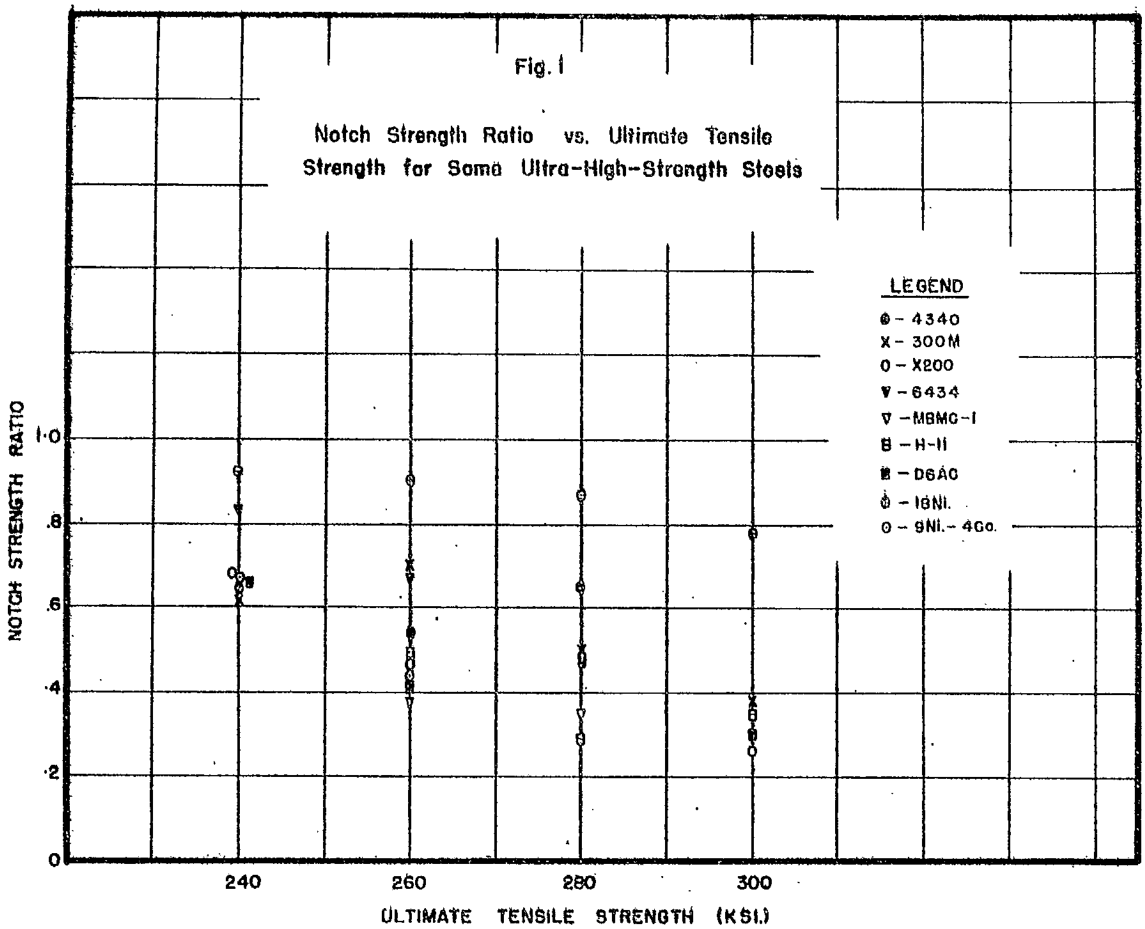


Fig. 1
Notch Strength Ratio vs. Ultimate Tensile Strength for Some Ultra-High-Strength Steels
LEGEND
- 4340
X - 300M
O - X200
▼ - 6434
▽ - MBMC-I
- H-II
- D6AC
- 18NI.
- 9NI.- 4Co.
NOTCH STRENGTH RATIO
1.0
.8
.6
.4
.2
0
240
260
280
300
ULTIMATE TENSILE STRENGTH (KSI.)

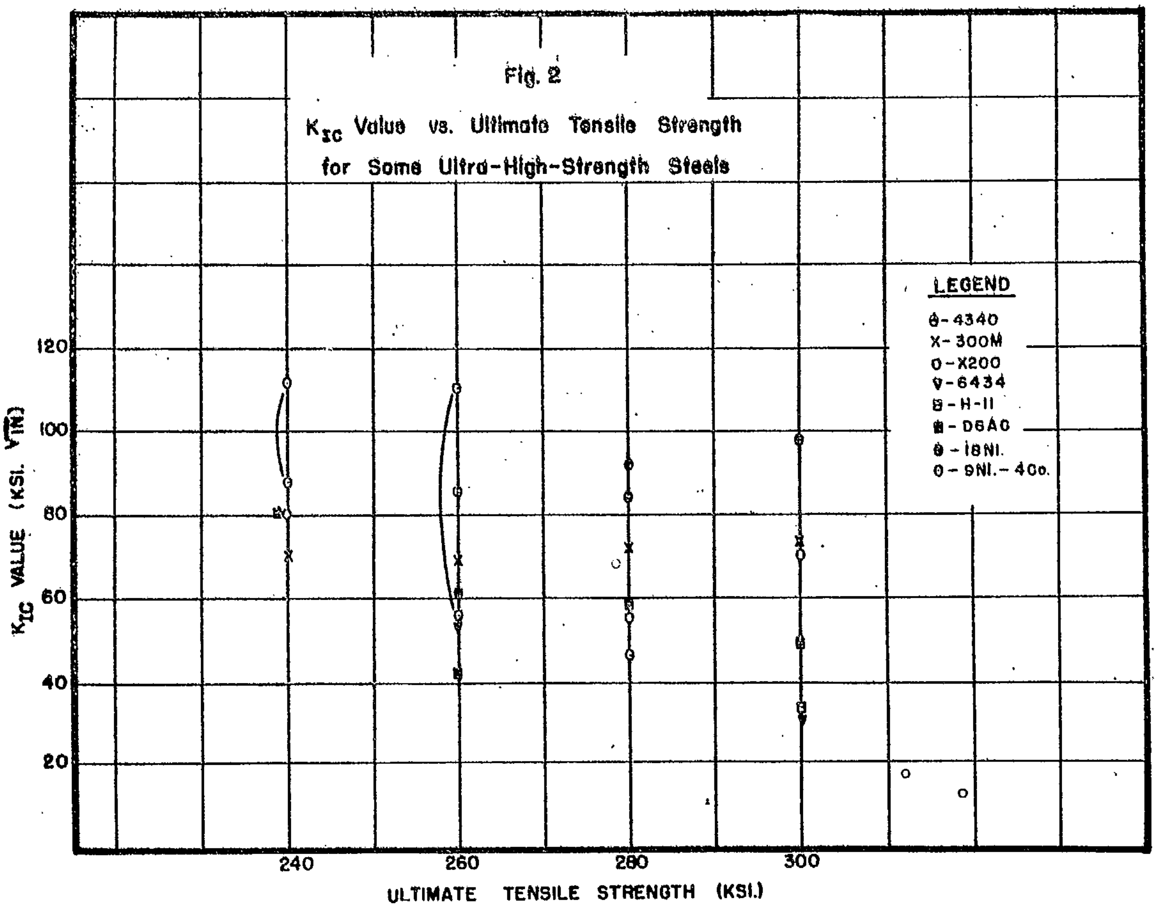


Fig. 2
K_{IC} Value vs. Ultimate Tensile Strength for Some Ultra-High-Strength Steels
LEGEND
- 4340
X - 300M
O - X200
▽ - 6434
- H-II
- D6AC
- 18NI.
- 9NI.- 4Co.
K_{IC} VALUE (KSI. $\sqrt{IN}$)
120
100
80
60
40
20
240
260
280
300
ULTIMATE TENSILE STRENGTH (KSI.)

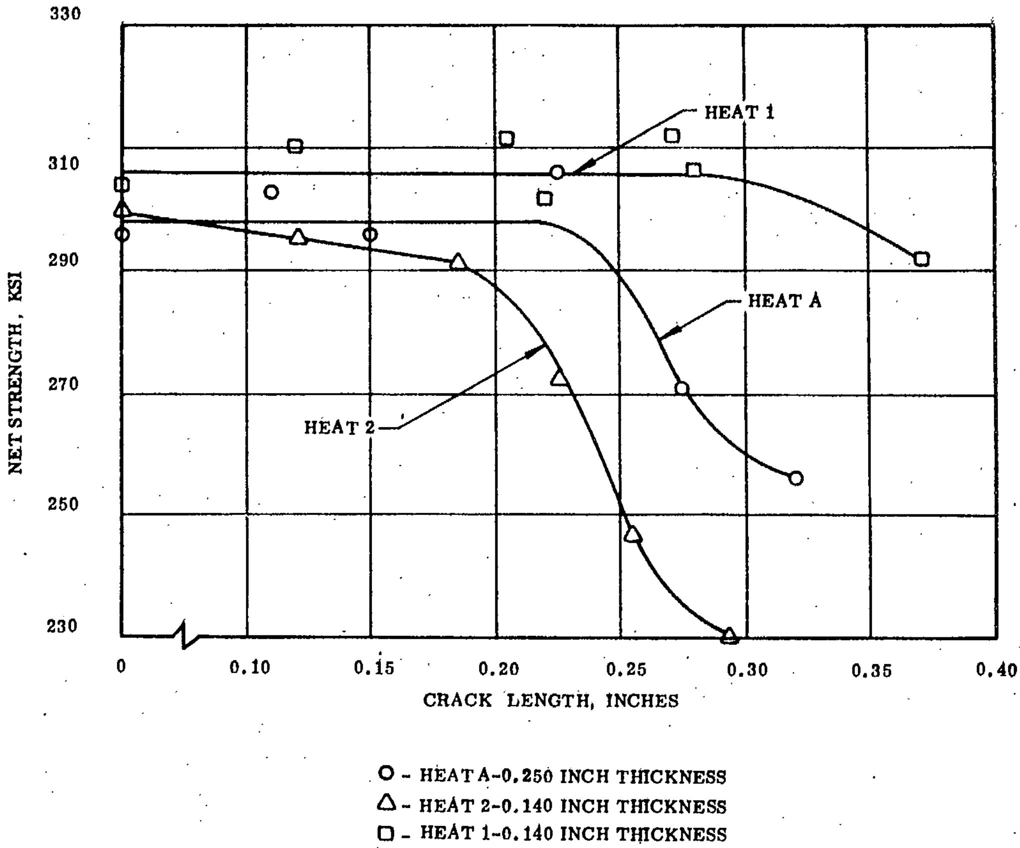
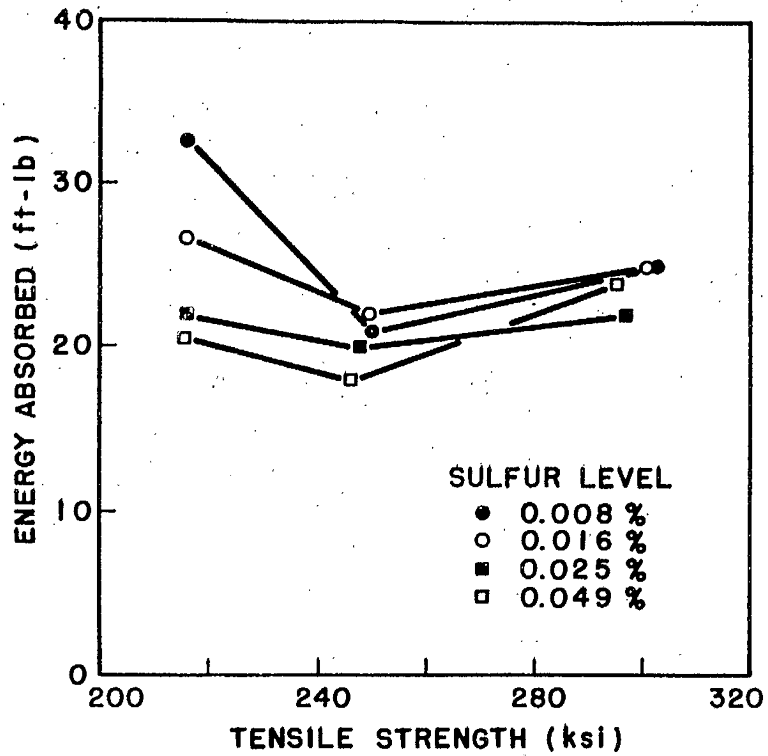


Figure 3. Relative Crack Resistance of Three Maraging Steel Heats (Ref. 25).



Effect of Sulfur Level on Charpy V-Notch Impact Energy
of AISI 4345 Steel at Room Temperature

Figure 4

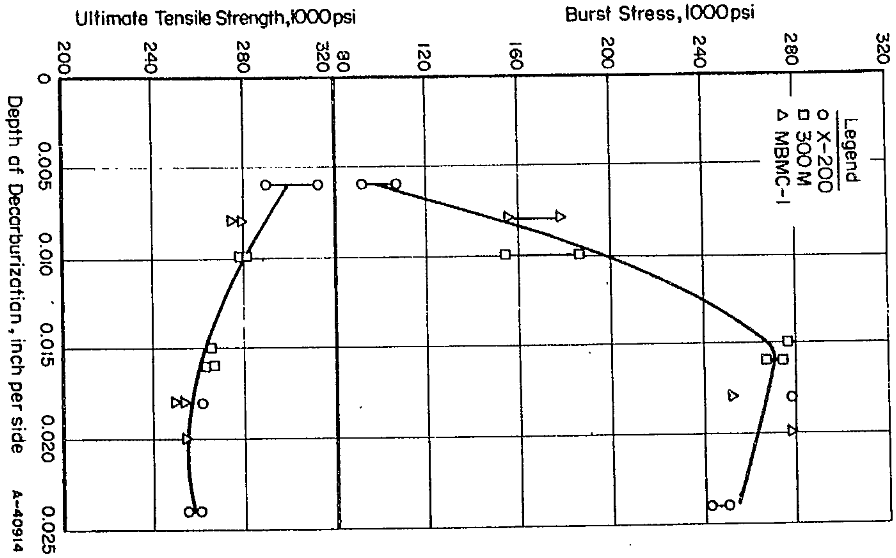


Figure 5. Effect of Decarburization on Burst Stress and Tensile Strength (Ref. 31).

ORIENTATION	G_{IC}, ipsi
A	245
B	230
C	310
D	150

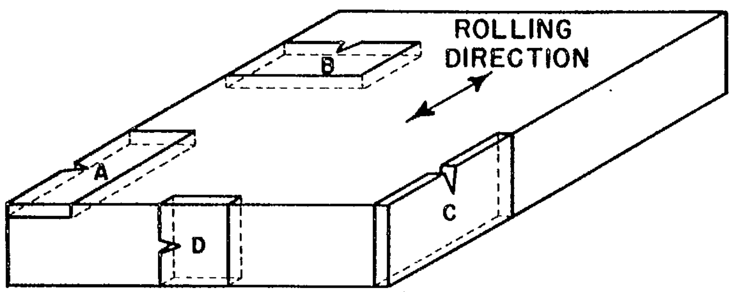


Figure 6. Effect of Orientation on G_{Ic} value - 18% Ni steel (Ref. 44).

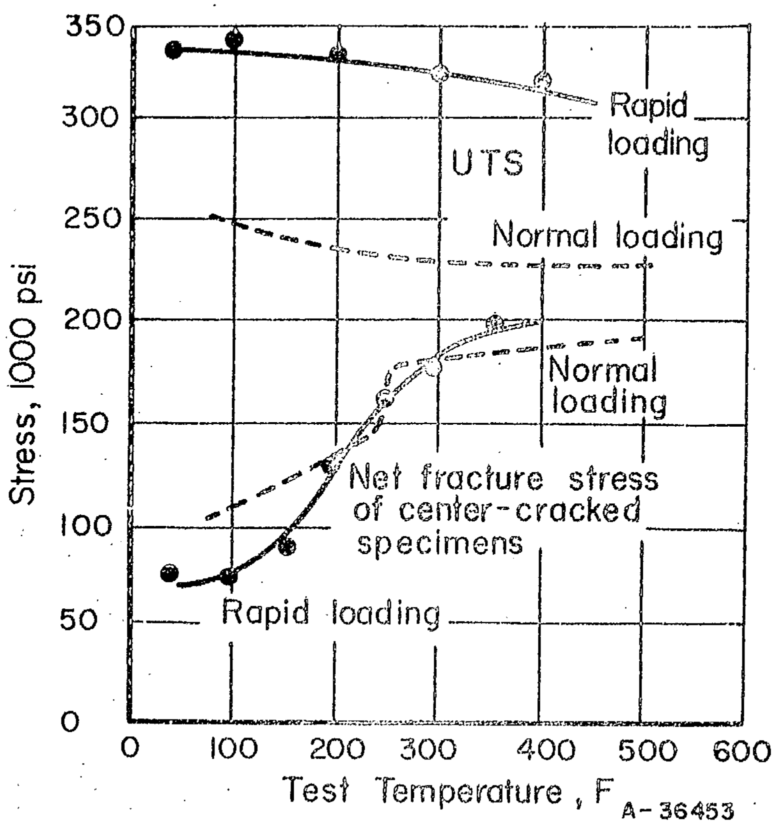
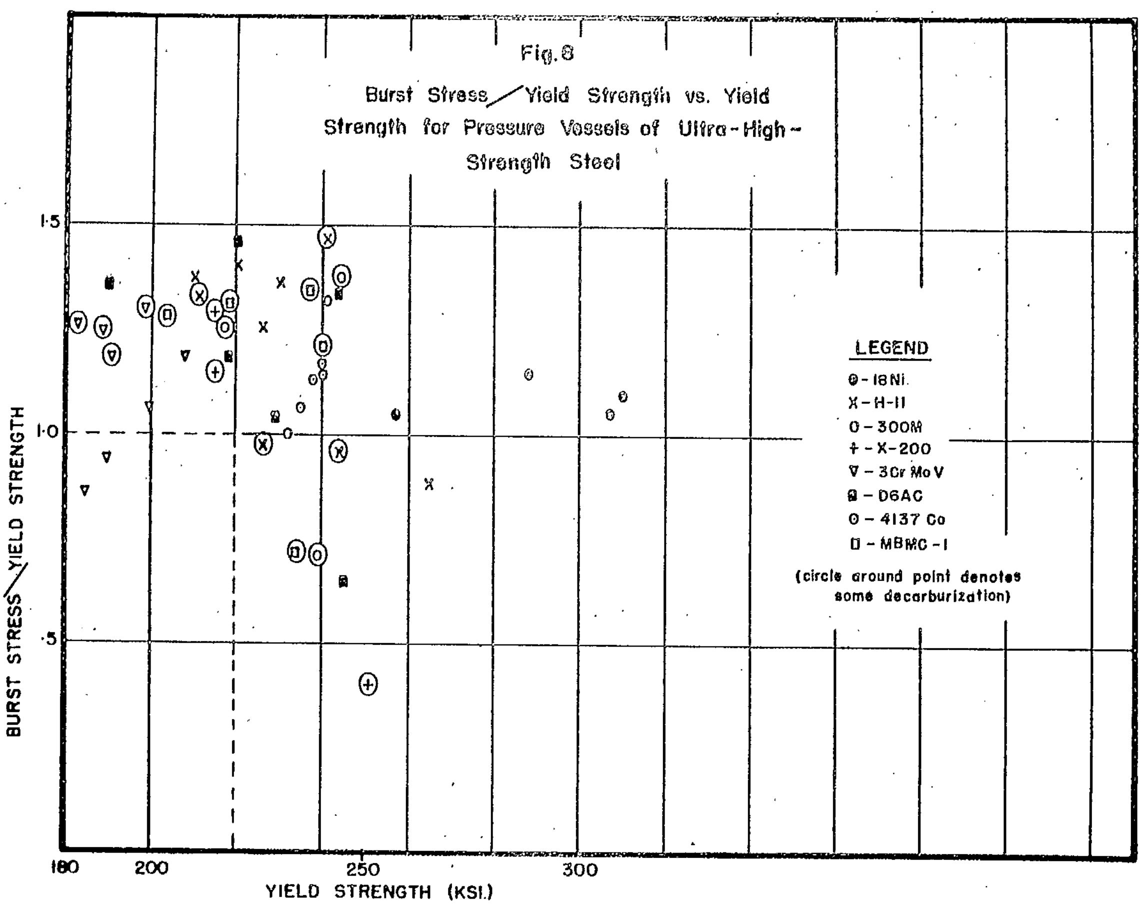
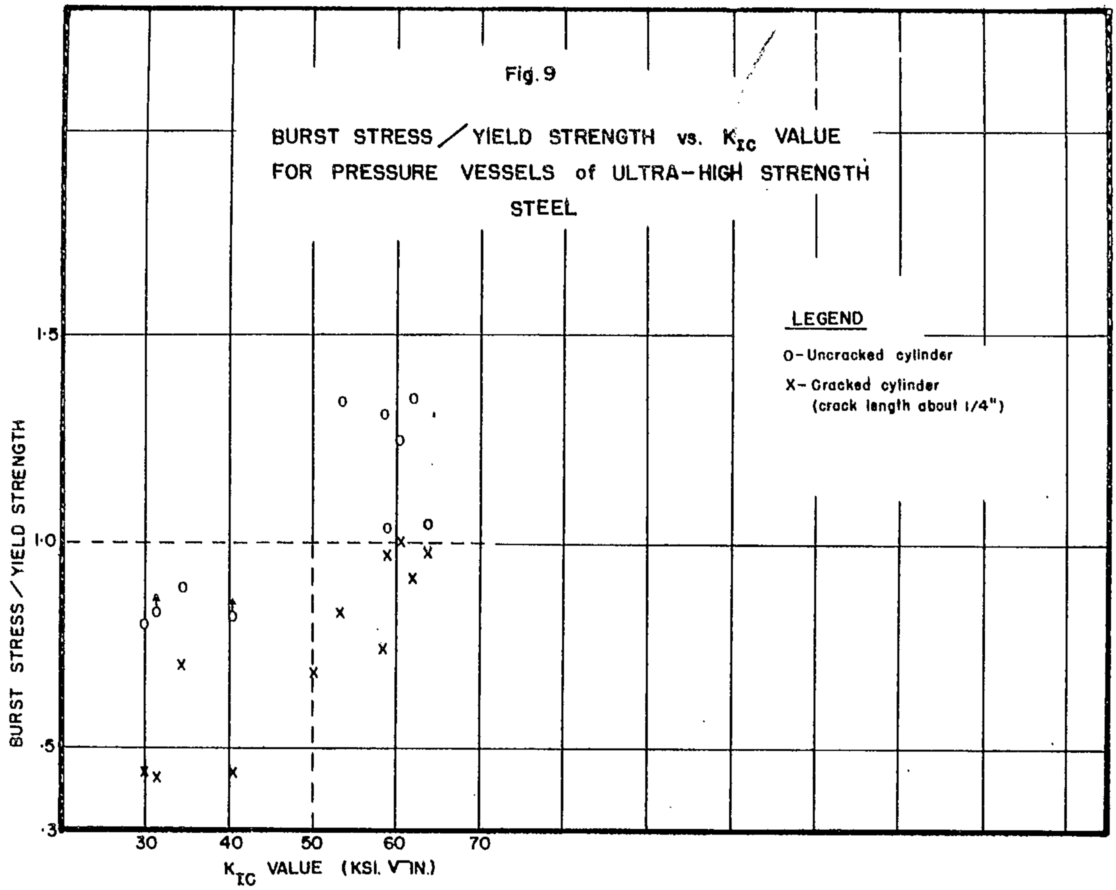


Figure 7. Effect of Loading Rate on Net Fracture Stress Transition Temperature - 422 M steel (Ref. 53).





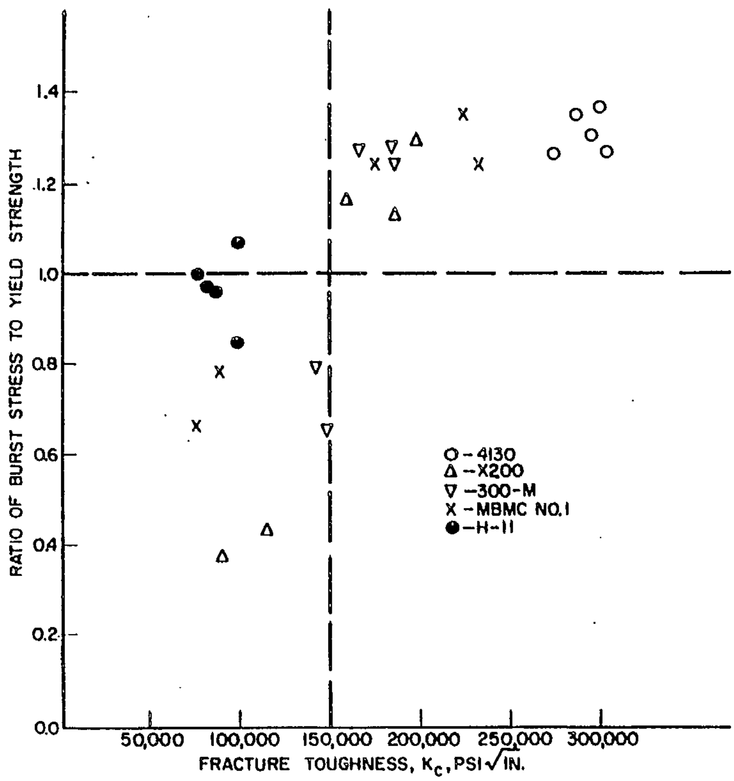


Figure 10. Burst Stress/Yield Strength vs. K_c value for Pressure Vessels of Ultra-High-Strength Steel (Ref. 32).

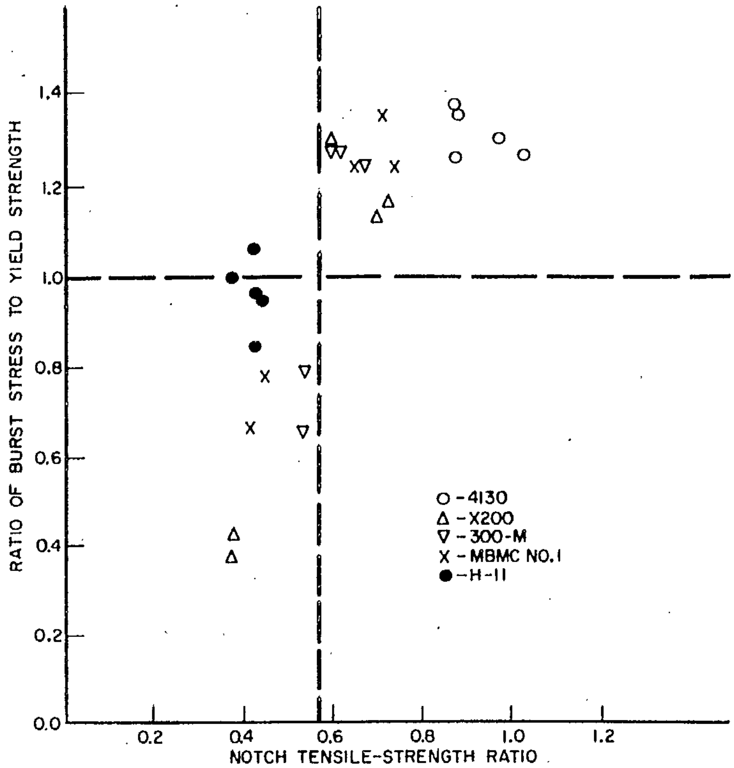


Figure 11. Burst Stress/Yield Strength vs. Notch Strength Ratio for Pressure Vessels of Ultra-High-Strength Steel. (Ref. 32).

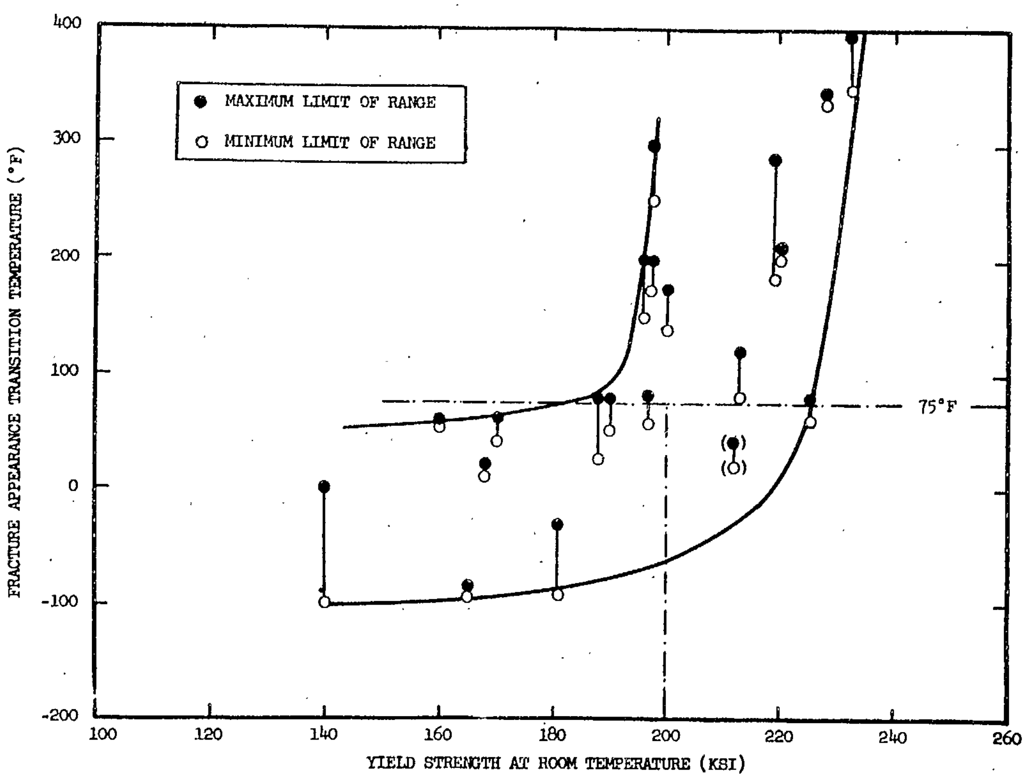


Figure 12. Fracture Appearance Transition Temperature vs. Yield Strength for Some High-Strength Steels (Ref. 72).

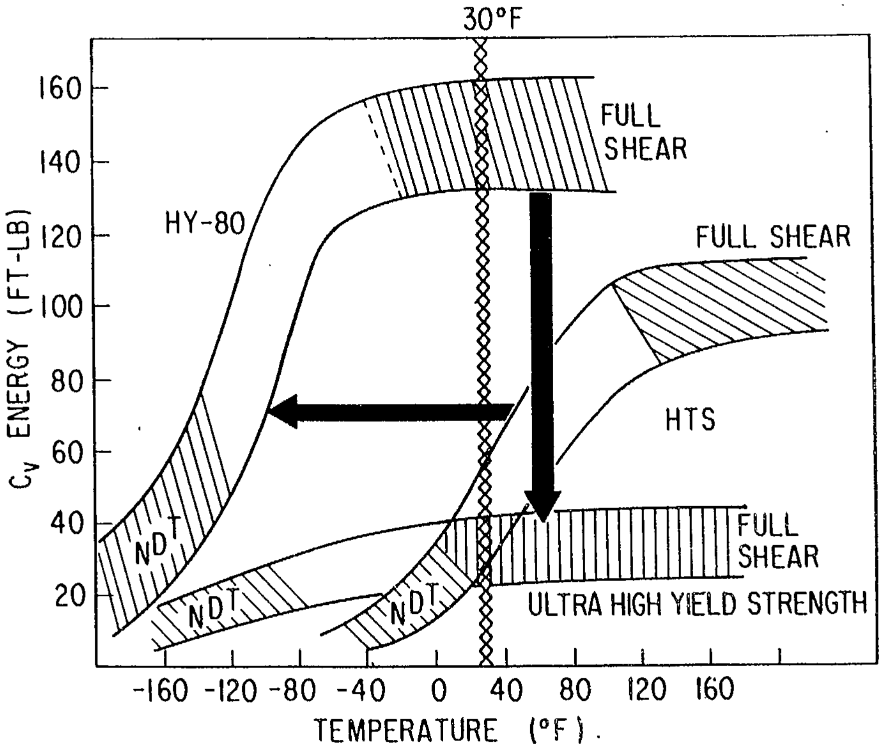


Figure 13. Representative Charpy-V Transition Curves for High-Strength and Ultra-High-Strength Steels (Ref. 75).

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