Steel, physical and mechanical properties

Fracture toughness

Rocket motor cases

RESISTANCE TO CRACK PROPAGATION OF HIGH-STRENGTH SHEET MATERIALS FOR ROCKET MOTOR CASINGS

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ABSTRACT

The ratio of the fracture strength of a standardized center-crack-slotted tensile specimen to the yield strength of the material was employed as a comparative measure of crack propagation resistance. Fracture strength ratios and tensile properties were determined at various levels of yield strength for nine commercial steels, four titanium alloys, and twenty-six experimental steels, including six ausforming compositions. Most of the tests were conducted at room temperature on longitudinal specimens, nominally 1/16 inch thick. Supplementary data were obtained on effects of temperature, thickness, and orientation. Some of the fracture surfaces were studied with the electron microscope.

The highest fracture strength ratios, in relation to the ratio of yield strength to density, were obtained with titanium alloys, notably 16V-2.5Al. The steels were markedly inferior at ratios of yield strength to density exceeding 900,000. All materials showed a limiting trend for fracture strength ratio to decrease with increasing strength level. High yield strengths were obtained in ausrolled steels, but the fracture strength ratios followed essentially the same trend as for quenched and tempered steels. Indications of effects of composition in quenched and tempered low-alloy steels were obtained and were related, in part, to prior austenite grain size and to temperability. Vanadium (0.2%) was specifically beneficial, and Mo and Cr were also beneficial. Further work on composition is needed, but the existing data does not suggest that the limitations of existing low-alloy steels might be substantially surmounted.

RESISTANCE TO CRACK PROPAGATION OF HIGH-STRENGTH SHEET MATERIALS FOR ROCKET MOTOR CASINGS

INTRODUCTION

One of the limitations on performance of large solid-propellant rockets is the weight of the inert parts such as the casing and the nozzles. Design of casings is therefore aimed at minimizing weight by using materials of high strength/weight ratio at high factors of utilization. The ratio of hoop stress to yield strength may approach quite closely to unity. The strength level at which casing materials may be used with acceptable confidence is restricted by the additional need for tolerance of small crack-like defects, which may occur during manufacture, testing, storage, or service of the vessel. It would be unrealistic to expect that such defects could be avoided entirely, but in principle, an upper limit of size corresponding to some acceptably low level of incidence could be set. This implies a corresponding lower limit on the crack propagation resistance of the material, which must be greater the higher the maximum applied stress. It is therefore necessary that all materials considered for the construction of rocket motor casings be evaluated in respect to this particular characteristic. Since resistance to crack propagation is sensitive to material factors of composition, thermal and mechanical treatment, "quality," thickness and temperature, the exhaustive evaluation of even a single specified composition could be a considerable undertaking.

Accounts of a number of proposed methods for evaluating the crack propagation resistance of high-strength sheet materials have been published in reviews and symposia collections (1-7). The most common approach is to conduct tension tests of sheet specimens which are provided either with very sharp edge notches or with central slots with very sharp ends which simulate cracks. The type of specimen used for this investigation has a narrow, transverse slot which terminates in fatigue cracks formed at low nominal stress levels. This is considered to represent the closest possible approach to a natural crack, and considerable evidence exists to support this view (4,8). The comparative measure of crack propagation resistance is taken to be the ratio of the nominal net fracture strength of a specimen of standard dimensions to the yield strength of the material (determined in a separate test using a standard tensile specimen). This ratio will be referred to more briefly as the fracture strength ratio. It is an arbitrary quantity in that it depends upon the specimen dimensions, but, as is shown in the Appendix, it has a rational interpretation in terms of fracture mechanics. When, as in the present investigation, the dimensions are standardized, the fracture strength ratio depends only upon the material characteristics and temperature.

The scope of the project was to determine the fracture strength ratio for a variety of conditions of a number of different high-strength sheet or strip materials, and thus to compare these materials on the basis of crack propagation resistance in relation to yield strength (or ratio of yield strength to density). These materials may be divided into the following categories: (a) commercial sheet steels, (b) commercial titanium alloys, (c) high-hardenability, low-alloy experimental steels suitable for ausforming to high-strength levels, (d) low-alloy experimental steels selected to display effects of various individual alloying elements in quenched and tempered steels, and (e) proprietary experimental steel compositions. While it was intended to devote some effort to variables such as testing temperature, thickness, and orientation, the primary emphasis was to be on the factors of composition and heat treatment. Certain limitations of scope were necessary, and the majority of the tests were conducted on longitudinal specimens of standard thickness (nominally 1/16 inch, with one or two exceptions) and at room temperature. In almost all cases some supplementary tests were conducted at higher or lower temperatures. In conjunction with the testing program, studies of fracture surfaces of certain of the steel specimens were conducted with an electron microscope. The purpose of these studies was to attempt to relate the fracture features to the crack propagation resistance of the material.

MATERIALS

The compositions and nominal thicknesses of the commercial steels which were included in the investigation are given in Table 1. With the exception of P-11, these are proprietary compositions and are accordingly identified in this report only by arbitrary identification symbols, the letter or pair of letters indicating a particular type of steel, and the number a particular thickness of sheet from a particular heat. The compositions given are the results of analysis conducted at NRL on the actual sheets. Steel P-11 was supplied according to SAE Aeronautical Materials Specification 6434 and represents a material which has been widely used for rocket motor casings. Steel C is a modification of AISI 4340 having higher silicon for increased tempering resistance and a small amount of vanadium for grain refining. Steel G is essentially AISI type H-11 tool steel (currently quite widely used for high-strength structural applications). The other commercial steels have compositions which differ considerably from any of the standard designations.

Table 2 gives the compositions and nominal thicknesses of the titanium alloys which were included in the investigation. Again these are identified by arbitrary symbols, but will be recognized from the compositions as Department of Defense Titanium Sheet Rolling Program alloys.

Table 3 gives the compositions of experimental steels produced at NRL. These include the high-hardenability compositions for ausrolling, the low-alloy steels containing about 0.4% carbon for study of the effects of individual alloying elements, and high-nickel-content steels of proprietary compositions which were produced because of difficulty in procuring commercially-produced sheet. The nominal thickness of all these NRL steels was 0.063 inch. These steels were produced by the vacuum induction melting of 25-pound heats of electrolytic iron and high-purity alloying elements of ferroalloys, adding the requisite carbon as graphite rod under argon, and pouring into tapered, square-ingot molds. Addition of aluminum as a deoxidizer was not necessary to produce sound ingots by this procedure. Accordingly, no aluminum additions were made in order that this variable should be eliminated in this series of heats. It was intended that supplementary heats would be made to determine the effect of aluminum additions on the most promising compositions. The ingots were hammer forged to flat bar stock 1.75 inch by 0.375 inch. This bar stock was cut into 5-inch lengths and surface ground to remove equal stock from both sides and to finish at a thickness of 0.250 inch. These bars were then either hot rolled from an inert atmosphere furnace at 1500°F, or were ausrolled in the range 850° to 1050°F after being austenitized at an appropriate temperature and being allowed to cool to 950°F. The final thickness was 0.063 ± 0.003 inch, and the length of each strip was 19 to 20 inches. Thus each strip provided two specimen blanks, 9 by 1.5 inches, for either crack propagation or tensile specimens.

Details of heat treatments are given in the headings of the various tables of test results. It should be noted that the specimens were not normalized after rolling, prior to the hardening and tempering heat treatment. While normalizing might have resulted in some refinement of the austenite grain size, it might also have substantially increased the amount of surface decarburization. The latter consideration was regarded as the more important.

TEST PROCEDURE

The designs of the crack propagation and tensile test specimens are shown in Fig. 1. During testing the load is transmitted through hardened pins which pass through the holes in the ends of the specimen. There are additional pin joints between the specimen loading pins and the spherical bearings of the testing machine so that axiality of loading is insured.

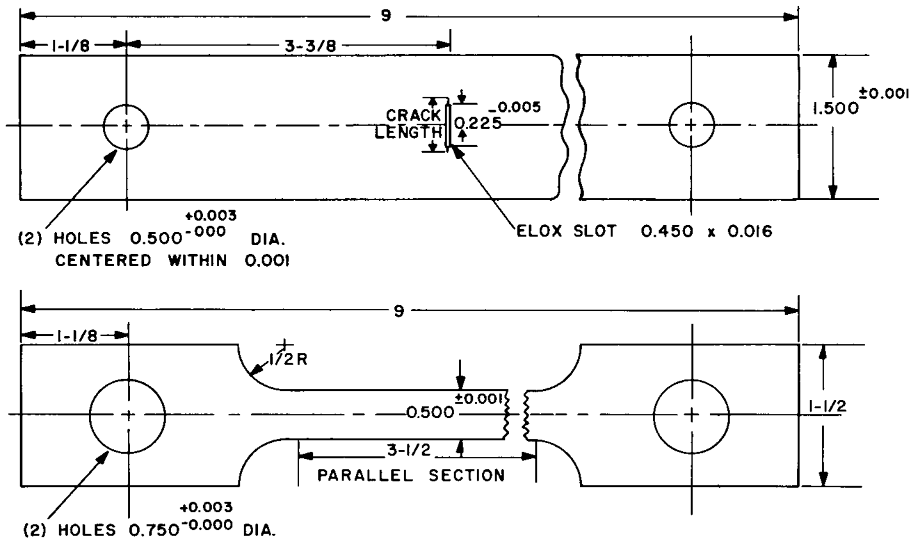

Fig. 1 - (a) Crack propagation and (b) tensile test specimens

The crack propagation specimen has a narrow transverse slot, formed by electrical discharge machining (Elox process). Cracks are formed at the ends of this slot by tension-tension fatigue stressing. The overall distance between fatigue crack tips is 0.5 inch, or slightly greater. Small variations in crack length do not have a significant influence on the nominal net fracture strength because the strength has a flat maximum in the range of the ratio of crack length to specimen width 0.3 to 0.4 (8) (see also the Appendix). This is the reason for the selection of the crack length of 0.5 inch. Fatigue cracking is carried out at a nominal maximum stress not exceeding half of the yield strength of the material in the condition in which it is cracked. Usually fatigue cracking was carried out prior to heat treatment because specimens are less liable to be broken in a relatively soft condition. In some cases it was necessary to fatigue crack the specimens in the condition in which they were to be tested, for instance, the ausrolled specimens. In the past, tests have been conducted on replicate specimens, some of which were cracked before heat treatment and some after heat treatment. No significant difference in results was found - see for instance Ref. 3, Fig. 20.

The arrangement for heating or cooling specimens for tests at other than room temperature has been described elsewhere (9). A pair of brass or copper plates are held in contact with the faces of the specimen and are maintained at the desired temperature by resistance heating elements or by precooled air passing through cooling ducts attached to the plates. Thermocouples, with their junctions at the plate surfaces which are in contact with the specimen, are incorporated for monitoring temperature.

NOMINAL NET FRACTURE STRENGTH AND FRACTURE STRENGTH RATIO

The nominal net fracture strength is the quantity determined in the crack propagation test and is defined as the maximum load divided by the product of specimen thickness and the difference between width and initial crack length. It does not take account of slow crack extension which may occur during the test before the maximum load is reached. It is therefore analogous to the ultimate tensile strength determined in an ordinary tensile test, which is based on the original cross-sectional area and does not take account of the reduction in area which occurs before the maximum load is reached. The initial crack length is determined after the specimen has been broken by measuring the maximum distance between the two fatigue crack fronts.

The measure of crack propagation resistance adopted here is the ratio of the nominal net fracture strength to the yield strength of the material, called briefly the fracture strength ratio. This may be related to the Irwin crack toughness K_c, which is discussed in Ref. 3, as shown in the Appendix of this report. However, it was not intended to determine K_c values with precision, but rather to investigate the comparative effects of factors which affect crack propagation resistance. More precise K_c measurements need then be made only on the more promising materials.

The fracture appearance index, or shear fraction, is a complementary indication of crack propagation resistance. It is given in the tables when it could be meaningfully determined. In a clear-cut example the fracture surface consists of a central band of transverse fracture with uniform oblique shear borders on either side. The shear fraction is defined as the difference between the specimen thickness and the average breadth of the uniform transverse band divided by the thickness. Values are given to the nearest 0.05, since greater precision is not warranted. The shear fraction may be used to define a fracture mode transition temperature, called the full-shear temperature, being the lowest temperature at which the shear fraction is 1.00. In some materials the full-shear temperature is very sharply defined for a given thickness.

RESULTS AND DISCUSSION

(a) Commercial Steels

The results of the tensile and crack propagation tests for the various commercial steels are given in Tables 4 through 20. Excepting Table 10 and, in part, Tables 19 and 20, these results are for longitudinal specimens, and it will be implicit in the discussion that longitudinal properties are under discussion unless explicit reference is made to transverse properties.* Each of the low-alloy steels, C, L, P, X, and G, was tested in a number of different tempered conditions (Tables 4 through 14). The highly alloyed precipitation-hardening steels, W, K, AC, and AD, were each tested in only one or two conditions recommended by the producer (Tables 15 through 20). Heat treatment details are given in table headings.

Table 21 summarizes the room temperature results from the previous tables in terms of the maximum fracture strength ratio for each steel according to a series of minimum

*A longitudinal specimen is one that is oriented so that its length is in the direction of rolling of the sheet or strip from which it is obtained. Thus the applied load during a test of a longitudinal specimen is in the rolling direction. In a crack propagation test the direction of crack propagation is transverse to the applied load, and in the case of a longitudinal crack propagation specimen, the direction of crack propagation is transverse to the rolling direction.

yield strength levels. The fracture strength ratios given in each column are the maxima corresponding to yield strengths not less than the value given at the head of the column. This type of presentation avoids the need for interpolation between actual fracture strength ratio values, which is considered inadvisable. For example, assuming a minimum yield strength of 220,000 psi is required, the highest fracture strength ratio is for steel AD in the longitudinal direction. However, this steel is cold worked and aged, and if the choice is restricted to steels which are hardened by heat treatment only, then steel P has the highest fracture strength ratio.

A graphic comparison of the room temperature data is shown in Fig. 2, where fracture strength ratios are plotted against yield strength and the upper bounding values are identified by the symbol for the appropriate steel. Two features of this plot should be noted: (a) the tendency for the highest values of the fracture strength ratio to decrease with increasing yield strength, and (b) the wide variation in fracture strength ratios at all yield strength levels. The interpretation suggested is that yield strength is a major factor in restricting toughness, but that there are other independent factors which can drastically reduce toughness below the limit determined by the strength level.

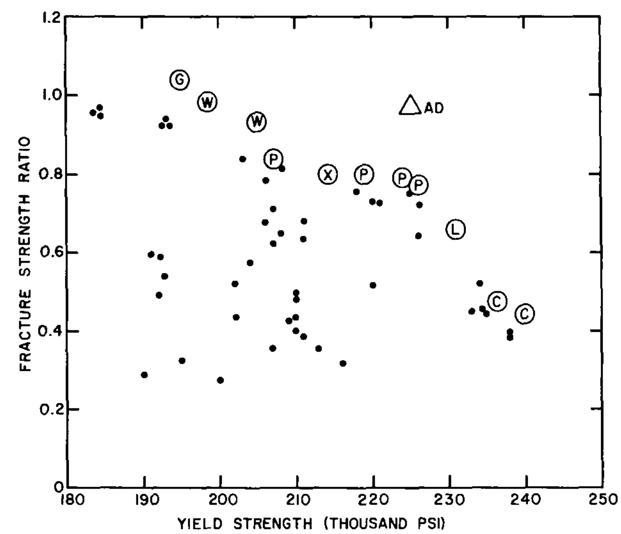

Fig. 2 - Fracture strength ratios versus yield strength for the commercial steels at room temperature

The exceptionally high fracture strength ratio of steel AD at a yield strength of 225,000 psi is for the longitudinal direction, the value for the transverse direction being considerably lower, as shown in Table 20. This difference is presumably due to the fact that this steel is strengthened by cold rolling and aging. It would be expected that there would be less difference between longitudinal and transverse properties in the case of a steel which derived its strength from quenching and tempering, and this is seen to be the case for steel P, as shown in Fig. 3 (data of Tables 9 and 10).

With the exception of AD, steel P has the highest fracture strength ratios in the yield strength range of 205,000 psi to 225,000 psi (the maximum attained with steel P). There is a sharp change in the trend of upper bounding values in Fig. 2 at this yield strength level, probably because the steels which developed higher yield strengths, C and L, were not of optimum composition for high crack propagation resistance. Both contain about 1.5% silicon which is suspected to be embrittling.

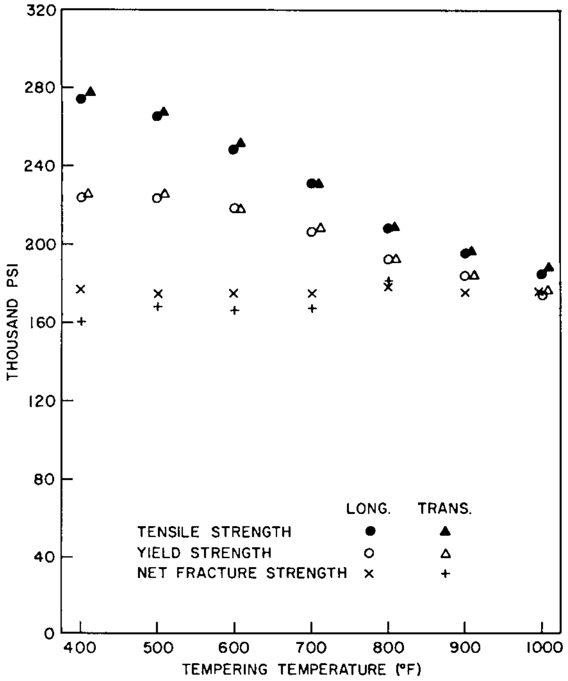

Fig. 3 - Longitudinal and transverse properties of steel P11 at room temperature versus tempering temperature (tempered 2 hours)

Steels C2 and C5 (Tables 4 and 5) are, respectively, air melted and consumable electrode vacuum remelted from the air melt material. The inclusion ratings, according to ASTM Recommended Practice E 45-51, were Type A, 2 thin, O heavy for steel C2, and Type A, 1 thin, O heavy for steel C5. Both steels had an austenite grain size of ASTM No. 7. Comparison of the data indicates no appreciable benefit from the vacuum remelting. This seems surprising and probably not indicative of what is to be expected generally. Possibly it is related to the comparatively low crack propagation resistance of this composition in the sense that some embrittling factor overshadowed any benefit which might have derived from vacuum remelting. In contrast, previous results on AMS 6434 steel have indicated a definite benefit of vacuum remelting (10,11).

In most cases the low-alloy steels were heat treated by quenching to obtain maximum transformation to martensite and then tempering at various temperatures to vary strength level. It is generally believed that maximum toughness at a given strength level is achieved in this way. However, it seemed worth while to attempt to compare a steel in the austempered condition (lower bainite structure) with the same steel in a quenched and tempered condition at about the same yield strength. Steel C5 was selected for this purpose and an austempering temperature of 600°F appeared to be appropriate to obtain a high-strength level according to the TTT diagram. A sufficient number of specimens were heat treated for tests to be conducted over a temperature range from -280 to 400°F so that temperature transition behavior could be compared. A similar set of specimens were quenched and tempered at 600°F. Unfortunately the yield strength of the austempered material at room temperature was only 157,000 psi - much lower than had been anticipated. The results, given in Table 6, are therefore of little relevance to the purposes of the investigation. The results for the material quenched and tempered at 600°F are given in Table 7, which complements the data of Table 5. An austempering treatment applied to steel L also resulted in a yield strength too low to be of interest (Table 8, last heat treatment listed).

Steel P was tested in the as-quenched condition and after tempering at 212°, 300°, and 350°F, as well as after tempering at higher temperatures as in the case of the other steels. The results (Tables 9 and 10) show that the yield strength at first increases with increasing tempering temperature and then declines with further increase in tempering temperature, the maximum occurring in the range 300° to 500°F. However, the fracture strength ratio increases with tempering temperature over the whole range. Thus for some values of the yield strength the fracture strength ratio may be relatively low or high, depending upon whether the material was tempered at a temperature lower than or higher than that corresponding to maximum yield strength. Indications of similar behavior are apparent with other low-alloy steels, for instance steel X (Table 11) and some of the experimental steels discussed later.

Analogous behavior was found with steel G (Table 12), although the tempering temperatures for this steel are higher. The yield strength was highest for triple tempering at $1000^{\circ}F$; tempering at lower temperatures resulted in lower yield strengths and lower fracture strength ratios. It was also found that oil quenching this steel did not result in better fracture strength ratios than the air hardening procedure which is usual.

Figure 4 (data of Table 13) is a typical example of fracture mode transition, from oblique shear to transverse, with increasing thickness at constant temperature, showing a maximum fracture strength ratio corresponding to the greatest thickness for which the fracture mode was full shear. Specimens were obtained from sheet G8, which was initially 0.13 inch thick, by surface grinding equal stock from both surfaces to obtain the required range of thicknesses. Table 19 gives results over a wide range of temperature for thicknesses of 0.05 and 0.10 inch from the same sheet with a slightly different tempering treatment. The full-shear transition temperature is sharply defined between 40° and $50^{\circ}F$ for the thinner specimens, and between 80° and $110^{\circ}F$ for the thicker specimens. Outside the interval between these transition temperatures the fracture strength ratios for the two thicknesses are not markedly different.

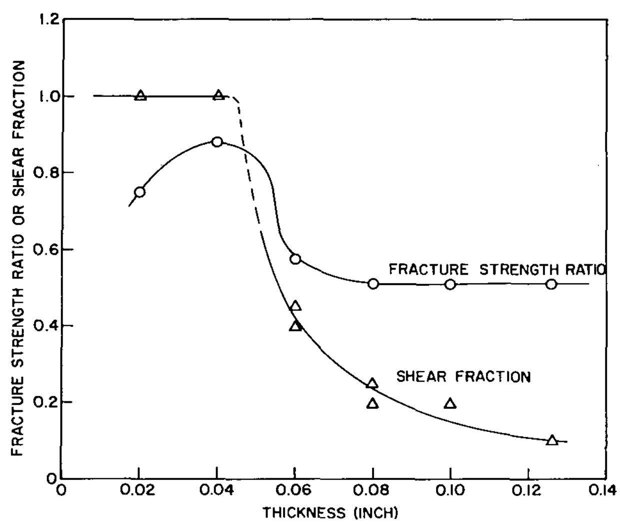

Fig. 4 - Shear fraction and fracture strength ratio versus specimen thickness for steel G8 (tempered at $1050^{\circ}F$ for 2 + 2 + 2 hours) at room temperature

Temperature and thickness effects were also studied in steel W2. In the as-received thickness of 0.10 inch the transition temperature was above $200^{\circ}F$ even in the relatively soft TH 1050 condition, Table 16, and above $300^{\circ}F$ in the RH 950 condition, Table 15. A range of thicknesses were tested at 300°, 80°, and $-280^{\circ}F$ in the TH 1050 condition (Table 18) with the purpose of observing the variation in fracture strength ratio with thickness for full-shear fracturing, transition behavior, and transverse mode of fracturing, respectively. The choice of steel seems to have been unfortunate for this purpose because only slight effects of thickness were found. In the transverse mode no effect of thickness is expected and the data show no appreciable effect.

The results obtained for these commercial steels may be compared with results obtained by Espey, Jones, and Brown, using their sharp-edge-notch tensile test (11).

Several of the compositions tested by these investigators were essentially the same as those used in the present investigation. In general there is good agreement as to the relative merits of the various steels, as may be seen by comparing Espey's Table IV with Table 21 of the present report.

(b) Titanium Alloys

The results for the titanium alloys are given in Tables 23 through 28. Tests were conducted on three of the alloys in a variety of aged conditions so that the yield strengths varied up to the maxima attainable by heat treatment. Expected stocks of the fourth alloy (6Al-4V) were not received and the only available results for this material are for the annealed condition (Table 26). These results were obtained on a small amount of material supplied by Messrs. Orner and Hartbower of Watertown Arsenal for the specific purpose of comparing results for the annealed condition.

Figure 5 shows the fracture strength ratios at room temperature versus the ratios of yield strength to density for the three titanium alloys tested in aged conditions. Points for commercial steels, corresponding to the upper bounding values in Fig. 2, are also shown for comparison with the titanium data. It is apparent that substantially higher fracture strength ratios at a given strength level were obtained with the titanium alloys. Alternatively, for a given minimum fracture strength ratio the highest ratio of yield strength to density was substantially greater for titanium alloys than for steels. For example, assuming a minimum required fracture strength ratio of 0.8, the highest ratio of yield strength to density for a low-alloy steel was 0.785 (steel P), or for the cold-worked and aged steel AD, 0.805, whereas for titanium-16V-2.5Al the ratio of yield strength to density was 0.925 in both longitudinal and transverse directions. The results for this alloy were generally better than for the other two titanium alloys, but all three alloys gave substantially better results than the commercial steels. The same conclusion was reached by Espey, Jones, and Brown, who have evaluated the same titanium alloys using their sharp-edge-notch tensile test and have compared the results with those for a variety of steels (12). The results shown in Fig. 5 of this report are in excellent agreement with Fig. 9 of Ref. 12.

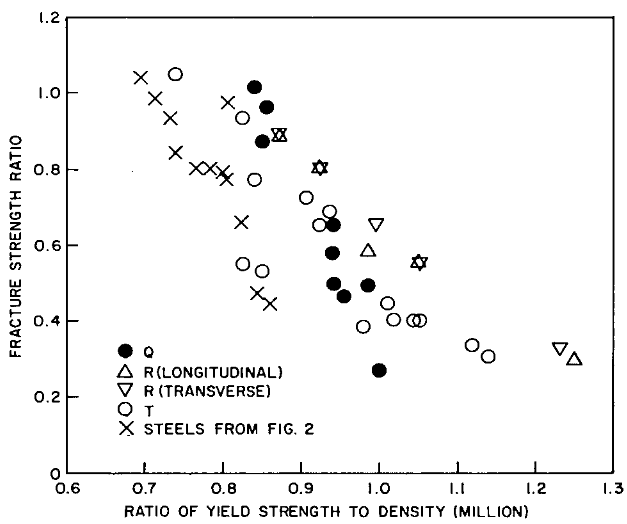

Fig. 5 - Fracture strength ratios versus ratio of yield strength to density for titanium alloys at room temperature

The fracture strength ratio of the 4Al-3Mo-1V alloy (Table 22) decreased rapidly and rather consistently with yield strength level. However, it is worth noting that aging at 1000°F for 24 hours resulted in a substantially higher fracture strength ratio than aging for 12 hours at 1000°, 1025°, or 1050°F, while the yield strength was almost the same for each of these treatments.

Tables 23 and 25 show the effect of aging temperature on the longitudinal and transverse properties, respectively, of the 16V-2.5Al alloy. The yield strength decreases steadily with increasing aging temperature and the fracture strength ratio increases consistently as the yield strength decreases. The range of yield strength is considerably greater than for the 4Al-3Mo-1V alloy, and the rate of decrease of fracture strength ratio with increasing yield strength is correspondingly slower. It is also notable that the fracture strength ratio in the transverse direction was not inferior to that in the longitudinal direction at any strength level.

Another point of interest is that the fracture mode of the 16V-2.5Al alloy aged for 4 hours at 1020°F was full shear at 0°F, although the ratio of yield strength to density was almost 1,000,000 psi per pound per cubic inch. For this reason a transition temperature series of tests was conducted for this condition, the results being given in Table 24. From these results the full-shear temperature is less than -100°F and the net fracture strength decreases only slightly in the temperature range from 80° down to -200°F. However, since the yield strength increases with decreasing temperature the fracture strength ratio decreases correspondingly.

Table 26 gives longitudinal and transverse test results as a function of temperature for the 6Al-4V alloy in the annealed condition. The ratio of yield strength to density at room temperature is 0.825, about the same as for a steel of 230,000 psi yield strength. The corresponding fracture strength ratios were 0.915 longitudinal and 1.00 transverse, comparable to the other titanium alloys and considerably better than the steels. The fracture strength ratios decrease quite gradually with decreasing temperature, but the transverse values decrease more rapidly than the longitudinal values below 0°F.

The effect of aging variables on the properties of the beta 13V-11Cr-3Al alloy is shown in Table 27. In addition to standard 100-hour aging treatments at various temperatures, some shorter aging treatments and some double aging treatments (13) were included in case these might result in better crack propagation resistance. A plot of the results is shown in Fig. 6 in which the shorter aging treatments and the double aging treatments are distinguished from the 100-hour aging treatments. Both double aging and short-time aging result in substantially better fracture strength ratios than obtained with the 100-hour aging treatments, but the improvement decreases with increasing yield strength to virtually none at 180,000 psi. These results emphasize the point that commonly recommended heat treatments may not be optimum for high resistance to crack propagation and show that there are possibilities of substantial gains from further exploration of heat treatment variables.

The results discussed so far were obtained with material nominally 1/16 inch thick. Stocks of 0.130 inch thick sheet of the 13V-11Cr-3Al alloy were available, and Table 28 and Fig. 7 show the effect of thickness on the properties of this alloy at room temperature for a double-aged condition. The aging treatment, 48 hours at 900°F followed by 30 minutes at 1100°F, was selected on the basis of the previous results in order to produce a high yield strength with a moderately high fracture strength ratio. In general the shear borders of the fractures of this alloy are not clearly defined and do not permit useful estimates of the shear fraction. For this reason such estimates are not given in Table 27 (the same applies to the alloy 6Al-4V, Table 26, and to the 4Al-3Mo-1V alloy, Table 22). However, in this particular series of tests there was clearly a change from full shear at 0.02 inch thick to decreasing shear fractions with increasing thickness, and it was possible to make fair estimates of the shear fraction values by intercomparison

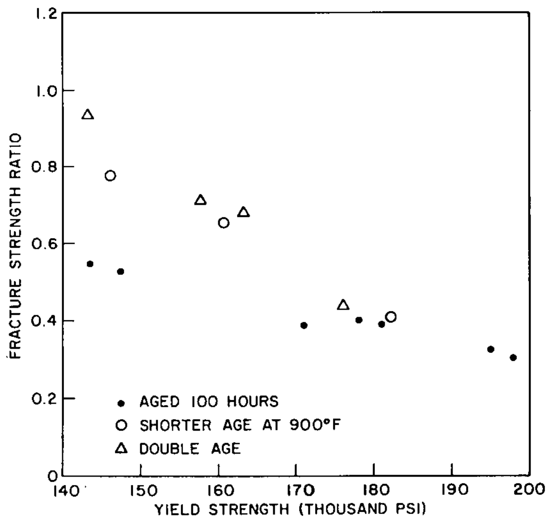

Fig. 6 - Fracture strength ratios versus yield strength for titanium-13V-11Cr-3Al at room temperature, showing the contrast between under-aging and double-aging treatments as compared to the standard 100-hour aging treatments

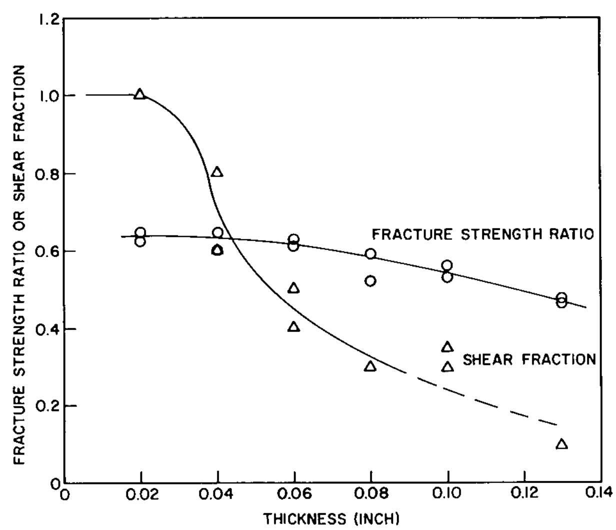

Fig. 7 - Shear fraction and fracture strength ratio as a function of specimen thickness for titanium-13V-11Cr-3Al at room temperature. The alloy was aged 48 hours at 900°F, then 30 minutes at 1100°F.

of the fractures for different thicknesses. The change in fracture strength ratio with thickness is very gradual compared with that for steel G8 (Fig. 4), suggesting that there is not much difference between the fracture toughness for full-shear fracturing and that for fracturing in the transverse mode for this condition of the titanium alloy. There is possibly a connection between this and the lack of definition of the shear borders.

(c) Steels for Ausforming

Ausforming is a term commonly used to denote a process of simultaneously shaping and strengthening steels of sufficient hardenability whereby the yield strength and tensile strength are raised to levels which could not be reached by straightforward heat treatment of the material. The steel is austenitized and then cooled to the working temperature range where the austenite is unstable but will not undergo appreciable transformation within the period of time in which the working is to be accomplished. To achieve high-strength levels, the degree of working in this range must be considerable, for instance, 75% or more reduction by rolling or drawing. The worked steel is quenched to martensite and then tempered. Literature relating to the subject has been reviewed by Morral (14), and more recently by Zackay (15). The process might be applied to the manufacture of rocket motor casings, and therefore the crack propagation resistance of ausformed steels is of interest.

The six compositions selected as representing a variety of steels likely to be quite responsive to strengthening by ausforming are NRL heats 200 through 204, and 217, given in Table 3. Heat 200 is virtually the same composition as steel A-41 of Shyne et al., which they showed to be very responsive to ausworking (16). Heat 203 was intended to be the same except for lower silicon content and was intended to check whether high silicon had an embrittling effect. However, as it happened, the carbon content was also considerably lower than that of Heat 200. Heat 201 is simply a 5% Cr steel. Heat 202 is a high-carbon version of the British high-nickel EN30B composition. Heat 204 is similar to the proprietary composition steel X (Table 1), which has been claimed to be particularly suitable for ausforming by forging. Heat 217 is essentially type 410 stainless steel and appeared to offer the interesting possibility of high yield strength at an unusually low-carbon level.

The test results for these steels are given in Tables 29 through 34. In each case, strip was produced by ausrolling in the range 850° to $1050^{\circ}F$ from a thickness of 0.25 inch down to 0.063 inch, that is, a 75% reduction, then oil quenched. Some specimen blanks were then tempered and the specimens tested in the ausrolled and tempered condition. Except in the case of Heat 200, other specimen blanks after ausrolling and quenching were re-austenitized, oil quenched, and tempered in order to determine the differences between properties obtained by ausrolling and tempering and those obtained by heat treatment only. In the case of Heat 200 the comparison specimens were hot rolled and heat treated instead of being ausrolled, quenched, and re-austenitized, etc. Details of the treatments are given in the Table headings.

Tests were conducted at room temperature only, and the data are summarized for comparison in Table 35 in the same manner as Table 21 for the commercial steels. The fracture strength ratios are plotted versus yield strength in Fig. 8 with the upper bounding points identified by the last digit of the heat number. These can be compared with the upper bounding points for the commercial steels, which are also shown.

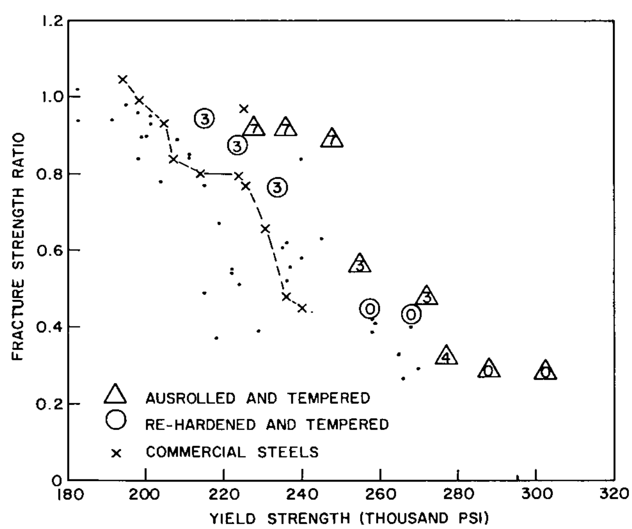

Fig. 8 - Fracture strength ratios versus yield strength for experimental ausforming steels at room temperature

Except for Heat 202 the expected high levels of yield strength were attained in the ausrolled and tempered conditions. However, the general tendency for fracture strength ratio to decrease with increasing yield strength prevailed with these steels as with the commercial steels. While the upper bounding data points do represent an appreciable improvement over those for the commercial steels, it must be kept in mind that these are the results of longitudinal tests on a narrow strip. Equipment limitations prevented ausrolling of sheet sufficiently wide to determine transverse properties, but it seems likely that ausrolled and tempered material would be markedly anisotropic. Furthermore, when these results are compared with those for the hot rolled and conventionally heat treated experimental steels, which are discussed later, the upper bounding values are not substantially different. Thus the data do not appear to provide any basis for supposing that ausforming is an exceptional means of obtaining toughness in very-high-strength steels. On the other hand, neither are such steels abnormally lacking in toughness when tested in the longitudinal direction.

The highest yield strengths were reached with Heat 200, Table 29 (Shyne's A-41 composition). This steel also had high yield strengths when hot rolled and heat treated in the usual manner. The fracture strength ratios, however, compare unfavorably with those for the low-silicon version of this composition, Heat 203, Table 32, as may be seen from Table 35. Although these two heats differ considerably in yield strength for a given tempering temperature because of the difference in carbon contents, the ranges of yield strengths overlap, so comparisons can be made. The superior toughness of the low-silicon composition is apparent in both the ausformed and the reheat treated conditions. It will be recalled that silicon was suspected of inducing brittleness in the commercial compositions C and L. Yield strengths exceeding 250,000 psi were also reached with Heat 204 by ausrolling and tempering, but the fracture strength ratios for this composition were also lower than those for Heat 203.

The greatest proportionate improvement in yield strength from ausforming was obtained with the 12% chromium steel, Heat 217, Table 34. When hardened by heat treatment only, the yield strength did not much exceed about 150,000 psi, which is consistent with the extensive published data for type 410 stainless steel. The maximum yield strength after ausforming and tempering was 248,000 psi, an increase of about 60%. At the same time, the fracture strength ratios for the ausformed and tempered conditions were quite high, in fact they are the highest values obtained with any steels of comparable yield strengths in the course of the investigation.

Heat 201 showed moderate response to ausrolling but the fracture strength ratios were much lower than for Heat 217. Heat 202 showed practically no response to ausrolling.

(d) Experimental Low-Alloy Steels

While a considerable variety of commercial steels have been proposed as suitable for highly stressed applications in thin sections, there is no systematic information available on the effects that the various alloying elements might have on crack propagation resistance. The absence of such information in a recent review of the characteristics of such steels is notable (17). The series of NRL vacuum induction melted steels considered in this section represents an initial effort at gaining some insight into this matter. The compositions of these heats are given in Table 3, under the subheading "Low-Alloy Steels." The intended carbon content of all these heats, except No. 233, was 0.40%, but inevitably the carbon content varied among the heats between 0.41 and 0.45%. Heat 233 was intended to have the same composition as No. 228 except for a somewhat higher carbon level.

All the compositions were hot rolled to 1/16 inch thick strip from which longitudinal tensile and crack propagation specimens were obtained. These were hardened by oil quenching from 1550°F (except in one or two cases to be noted later), cooled to -100°F to promote more complete transformation of austenite to martensite, then tempered variously at 400°, 500°, 600°, 700°, 800°, and 1000°F to obtain a wide range of yield strengths. The same austenitizing temperature of 1550°F was used for all compositions, although it was recognized that the optimum austenitizing temperature would depend upon composition. Given sufficient material, this factor should have been included as a variable in the program. However, it is probably of marginal significance compared to that of tempering temperature, and it was considered not essential to the purpose of the program. For each steel and each tempering temperature, one tensile specimen and one crack propagation specimen was tested at room temperature. Then an additional crack propagation specimen was tested at either 0° or 200°F, depending upon the room temperature result. The testing was limited to this number of specimens by the amount of material available from each 25-pound ingot. However, these appeared to be sufficient for survey purposes.

Heat 207 contained no added elements other than carbon. Heats 208, 209, 213, 214, 215, 216, and 234 were intended to contain, respectively, 1.5% Si, 1% Mn, 2% Ni, 2% Co, 0.5% Mo, 0.2% V, and 2% Cr. It was realized that full hardening, even of 1/16 inch thick strip, could not be expected of some of these compositions, but they were included for completeness. Heats 207, 208, and 214 were actually brine quenched rather than oil quenched. It is evident from the tensile properties in Tables 36, 41, and 43 that Heats 207, 214, and 216 (plain C, Co, and V) were very far from being fully hardened by the quenching. The remaining compositions responded satisfactorily to quenching, as did the multialloyed steels. Examination of the microstructures showed that Heats 207, 208, 213, 214, and 215 were predominantly martensitic with amounts of pearlite or ferrite of the order of 5%, and that 216 had perhaps 50% pearlite and ferrite intermingled with the martensite. All the other compositions were apparently fully martensitic. The remaining heats were intended to contain 1% Mn, 0.5% Si, and various combinations of the alloying elements Ni, Cr, Mo, V, and Co. The first four of these alloying elements are those commonly used for conferring hardenability in quenched and tempered steels. Cobalt was included because it has been claimed to be uniquely beneficial in high-strength steels (18).

Test results are given in Tables 36 through 54 for the individual steels, and a summary of maximum fracture stress ratios according to yield strength level for all the steels is given in Table 55. This is of the same form as Table 21 for the commercial steels and Table 35 for the ausrolled steels, and can thus be compared with them.

Fracture strength ratios at room temperature are plotted versus yield strength in Fig. 9. Results for yield strengths less than 180,000 psi are not shown since they are of little interest in the present context. The upper bounding values for commercial steels, taken from Fig. 2, are also plotted for comparison. Upper bounding values for the experimental steels are identified by the last digit of the heat number, as indicated in the bottom left-hand corner of the figure. It appears that the better experimental compositions are generally superior to the commercial steels that were tested; in fact, the upper bounding values fall in about the same band as those for the experimental ausforming steels shown in Fig. 8. This apparent superiority to the commercial steels is of minor interest, however, because it could be largely or entirely due to the circumstances that the experimental heats were made from relatively pure raw materials, cast into small ingots, and rolled into narrow strip. The main interest is in comparisons among the experimental steels.

Among the experimental steels, as with the other materials that have been considered, the fracture strength ratio varies widely at any given strength level, but the maximum of its range tends to decrease with increasing yield strength. Resistance to plastic deformation is a major factor in limiting toughness, but there are additional, independent factors which may drastically reduce fracture toughness. Obviously it is desirable to identify these factors so that they can be minimized by appropriate choice of composition and other variables.

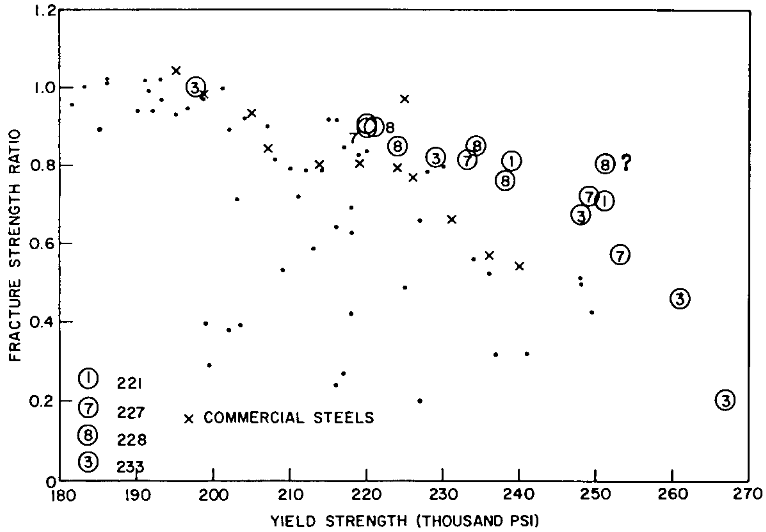

Fig. 9 - Fracture strength ratios versus yield strength for experimental low-alloy steels at room temperature

The clearest indication, that emerges from a study of the data, of the effect of a compositional factor is that of vanadium. All four of the compositions included in the upper bounding values in Fig. 9 contained about 0.2% V; none of the other compositions did, except Heat 216, the carbon-vanadium steel which was insufficiently alloyed for hardening by quenching (Table 43). In Fig. 10 the data is replotted using symbols to distinguish the points according to whether the steels contained vanadium, molybdenum, both, or neither. At the higher yield strength levels the division is quite marked; at a given level of fracture strength ratio the vanadium-containing steels have the highest yield strengths, those containing molybdenum without vanadium have intermediate yield strengths, and those containing neither vanadium nor molybdenum have the lowest yield strengths. Insofar as the yield strength ranges overlap, the fracture strength ratios are highest for the vanadium-containing steels and lowest for those containing neither vanadium nor molybdenum. Below 200,000 psi yield strength almost all fracture strength ratios are high, and differences according to composition are slight (the one or two specific exceptions which are to be found in the tables are discussed later).

Of the four hardenable vanadium-containing compositions, Heat 221 contained only chromium and vanadium, while the other three contained molybdenum, and two of them nickel. The results for Heat 221 were not inferior to the others, as can be seen from Figs. 9 and 10, and thus it can be concluded that the beneficial effect derives from the vanadium, rather than from vanadium in conjunction with molybdenum. It is clear from Fig. 10 that molybdenum in the absence of vanadium, while somewhat beneficial, is less effective than vanadium.

There are two factors which are significant in relation to the effect of vanadium, and also of significance in general. In the first place the primary reason for including vanadium in a low-alloy steel composition is control of the austenite grain size for quenching. Grain growth at usual austenitizing temperature is inhibited by finely distributed vanadium carbides, and a quenched and tempered vanadium-containing steel is likely to be finer grained than a similar steel lacking only the vanadium. It might well be, therefore, that the beneficial effect of vanadium on crack propagation resistance is closely associated with its effect in refining austenite grain size. Of course the austenite is transformed to martensite on quenching, and this in turn undergoes transformation in several stages during tempering. Nevertheless, there is considerable evidence that the quenched

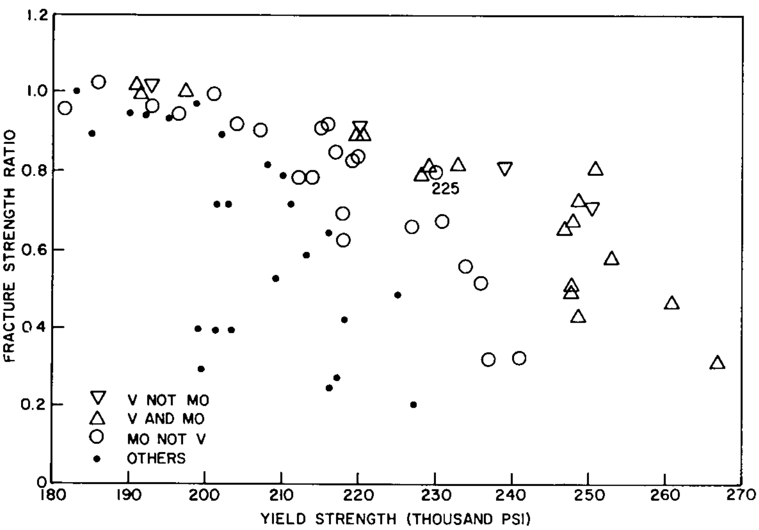

Fig. 10 - Fracture strength ratio versus yield strength for experimental low-alloy steels at room temperature, distinguished according to whether or not composition includes vanadium and/or molybdenum

and tempered structures retain features which derive from the prior austenite grain boundaries, and that the effects of these features are not insignificant unless the steel is tempered to a rather soft condition. For instance, the phenomenon of temper embrittlement is undoubtedly connected with prior austenite grain boundary regions. That toughness should improve with decreasing prior austenite grain size, other factors being constant, seems reasonable and is probably quite generally believed, though not yet thoroughly demonstrated because of the many complicating factors. The analogous relationship between toughness and grain size in simpler materials, such as ferritic steels, has been demonstrated and explained in terms of dislocation theory (19).

Austenite grain size measurements were made, and the vanadium-containing steels 221, 227, 228, and 233 all had grain sizes of ASTM No. 11 or smaller, that is, an average idealized grain diameter of 8 microns or less. All the other experimental steels, except Heats 216 and 225, had grain sizes of ASTM No. 6, 7, or 8, that is, an average idealized grain diameter of 25 microns or greater. Heat 216 (carbon-vanadium) had a grain size of ASTM No. 10, and Heat 225 (nickel-chromium-molybdenum) had a grain size of ASTM No. 10 to 11. The reason for this latter exception is not known, but it is significant that this steel (Table 50) had the best fracture strength ratios at high yield strengths of any of the steels which did not contain vanadium. For instance, the point at 0.795 and 230,000 psi in Figs. 9 and 10 is for Heat 225. There is therefore good evidence that austenite grain size is an important factor affecting crack propagation resistance at high-strength levels, and that the beneficial effect of vanadium may be largely or entirely a result of its role in refining austenite grain size.

The second factor which needs to be considered is tempering temperature. In general, if the duration of tempering is kept constant, the yield strength at first increases from the as-quenched value as the tempering temperature is increased, reaches a maximum within some range of tempering temperature that depends upon composition, and then declines with further increase in tempering temperature. On the other hand, the fracture strength ratio is lowest at low tempering temperatures and increases continuously as the tempering temperature is increased. The rate of increase with tempering temperature is usually rapid at first over a certain range, and then becomes more gradual at higher tempering temperatures. The change from rapid to gradual increase may be rather abrupt, defining what might be called a minimum useful tempering temperature. None of the steels tested showed clear evidence of embrittlement within an intermediate range of testing temperature, that is, of decreasing fracture strength ratio with increasing tempering temperature over some part of the range. However, there was a suggestion of this in the data for Heat 228, Table 53. It is considered that the available data for this steel requires confirmation, and for that reason one of the results is regarded with suspicion and is marked with a query in Fig. 9.

In Fig. 11 the same data points as in Figs. 9 and 10 are shown, this time distinguished according to tempering temperature (data points for tempering at 1000°F are omitted because none of these fracture strength ratios was less than 0.99, and the yield strengths were mostly less than 180,000 psi, and all were less than 200,000 psi). It is at once apparent that almost all the higher fracture strength ratios correspond to tempering temperatures of 600°F or higher, and that nearly all the lower fracture strength ratios correspond to tempering temperatures of 400° or 500°F. If the data for tempering at 400° or 500°F are disregarded, there is a fairly close correlation between fracture strength ratio and yield strength, with four marked exceptions which are identified by their heat numbers. It will be recalled that the commercial steels also tended to be brittle when tempered at low tempering temperatures. For instance, for AMS 6434 (Tables 9 and 10) the minimum useful tempering temperature was about 400°F.

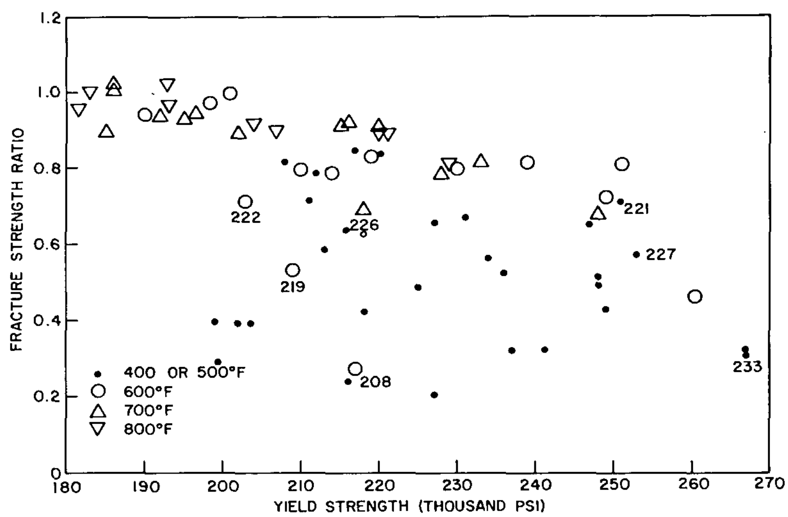

Fig. 11 - Fracture strength ratios for experimental low-alloy steels at room temperature, distinguished according to tempering temperature

From Fig. 10 in conjunction with Fig. 11 it is clear that the beneficial effect of vanadium, and the lesser beneficial effect of molybdenum, are strongly coupled with the tempering temperature factor. In the range of yield strengths 220,000 psi to 250,000 psi the higher fracture strength ratios of the vanadium-containing steels, and of the fine-grained molybdenum steel 225, result from tempering at 600°F or higher, while the lower values for other compositions result from tempering at 400° or 500°F. In the range of yield strengths 200,000 to 220,000 psi the same is largely true of the molybdenum steels as compared with those containing neither molybdenum nor vanadium. Thus, subject to some qualifications which are not yet clear, composition factors which increase the maximum tempering temperature for attainment of a given yield strength tend to favor high crack propagation resistance. This again is in accord with general experience and might well be related to relief of residual microstresses. Until the structures of tempered martensites are better understood, a more adequate explanation cannot be given.

Whether there is an interaction between austenite grain size and tempering resistance, or whether these two factors operate independently to augment one another in the case of the vanadium steels and of Heat 225, is not clear. It seems likely that small austenite grain size would enhance resistance to tempering to some extent, but also that it would be a favorable factor independently of the tempering factor.

The relationship between yield strength and tempering temperature depends upon the carbon content as well as other factors of composition. Heat 233 was intended to provide some insight into the effect of carbon and is essentially the same composition as Heat 228 except that the respective carbon contents were determined to be 0.48% and 0.45%. There is less difference in carbon contents than had been intended but the yield strengths of the higher carbon steel were consistently higher, as is seen by comparing Tables 53 and 54. The results for the two steels are plotted in Fig. 12, the numbers beside the data points indicating tempering temperature in hundreds of degrees Fahrenheit. As mentioned earlier, the fracture strength ratio of 0.805 at a yield strength of 251,000 psi for Heat 228 appears to be questionable in relation to the other data points. Additional specimens of this steel were heat treated and tested, with somewhat inconclusive results. From the results for 0^{0}F it appears that the full-shear transition temperature for specimens of this steel tempered at about 600^{0}F is not far from room temperature, and it may well be that the characteristics at room temperature are rather sensitive to minor variations in conditions of tempering or quenching. Disregarding the doubtful data point, the relationship between fracture strength ratio and yield strength appears to be virtually the same for both steels. Thus, while the tempering temperature for any given yield strength had to be higher for the higher carbon steel, the corresponding fracture strength ratios were neither higher nor lower. This suggests that carbon content may affect toughness through more than one mechanism. Insofar as it affects the selection of tempering temperature, increasing the carbon content improves toughness, but by some other mechanism the toughness tends to decrease with increasing carbon content, as is generally supposed. In the present example the two contrary effects appear to balance. Under other conditions one or the other tendency might predominate.

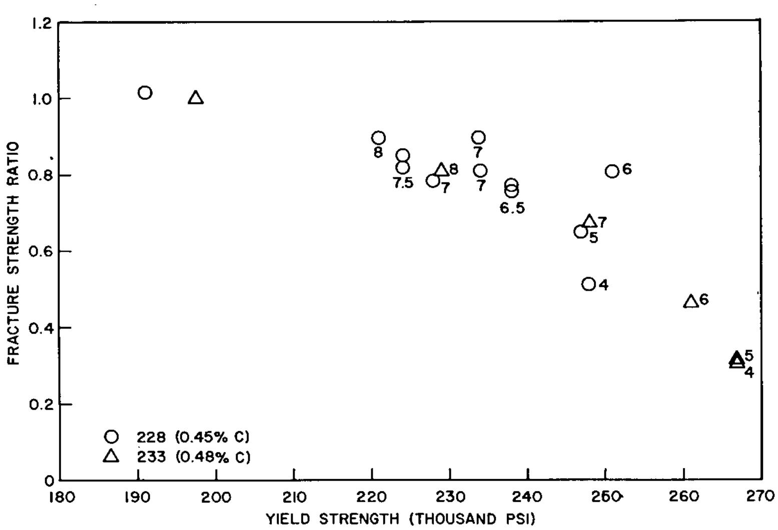

Fig. 12 - Comparison of fracture strength ratios at room temperature of Ni-Cr-Mo-V steels of two different carbon contents (as indicated at left). Figures by symbols indicate tempering temperatures in hundreds of degrees Fahrenheit.

There appears to be considerable evidence that chromium is a desirable alloying element. All four of the hardenable vanadium containing steels also contained chromium, and Heat 221 which contained only chromium and vanadium compared well with the Cr-Mo-V and Ni-Cr-Mo-V heats. Other indications can be seen in Table 55; Heat 225 (Ni-Cr-Mo), for instance, has the highest fracture stress ratio at the yield strength level of 230,000 psi of any of the steels that did not contain vanadium. Heats 234 (Cr) and 220 (Cr-Mo) also had good characteristics. The other three steels which contained chromium had unsatisfactory features. These are 219 (Ni-Cr), 222 (Co-Cr), and 226 (Co-Cr-Mo), and are identified in Fig. 11 as having exceptionally low fracture strength ratios for tempering temperatures of 600° or 700°F. However, there is no reason to suppose that this is a direct effect of chromium. In fact, comparison of the data of Table 55 for Heats 219 and 222 with that for the chromium steel 234 suggests otherwise, and this is supported by more detailed comparison of Tables 44 and 47 with 40. Similarly, Heat 226 (Co-Cr-Mo, Table 51) can be compared with Heat 220 (Cr-Mo, Table 45). Thus there is no real evidence of any embrittling effect of chromium.

The evidence regarding nickel is less conclusive. Referring again to Table 55, Heat 213, containing nickel only, and Heat 219 (Ni-Cr) had poor characteristics. On the other hand, Heat 223 (Ni-Mo) and the Ni-Cr-Mo-V and Ni-Cr-Mo heats all had good characteristics. Thus nickel clearly does not necessarily have an embrittling effect, but neither does the data indicate that toughness was improved by the inclusion of nickel in a composition.

Study of the data for the three cobalt-containing steels 222, 224, and 226 (Tables 47, 49, and 51) appears to indicate that this element is somewhat detrimental. However, more definite evidence is required as it is possible that the effect may be different when larger quantities of other alloying elements are present. Certainly there is no evidence of a beneficial effect of cobalt as far as the present results are concerned.

In discussing the commercial steels and the ausforming compositions it was noted that steels containing 1.5% silicon had a marked tendency to be brittle compared to those with more usual silicon contents of 0.5% or less. The data for Heat 208, Table 37, appears to support the supposition that high silicon has a specific embrittling effect. By comparison, the data for Heat 209, Table 38, appears to indicate that 1% manganese is not detrimental.

Further clarification of the effects of the various alloying elements by testing of additional compositions is clearly desirable. The effect of a particular element may depend strongly upon the amount present, possibly also on the amounts of other elements present, and the available data is restricted to one level of each element. However, within the limitations of the available data, the indications regarding choice of composition for crack propagation resistance may be summarized as follows: (a) adequate hardenability is necessary; (b) the carbide stabilizing elements V, Mo, Cr, and Mn appear to be beneficial and, in part, related to tempering behavior such that high yield strengths result from tempering at 600°F or higher; (c) the particularly beneficial effect of 0.2% V is associated with refinement of austenite grain size; and (d) the elements Ni, Co, and Si do not appear to be beneficial - high Si is probably detrimental and Co may be detrimental.

(e) High-Nickel Steels

Recently it has been reported that certain proprietary steels having nickel contents of the order of 20% are capable of very high yield strengths combined with good crack propagation resistance (20,21). Unfortunately, efforts to obtain stocks of these steels from commercial sources during the contract period were not successful. However, 25-pound vacuum induction melted heats of the reported compositions were produced at NRL for rolling into 1/16 inch thick strip in the same manner as for the low-alloy compositions. The compositions of these heats, 240, 241, and 246 through 250, are given in

Table 3 under the heading "High-Nickel Steels." Heats 240 and 241 are, respectively, 20% and 25% nickel steels with Ti, Cb, and Al and represent two of the proprietary steels. Heats 246 through 250 are variations of a third proprietary steel containing 18.5% Ni and 9% Co with Mo and Ti. This latter composition appears not yet to have been standardized, and it was therefore decided to explore variations in Mo and Ti contents (20). The specimens of these 18.5% Ni steels were not ready for testing before the end of the contract period. These results will be reported separately at a later date.

Results for Heats 240 and 241 are given in Tables 56 and 57, respectively, the heat treatments being detailed in the table headings. These heat treatments are as suggested in Ref. 20, and since this reference does not suggest variations in heat treatment for varying strength level, each steel was tested in a single heat-treated condition only. In each case tests were conducted at a number of different temperatures in the range -280° to 580°F.

The results do not support the claim that these compositions have good crack propagation resistance at high yield strengths. The yield strength of heat 240 at room temperature was 254,000 psi, but the corresponding fracture strength ratio was only 0.26, and the full-shear temperature was between 140° and 170°F. The yield strength of Heat 241 at room temperature was only 207,000 psi, the corresponding fracture strength ratio was 0.50, and the full-shear temperature was between 120° and 140°F. There is no obvious explanation for the discrepancies between these results and the claims which have been made for these steels. The compositions of the NRL heats do not appear to differ sufficiently from those given in Ref. 20 to suggest an explanation. The results must be regarded as inconclusive pending comparable testing of sheet obtained from approved commercial sources.

(f) Electron Microscope Examination of Fractures of Low-Alloy Steels

During the course of the investigation the microscopic details of the fractures of some of the low-alloy steel specimens were examined using an Elmiskop I electron microscope. The electron microscope is a uniquely useful instrument for the purpose since it has adequate depth of focus at high magnifications as well as high resolving power. While there have been some excellent studies of the microscopic details of fractures (see, for instance, the contributions by Low and by Crussard et al., in Ref. 22), there is still much to be learned about the fractures of high-strength materials. It was hoped that studies of fracture characteristics in relation to the results of the crack propagation tests of the present investigation might reveal useful information on the relationships between structure and crack propagation resistance or toughness. Because these studies were made while the tests were in progress and before all the results were available for analysis, the selection of specimens was somewhat arbitrary and the work incomplete. Nevertheless some interesting information was obtained which merits discussion.

Two extreme macroscopic fracture modes are commonly observed in high-strength-steel sheet crack propagation specimens and also in many titanium alloys. These are referred to as the transverse mode and the oblique shear mode. There is a transition between these two modes when some factor such as testing temperature or thickness is varied, and in the transition range the fractures are mixed, consisting of a central transverse band with oblique borders on either side. The characterization of these intermediate fractures in terms of the shear fraction was discussed in an earlier section. Examples of these different macroscopic appearances are shown in Figs. 13, 14, and 15. One point should be noted about the full-shear fracture shown in Fig. 14; there is a small triangular region of transverse fracture immediately adjacent to the starting fatigue crack at the right-hand end. This represents transverse crack extension which occurred at some stage of the test, well before the load reached its maximum value. The extent of the transverse crack extension was restricted by plastic deformation starting at the specimen

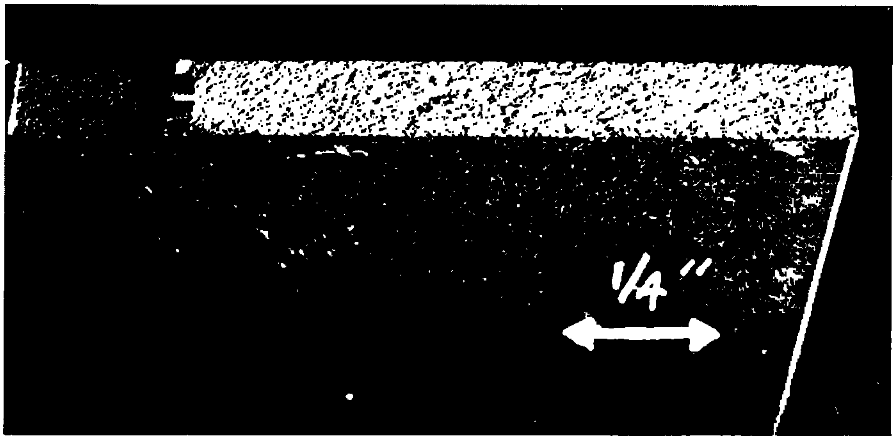

Fig. 13 - Fully transverse macroscopic fracture mode

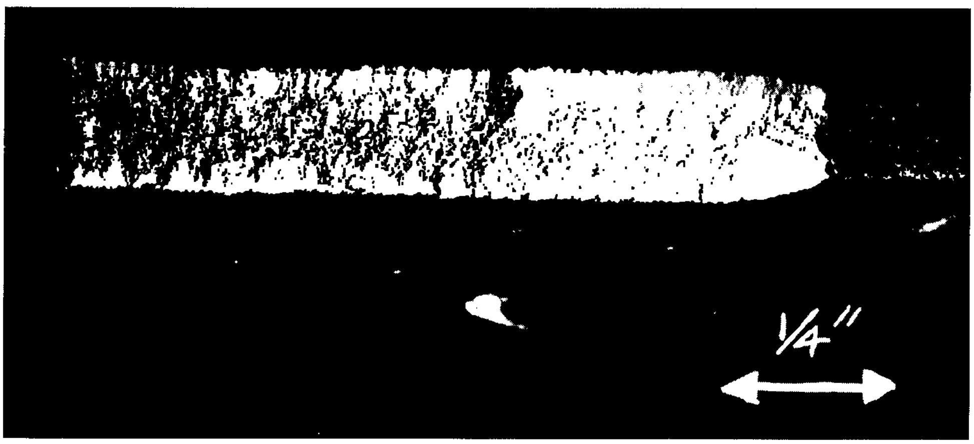

Fig. 14 - Fully oblique (full-shear) macroscopic fracture mode

Fig. 15 - Intermediate type of fracture - central transverse band with oblique borders

surfaces and extending inwards with increasing load. Eventually, the oblique separation surfaces developed and further crack extension was in the oblique mode. These triangular transverse regions invariably occur and correspond to the transverse bands of intermediate type fractures. The size of the region diminishes as some factor favorable to oblique shear separation, such as temperature, is increased. They are of interest here because they provide "samples" of transverse fracture in specimens which separate in full shear.

The mode in which a specimen fractures is thought to depend in part upon the resistance of the material to crack propagation in the transverse mode. The higher this resistance is, the more opportunity will there be for plastic deformation during the initial extension of the transverse crack, and the greater the likelihood that transverse crack extension will be stifled and superseded by oblique separation. Thus the features of the transverse fracture regions are likely to be more significant in relation to crack propagation resistance than those of the oblique regions.

To examine the detailed features of a fracture surface a high-fidelity replica is first obtained for use as an object in the electron microscope. Stereoscopic pairs of photographs of selected parts of the replica are taken and examined at leisure with a suitable viewer. Stereoscopic viewing is indispensable for proper appreciation of the topographical detail, but particular points can be illustrated by reference to suitable selected single photographs, as in this report. Of several replication methods that were tried, the one finally adopted as both convenient and adequate was a two-stage carbon replica method referred to as the Fax-film method. This was compared with a direct carbon method which is generally considered to give best fidelity - unfortunately it involves dissolving away the fracture surface to release the replica. After some practice the Fax-film method gave results which were about as good as those obtained with the direct carbon method. One surface of a piece of 0.005 inch thick cellulose acetate sheet is softened with acetone and pressed against the fracture surface. After it has hardened it can be stripped away from the fracture, which is left undamaged. The replica is then shadowed with palladium at an angle of 45°, followed by coating with carbon deposited at 45° while the specimen is rotated. Finally the cellulose acetate is removed by dissolving in acetone and the fragile carbon replica is mounted on a wire screen for insertion in the object holder of the microscope. The procedure is somewhat similar for replicating polished and etched microstructures except that the initial replica is obtained by dropping a solution of parlodion on the surface and allowing it to dry.

The oblique shear surfaces of the different specimens that were examined were all essentially the same type, illustrated by the example shown in Fig. 16 at a magnification of about 3500. This type of ductile rupture surface has been described previously by Crussard et al., who found it in a wide variety of materials (22). When viewed stereoscopically it is seen to consist largely of a mass of contiguous dimples. By replicating two mating fracture surfaces it can be shown that these dimples are concave on both surfaces - they represent mating depressions, not depressions mating with protuberances. It is considered that separation occurs by nucleation, enlargement, and connecting together of lenticular cavities, each dimple representing a cavity. This kind of appearance can be found in transverse fracture surfaces as well as oblique surfaces, as will be shown later. On oblique surfaces the dimples usually tend to be elongated in the apparent direction of shearing. There is some tendency for the dimples to be elongated preferentially in one direction in Fig. 16, but this is not pronounced. Furthermore, there are some near-vertical steps, seen better in stereo. These features suggest that separation by pulling apart of these surfaces was actually accompanied by relatively little shearing across, and that it is inaccurate to refer to the mode of separation as shear. The idea that the oblique mode of separation is actually an opening mode rather than a true shearing mode is supported by visual observations of the early stages of fracturing of specimens. It is also supported by the fact that some specimens have oblique borders on either side of a transverse band which are normal to one another rather than parallel - a situation difficult to reconcile with any imaginable shearing movement. Of course, these remarks refer to fractures

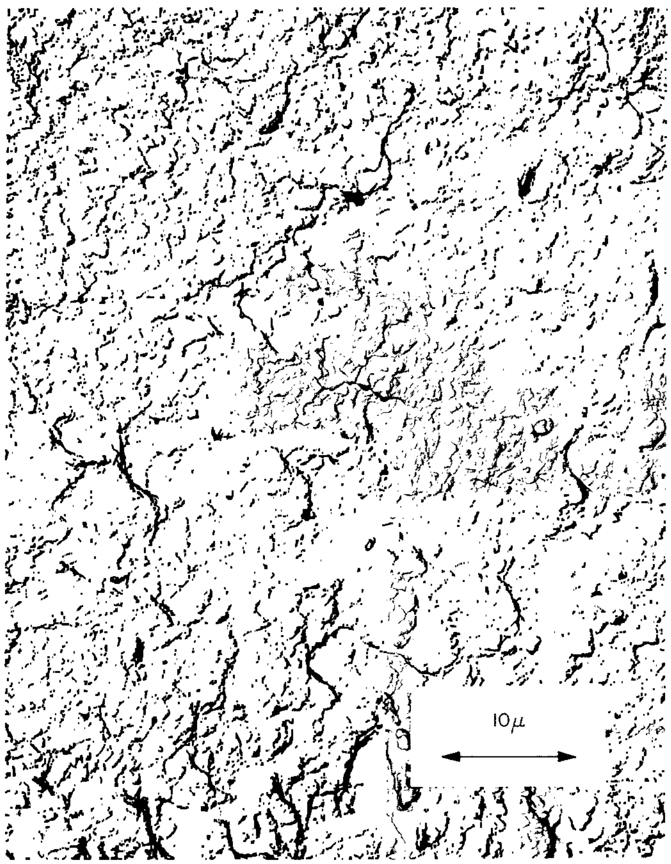

Fig. 16 - Typical oblique fracture surface appearance at a magnification of 3500. (Steel 221 tempered at 400°F and tested at room temperature.) Dimpled ductile rupture. (Reduction in printing approximately 50%.)

found in crack propagation specimens and are not intended to imply that true shearing separations are not to be found under different conditions of loading.

The oblique fracture surfaces of all the other specimens examined were very similar to that shown in Fig. 16, and no significant distinguishing characteristics between different specimens could be found. However, the specimens examined probably did not represent a wide range of variation in fracture toughness for the oblique mode of separation, and it is possible that further study of a wider variety of materials might reveal that certain features varied systematically. Less effort was devoted to the study of the oblique regions than to the transverse regions of the fractures because it was thought likely that the latter would be more rewarding. Essentially the basis for this is that while the effective resistance to crack propagation under a particular set of circumstances depends upon the macroscopic mode of separation, the mode itself is determined partly by the resistance to transverse crack propagation. The latter, being independent of thickness, can be regarded as a basic measure of the fracture toughness of a material.

Three distinct types of fracture were found in the transverse fracture regions of the various specimens examined. Figure 17 shows the dimpled ductile rupture type, similar to that of the oblique fracture regions. The magnification is about 6000, so that the size

Fig. 17 - Dimpled ductile rupture type of appearance in transverse fractured region at a magnification of 6000. (Steel P as-quenched and tested at room temperature.) (Reduction in printing approximately 50%.)

of the dimples in this example is not much larger than those shown in Fig. 16, as might be thought at first glance. Another example of dimpled ductile rupture in a transverse region is shown in Fig. 18, magnification about 8000. This illustrates a special feature which is sometimes found in this type of fracture, more commonly in transverse regions, but also in some oblique regions. Near the middle of the photograph is a nondimpled area having clearly defined parallel striations running nearly horizontally. Several other such areas are to be found elsewhere on the photograph. When viewed stereoscopically it is seen that these are near-vertical steps with the striations running vertically. It would appear that these represent places where cavities on different levels have joined up by true shearing separation, the shearing direction being normal to the gross plane of the fracture.

The second type of transverse fracture is illustrated in Fig. 19 at a magnification of about 3500. This is considered to be intergranular fracture through the prior austenite grain boundary regions, and it was confirmed that the size of the facets was consistent with the austenite grain size of the specimen.

The third type is illustrated in Fig. 20 at a magnification of about 20,000. It is believed that this may be a form of cleavage fracture, although the scale is very fine, and

Fig. 18 - Striated shear steps in dimpled ductile rupture type of transverse fracture at a magnification of 8000. (Steel P tempered at 350°F and tested at room temperature.) (Reduction in printing approximately 50%.)

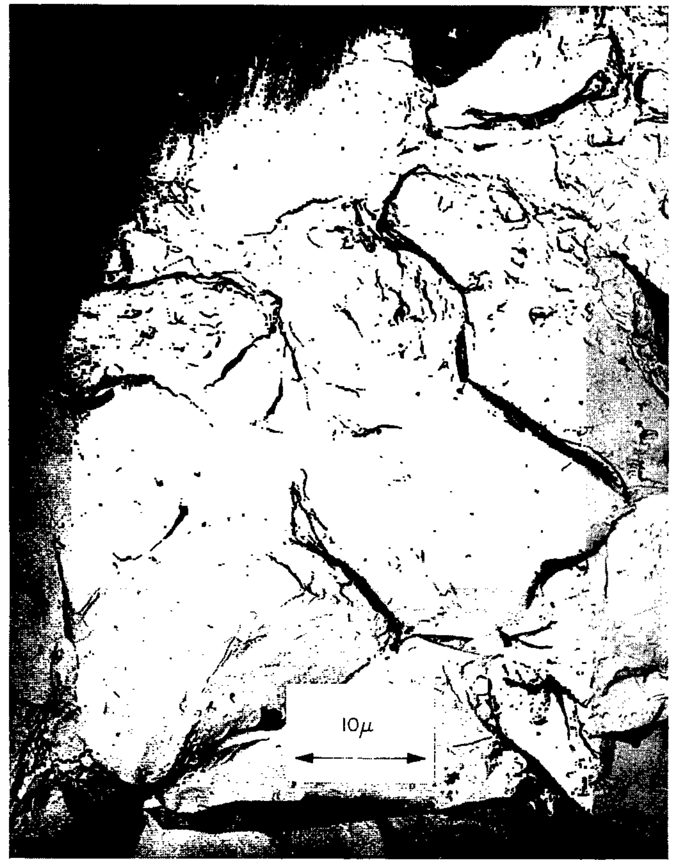

Fig. 19 - Intergranular type of appearance in transverse fracture region at a magnification of 3500. (Steel 219 tempered at 600°F and tested at room temperature.) (Reduction in printing approximately 50%.)

Fig. 20 - Quasi-cleavage type of appearance in transverse fracture region at a magnification of 20,000. (Steel G tempered at 1100°F and tested at 280°F.) (Reduction in printing approximately 50%.)

the well-developed river markings characteristic of cleavage fractures on a course scale are not clearly in evidence. Tentatively this will be referred to as a quasi-cleavage-type fracture. The quasi-cleavage facets in this figure are connected by surfaces which show marked signs of distortion. This, together with the numerous sharp changes of level and orientation of facets, suggests that the associated resistance to crack propagation would not be particularly low. In fact this was taken from a specimen of steel G tested at -280°F, and the crack propagation resistance was very low.

Now that the general types of fracture that were observed have been described, the further discussion will be concerned with studies in connection with certain pronounced trends in the crack propagation data, illustrated with reference to particular steels.

The tendency for fracture strength ratios to be particularly low for specimens tempered at low temperatures was evident in many of the steels. The series of specimens of AMS 6434, which were tested at room temperature, as-quenched, or tempered at 212°, 300°, 350°, and 400°F (Table 9), were studied in this connection. The transverse fracture surfaces of all these specimens were predominantly of the dimpled ductile rupture type, that of the as-quenched specimen being the example shown in Fig. 17. However, although there was no change in type of fracture with tempering temperature, there was a noticeable gradation among the fractures, especially with stereoscopic viewing. The clearest feature of this gradation was an increase in depth with increasing tempering temperature, that is, the differences in level between low and high points of the surface were greater at higher tempering temperatures. In the as-quenched specimen (Fig. 17), the dimples tend to be spread over plateaus, with relatively shallow steps between. There are also a number of small flat areas free of dimples which might be quasi-cleavage facets. In contrast, the specimen tempered at 350°F (Figs. 18 and 21) gives an impression of deep depressions, hills and ridges, with the dimples superimposed, and with occasional deep, vertical, striated steps. Nothing suggestive of a relationship to microstructure was apparent.

Fig. 21 - Transverse fracture region of steel P specimen, tempered at 350°F and tested at room temperature, at a magnification of 6000. Dimpled ductile rupture. (Reduction in printing approximately 50%.)

Specimens of three of the NRL experimental steels, Heats 225 (Ni-Cr-Mo), 227 (Cr-Mo-V), and 228 (Ni-Cr-Mo-V), which were tempered at 400° and at 600°F and tested at room temperature were also examined. These transverse fractures also were all of the dimpled ductile rupture type, with the exception that occasional quasi-cleavage facets occurred in Heat 225, tempered at 400°F. The main difference associated with tempering temperature was that in each case the size of dimples was smaller in the specimen tempered at 600°F than in that tempered at 400°F.

The experimental Heat 219 (Ni-Cr) was selected for study because of the very marked difference in crack propagation resistance at room temperature between specimens tempered at 600°F or less and those tempered at 700°F or higher (Table 44) and also because macroscopic fracture mode transitions occurred between 80° and 200°F in the case of specimens tempered at 600°F, and between 0° and 80°F in the case of specimens tempered at 700°F - a marked effect of tempering temperature upon fracture mode transition temperature. The transverse fracture regions of specimens tested at room temperature were predominantly intergranular in the specimens tempered at 600°F or less, as illustrated by Fig. 19, and were of the dimpled ductile rupture type with occasional intergranular facets in the case of the specimen tempered at 700°F. Thus the poor characteristics of this particular steel were due to some form of intergranular embrittlement which has so far not been identified.

In the case of the specimen tempered at 600°F and tested at 200°F the transverse fracture region consisted largely of dimpled ductile rupture regions intermingled with some intergranular regions, a typical field being illustrated in Fig. 22, which should be compared with Fig. 19. The transverse fracture region of the specimen tempered at 700°F and tested at 0°F was predominantly intergranular. Thus the macroscopic fracture mode transitions for specimens tempered at 600°F or at 700°F were associated with clearly defined changes in the microscopic characteristics of the fractures from intergranular to ductile rupture. This, however, is probably not the general rule.

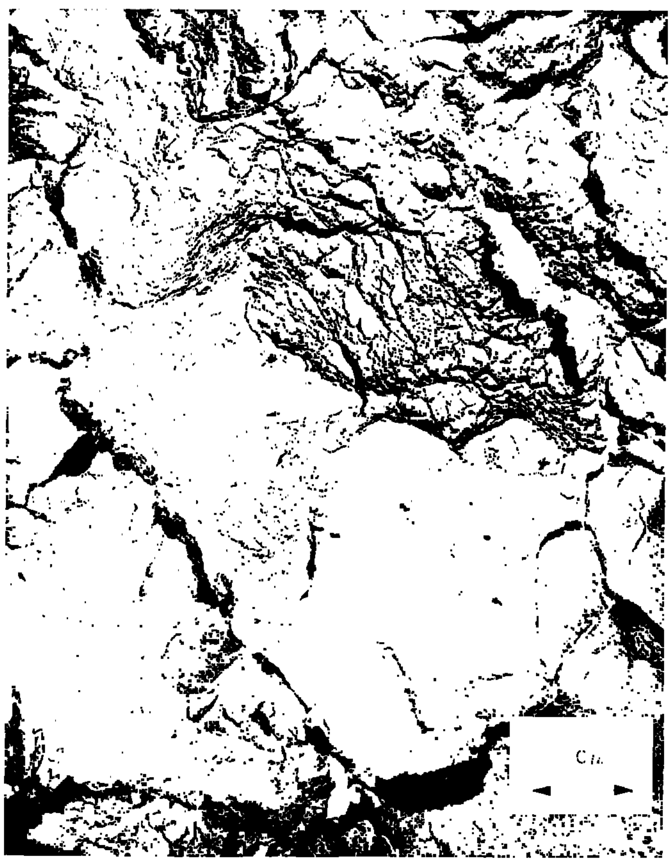

Fig. 22 - Transverse fracture region of steel 219 specimen, tempered at 600°F and tested at 200°F, at a magnification of 2500. Mixed intergranular and dimpled ductile rupture. (Reduction in printing approximately 50%.)

Fig. 23 - Transverse fracture region of steel G specimen, tempered at 1100°F and tested at -280°F, at a magnification of 2500. Quasi-cleavage type. (Reduction in printing approximately 50%.)

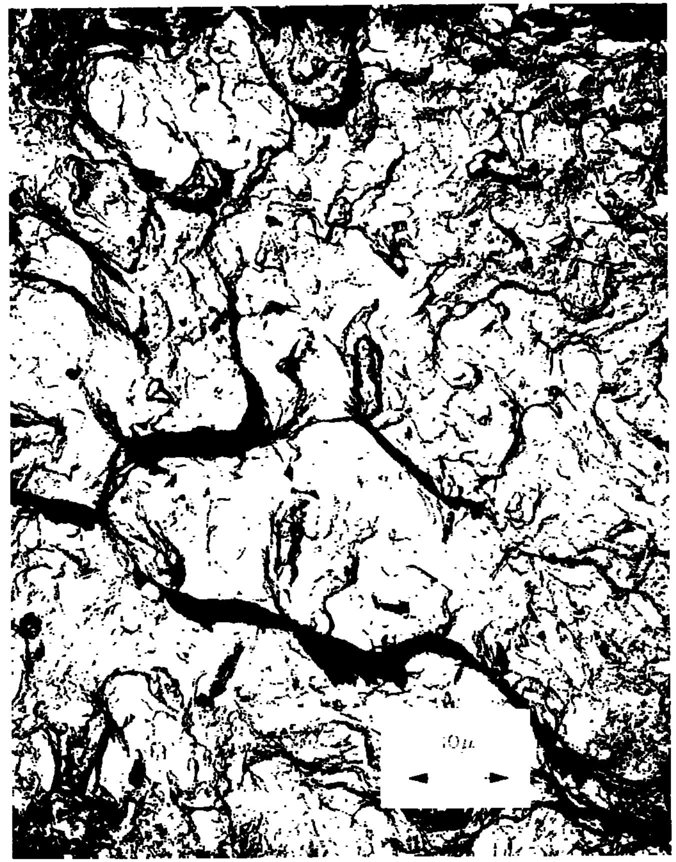

Fig. 24 - Transverse fracture region of steel G specimen, tempered at 1100°F and tested at 80°F, at a magnification of 2500. Predominantly ductile rupture type. (Reduction in printing approximately 50%.)

The change in microscopic characteristics of transverse fractures with testing temperature for a given material is of considerable interest in relation to the temperature dependence of the resistance to transverse crack propagation. Present indications suggest that the change with temperature is gradual and, in general, does not include any transition range of rapid change. The preceding case may be an exception, but the usually observed toughness transitions with temperature are associated with fracture mode transitions rather than with sudden increases in resistance to transverse crack propagation. Specimens of steel G8, tempered for 1 + 1 hours at 1100°F and tested at various temperatures, were examined in this connection (data of Table 19). Figure 23 shows the appearance of the transverse fracture region of a specimen tested at -280°F, and Fig. 24 that of a specimen tested at 80°F, both at 2500 magnification. The shear fractions were 0 and 0.70, respectively. The fracture obtained at the lower testing temperature appears to be composed largely of quasi-cleavage facets - Fig. 20 is from the same specimen at higher magnification. There are also numerous deep rifts representing secondary cracks normal to the general plane of the fracture. On the other hand, much of the room-temperature fracture consists of dimpled ductile rupture surfaces with some striated shear steps. Here and there an occasional quasi-cleavage facet occurs. At intermediate testing temperatures the fractures were between these extremes.

In summary of this section, a variety of different types of fracture topography were observed at high magnification, and three specific characteristic types were distinguished. Much yet remains to be done in the way of comprehension and classification. The evidence suggests that relatively high toughness is associated with the dimpled ductile rupture type of transverse fracture surface, but variations in character within this type are important and not yet understood. Low toughness was clearly associated in one case with the intergranular type of fracture.

CONCLUSIONS

Comparative crack propagation tests and tensile tests were conducted on a number of high-strength sheet or strip materials - nine different commercial steels, twenty-six experimental steels, including six which were selected for ausrolling, and four titanium alloys. With one or two exceptions the nominal thickness of the materials was 1/16 inch. The main objective was to compare the materials on the basis of crack propagation resistance in relation to yield strength, and most of the materials were tested at a number of levels of yield strength. The bulk of the data was obtained at room temperature, with supplementary tests at higher or lower temperatures. From a study of the data given in Tables 1 through 55, and illustrated in Figs. 1 through 12 and 25, the following conclusions may be drawn:

1. It cannot be too strongly emphasized that crack propagation resistance inevitably tends to decrease as strength level is increased. This tendency is most clearly seen in Fig. 25, which is a plot of fracture strength ratio at room temperature versus ratio of yield strength to density showing the best results for each of the four groups of materials. Some materials will have better resistance to crack propagation than others at a given strength level, but the strength level is a general limiting factor which cannot be circumvented. There is nothing novel about this conclusion, it is supported by a great deal of published data and by theoretical considerations, but it is nevertheless often insufficiently appreciated.

2. An outstanding feature of the data in general is that distinctly better results were obtained with titanium alloys than with steels. This is evident from Fig. 25, where the best results for the titanium alloys are contrasted with those for the three groups of steels. For a given fracture strength ratio the limit of strength level is considerably higher for the titanium alloys than for the commercial steels. For instance, at a fracture strength ratio of 0.8, the highest ratio of yield strength to density for a titanium alloy was 0.925,

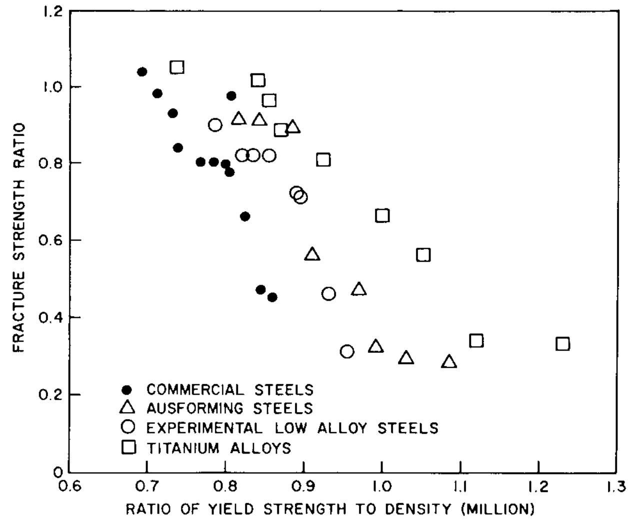

Fig. 25 - General comparison of the best results for the various classes of materials in terms of fracture strength ratio versus ratio of yield strength to density at room temperature

as compared with 0.785 for a commercial steel. There is no absolute basis for selecting a particular level of fracture strength ratio as a criterion, but the value of 0.8 has some merit on the basis of satisfactory use of the steel in question for rocket motor casings (see earlier discussion and also Appendix).

3. Among the titanium alloys the best results were obtained with the 16V-2.5Al alloy identified as R (Tables 23 and 25), as shown in Fig. 5. Both longitudinal and transverse specimens of this alloy were tested, with no appreciable difference in results. Figure 5 also shows that the results for two other titanium alloys were generally better than the best results for commercial steels. The 6Al-4V alloy was tested in the annealed condition only, but nevertheless had a ratio of yield strength to density equal to that of a steel at a yield strength of 230,000 psi, and a corresponding fracture strength ratio better than for any of the steels.

4. Titanium alloys for rocket motor casings should be heat treated so as to result in a suitable combination of yield strength and crack propagation resistance. The appropriate heat treatment is not necessarily one commonly recommended. For instance, Fig. 6 shows that short aging treatments and double aging treatments applied to the beta titanium alloy may be more satisfactory than aging for 100 hours. Results for the 4Al-3Mo-1V alloy also indicated a substantial effect of aging variables on fracture strength ratio independent of yield strength (Table 22).

5. In general, steel P (AMS 6434) had the best characteristics of the commercial steels. The cold-rolled and aged steel AD had an exceptionally high fracture strength ratio for the longitudinal direction, as indicated in Fig. 2, but the fracture strength ratio in the transverse direction was markedly lower. Higher yield strengths were attained with steels L and C, but the fracture strength ratios were relatively low and it is suspected that the high silicon content of these steels has an embrittling effect.

6. The best experimental steels had appreciably better characteristics than the commercial steels, as shown by Fig. 25. However, it is possible that the difference is due mainly to processing differences. None of the experimental steels represented a radical improvement over the commercial steels, and they were generally inferior to the best

titanium alloys. Thus the data does not suggest any possibility for substantially surmounting the limitations of current low-alloy steels, nor does the data suggest that ausforming has particular merit for obtaining high fracture strength ratios at high yield strength levels.

7. Among the steels which were ausrolled, the 12% Cr steel, Heat 217, was particularly interesting (Tables 34 and 35, and Fig. 8). This was the only steel with a fracture strength ratio comparable to the best titanium alloy - at a ratio of yield strength to density of about 0.88 (Fig. 25). The highest yield strengths among the ausrolled steels were obtained with a high-silicon Ni-Cr-Mo-V composition, Heat 200, Table 29. However, the fracture strength ratios for this steel were lower than for Heat 203, Table 32, which was similar in composition except for much lower silicon content. These results support the suspicion that high silicon reduces crack propagation resistance.

8. Results for eighteen experimental compositions were compared for indications of effects of individual alloying elements on crack propagation resistance of quenched and tempered low-alloy steels. With one or two exceptions, the fracture strength ratios at room temperature for specimens tempered at 600°F or higher were fairly consistently related to yield strength, decreasing with increasing yield strength. Lower fracture strength ratios were usually obtained for tempering temperatures of 400° or 500°F (see Fig. 11). The carbide stabilizing elements V, Mo, Cr, and Mn appeared to be beneficial in the quantities employed, partly because they contributed to raising the levels of yield strength resulting from tempering at 600°F or higher.

9. The effect of about 0.2% vanadium was the most pronounced of any of the alloying elements. This was associated with a markedly smaller prior austenite grain size in the vanadium-containing steels than in those without vanadium, except for a Ni-Cr-Mo steel which had grain size and fracture strength ratios comparable with the vanadium-containing steels.

10. There was no evidence of any beneficial effect of either nickel or cobalt in the quantities employed. There was some indication of a detrimental effect of cobalt similar to that suspected of silicon.

11. The data regarding effects of composition are very limited in that only one level of each alloying element was considered (1% of Ni, Co, or Cr, 0.5% of Mo, and 0.2% of V). Thus the inferences from the data are indicative rather than conclusive and require further investigation. However, the existing data do not encourage the hope that low-alloy steels with crack propagation resistance markedly exceeding that of currently available steels will be developed.

12. Studies of the fractures of low-alloy steels at high magnifications with the electron microscope have revealed that at least three general types of fracture may be distinguished: (a) dimpled ductile rupture, (b) intergranular, through prior austenite grain boundaries, and (c) quasi-cleavage. Oblique fracture surfaces were always of the first type, whereas transverse fracture surfaces were of all three types or mixtures of two types. High fracture strength ratios are associated with an oblique (full-shear) macroscopic fracture mode, but there is always a small initial transverse fracture triangle. Generally, comparatively high crack propagation resistance is associated with the dimpled ductile rupture type of fracture in the transverse region. In one case, low crack propagation resistance was clearly associated with the intergranular type of transverse fracture. Further studies are needed to develop understanding of the fracture topographies and their relationships to crack propagation resistance.

13. Results obtained for 20%Ni-Ti-Cb-Al and 25%Ni-Ti-Cb-Al steels did not indicate that these compositions have good crack propagation resistance at high yield strengths. For recommended heat treatments, the yield strengths at room temperature were 254,000 psi and 207,000 psi, respectively, and the fracture strength ratios were 0.26 and 0.50.

Full-shear temperatures were above room temperature. Since these were laboratory heats the results may not be representative of commercial sheet material and are therefore not regarded as conclusive.

14. Laboratory heats of 18.5%Ni-Co-Mo-Ti steels were produced, but specimens were not available for testing before the conclusion of the contract. Results for these steels will be reported separately.

ACKNOWLEDGMENTS

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APPENDIX

RELATIONSHIP OF NOMINAL NET FRACTURE STRENGTH TO THE FRACTURE MECHANICS CRACK TOUGHNESS PARAMETER K_c

The stress intensity factor K for a crack length $2a$ in a center-crack specimen of width W and yield strength F_{ty}, carrying a gross stress σ, according to Ref. 3, Eq. 16 is:

$$K = \sigma \sqrt{W \tan\left(\frac{\pi a}{W} + \frac{K^2}{2WF_{ty}^2}\right)}. \tag{A1}$$

Thus the net section stress can be expressed in terms of K, W, a, and F_{ty}:

$$\sigma_n = \frac{W\sigma}{(W - 2a)}$$

$$= \frac{K}{\left(1 - \frac{2a}{W}\right)\sqrt{W \tan\left(\frac{\pi a}{W} + \frac{K^2}{2WF_{ty}^2}\right)}}.$$

It is convenient to consider the ratio of the net section stress to the yield strength, which is a function of the width W (which is constant for a given specimen design) and of the two variables K/F_{ty} and a/W:

$$\frac{\sigma_n}{F_{ty}} = \frac{K}{F_{ty}} \frac{1}{\left(1 - \frac{2a}{W}\right)\sqrt{W \tan\left(\frac{\pi a}{W} + \frac{K^2}{2WF_{ty}^2}\right)}}. \tag{A2}$$

Figure A1 shows the dependence of σ_n/F_{ty} on $2a/W$ in the range of interest, 0.25 to 0.5, for a series of values of K/F_{ty} and for $W = 1.5$ inches. It is apparent that σ_n/F_{ty} is almost independent of $2a/W$ in this range and, thus, to a good approximation, can be regarded as a function of K/F_{ty} only.

At fracture the nominal net fracture strength $F_n = \sigma W/(W - 2a_o)$ is less than or equal to the true net section stress. It follows that:

$$\frac{K_c}{F_{ty}} \geq \frac{F_n}{F_{ty}} \left(1 - \frac{2a}{W}\right) W \tan\left(\frac{\pi a}{W} + \frac{K_c^2}{2WF_{ty}^2}\right). \tag{A3}$$

The right-hand side of this inequality is plotted in Fig. A2 versus F_n/F_{ty} for $W = 1.5$ and $2a = 0.5$ to 0.6. From this plot a lower bound of the value of K_c/F_{ty} can be obtained when F_n/F_{ty} is known.

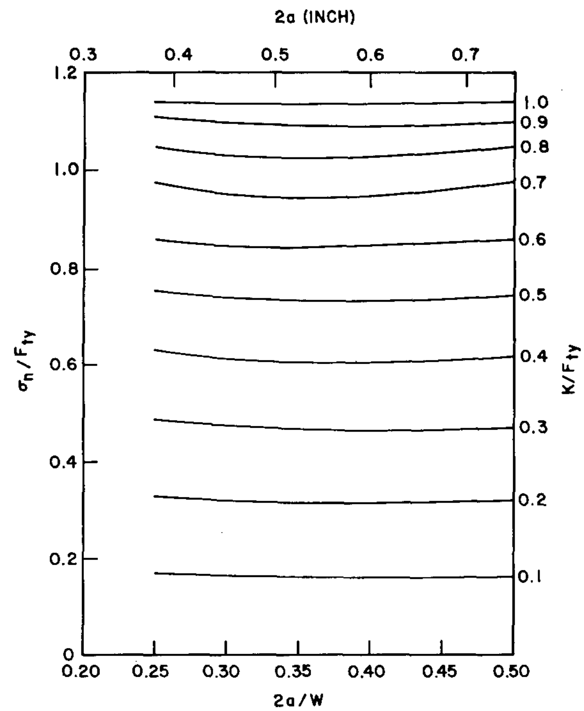

Fig. A1 - Variation of ratio of net section stress to yield strength (σ_n/F_{ty}) with ratio of crack length to specimen width ($2a/W$). The former is nearly independent of the latter in the range shown.

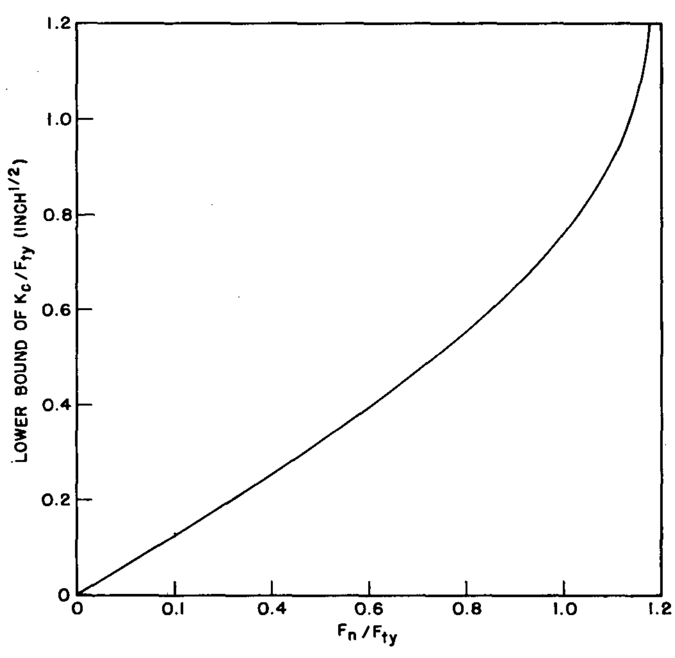

Fig. A2 - Relationship between K_c/F_{ty} and F_n/F_{ty} for specimens 1.5 inches wide with crack length of 0.5 inch

TABLE 1

COMPOSITIONS AND THICKNESSES OF COMMERCIAL STEELS

Ident.	Heat Type*	Gage (Inch)	COMPOSITION PERCENT										
			C	Mn	P	S	Si	Ni	Cr	Mo	V	Al	N
C2	EF	0.07	0.43	0.80	0.007	0.006	1.75	1.81	0.92	0.32	0.13		
C5	CEVM	0.07	0.44	0.65	0.007	0.006	1.74	1.82	0.98	0.36	0.15		
L2	CEVM	0.07	0.42	0.73	0.008	0.008	1.60	-	1.99	0.46	0.10		
P11	CEVM	0.09	0.39	0.70	0.007	0.008	0.53	1.81	0.56	0.37	0.17		
X2	CEVM	0.06	0.38	0.61	0.007	0.007	0.20	1.10	0.59	1.14	0.10		
G6	CEVM	0.07	0.41	0.35	0.012	0.007	0.91	0.09	4.81	1.42	0.56		
G8	CEVM	0.13	0.42	0.37	0.012	0.007	0.88	0.09	4.79	1.45	0.57		
W2		0.10	0.07	0.50	0.008	0.007	0.30	6.8	14.9	2.37		1.06	
K2		0.07	0.09	0.76	0.018	0.017	0.24	4.11	16.4	2.73		-	0.09
AC1		0.07	0.18	0.73	0.017	0.003	0.37	7.00	14.7	2.85		1.15	
AD1		0.07	0.20	0.85	0.015	0.003	0.23	4.19	14.1	2.90		0.02	

* CEVM - Consumable Electrode Vacuum Melted
EF - Electric Furnace (Open to Atmosphere)

TABLE 2

COMPOSITIONS AND THICKNESSES OF TITANIUM ALLOYS

Ident.	Gage (Inch)	COMPOSITION PERCENT							
		Al	V	Mo	Cr	Fe	C	N	H
Q1	0.06	4.2	0.9	3.1	-	0.10	0.015	0.014	0.009
R1	0.06	2.7	15.7	-	-	0.25	0.011	0.015	0.010
S4	0.06	6.3	4.0	-	-	0.17	0.01	-	-
T1	0.06	3.2	13.8	-	10.6	0.20	0.015	0.020	0.011
T4	0.13	3.2	13.5	-	11.5	0.28	0.02	0.019	0.005

TABLE 3

COMPOSITIONS OF NRL EXPERIMENTAL VACUUM INDUCTION MELTED STEELS

No.	C	Mn	P	S	Si	Ni	Cr	Mo	V	Co	Ti	Cb	Al
STEELS FOR AUSROLLING													
200	0.42	1.14	0.004	0.008	1.68	1.14	2.28	0.37	0.35				
201	0.38	0.57	0.004	0.013	0.33	0.09	5.00	0.04	0.02				
202	0.35	0.51	0.005	0.008	0.33	4.20	1.19	0.38	0.05				
203	0.34	0.99	0.006	0.008	0.22	1.06	2.59	0.32	0.31				
204	0.40	0.77	0.002	0.005	0.35	0.69	1.09	1.08	0.08				
217	0.20	0.12	0.007	0.018	0.01	0.07	13.0	0.06	0.01				
LOW ALLOY STEELS													
207	0.41	0.00	0.006	0.005	0.01	0.05	0.09	0.01	0.01	0.01			
208	0.45	0.02	0.005	0.004	1.52	0.05	0.09	0.01	0.01	0.01			
209	0.42	0.97	0.008	0.006	0.01	0.07	0.07	0.07	0.02	0.01			
213	0.41	0.01	0.002	0.008	0.01	2.04	0.12	0.01	0.03	0.04			
214	0.43	0.01	0.004	0.005	0.02	0.07	0.10	0.02	0.02	2.15			
215	0.41	0.00	0.007	0.005	0.01	0.05	0.05	0.56	0.02	0.01			
216	0.43	0.01	0.004	0.004	0.02	0.06	0.13	0.01	0.10	0.03			
219	0.42	1.02	0.004	0.004	0.50	0.99	1.04	0.01	0.01	0.06			
220	0.43	1.03	0.003	0.006	0.50	0.05	1.00	0.48	0.02	0.03			
221	0.41	0.52	0.002	0.005	0.54	0.06	0.87	0.03	0.19	0.03			
222	0.41	0.52	0.002	0.007	0.53	0.07	0.87	0.01	0.02	0.78			
223	0.42	0.54	0.005	0.006	0.53	0.99	0.03	0.49	0.01	0.03			
224	0.43	0.58	0.007	0.004	0.52	0.07	0.01	0.54	0.02	0.86			
225	0.43	1.03	0.003	0.006	0.59	1.03	0.98	0.59	0.02	0.01			
226	0.43	1.03	0.003	0.006	0.54	0.07	0.98	0.63	0.02	0.99			
227	0.45	0.99	0.004	0.006	0.54	0.05	1.03	0.53	0.18	0.02			
228	0.45	1.03	0.004	0.006	0.54	1.00	1.02	0.53	0.18	0.02			
233	0.48	1.02	0.002	0.003	0.54	1.03	0.99	0.52	0.19	0.00			
234	0.43	0.02	0.004	0.008	0.02	0.08	1.95	0.01	0.02	0.03			
235	0.40	0.96	0.001	0.006	0.50	1.00	1.04	0.58	0.18	0.02			
HIGH NICKEL STEELS													
240	0.04	0.03	0.005	0.007	0.02	19.5	0.01	0.00	-	-	1.60	0.32	0.2
241	0.03	0.01	0.005	0.005	0.02	24.3	0.01	0.00	-	-	1.60	0.52	0.2
246	0.02	0.02	0.002	0.007	0.01	18.5	0.06	5.2	-	9.1	0.38		
247	0.04	0.02	0.002	0.007	0.01	18.5	0.03	5.2	-	9.0	0.61		
248	0.03	0.02	0.002	0.005	0.01	18.0	0.06	4.7	-	8.9	0.98		
249	0.03	0.02	0.002	0.007	0.01	18.5	0.06	4.4	-	9.1	0.60		
250	0.04	0.02	0.002	0.005	0.01	18.5	0.03	8.3	-	9.1	0.60		

TABLE 4

TENSILE AND CRACK PROPAGATION TEST RESULTS ON STEEL C2, 0.07 INCH THICK, LONGITUDINAL DIRECTION. HEAT TREATMENT: 1700°F (SALT-BATH) FOR 30 MIN., AIR COOL; 1600°F (SALT-BATH) FOR 30 MIN., OIL QUENCH; DRY ICE FOR 1 HOUR; THEN TEMPERED AS INDICATED

Tempering Treatment	Test Temp. (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 In. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
500°F/1 + 1 hours	200				109.5	1.00	
	80	285.0	235.0	3.3			
	80				104.7	0.60	0.445
	0				104.5	0.45	
600°F/1 + 1 hours	200				114.5	1.00	
	80	281.0	236.0	5.0			
	80				111.8	0.65	0.475
	0				106.5	0.55	
700°F/1 + 1 hours	200				112.5	1.00	
	80	280.0	238.0	5.0			
	80				94.3	0.50	0.395
	0				99.2	0.40	
800°F/1 + 1 hours	200				118.0	0.65	
	80	248.0	210.0	6.0			
	80				91.2	0.30	0.435
	0				81.8	0.20	
900°F/1 + 1 hours	200				132.6	0.50	
	80	220.0	192.0	6.0			
	80				114.4	0.30	0.595
	0				84.2	0.10	
1000°F/1 +1 hours	200				141.0	0.40	
	80	217.0	195.5	7.0			
	80				105.0	0.30	0.54
	0				66.0	0.02	

TABLE 5

TENSILE AND CRACK PROPAGATION TEST RESULTS ON STEEL C5, 0.07 INCH THICK, LONGITUDINAL DIRECTION. HEAT-TREATMENT: 1700°F (SALT-BATH) FOR 30 MIN. AIR COOL; 1600°F (SALT-BATH) FOR 30 MIN. OIL QUENCH; DRY ICE FOR 1 HOUR; THEN TEMPERED AS INDICATED

TEMPERING TREATMENT	Test Temp. (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 In. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
500°F/1 + 1 hours	79	280.0	233.0	6.0			
	80				104.4	0.65	0.45
600°F/1 + 1 hours	79	278.0	234.5	5.5			
	83				107.0	0.60	0.455
700°F/1 + 1 hours	80	280.0	238.0	5.5			
	80				92.0	0.45	0.385
800°F/1 + 1 hours	83	248.0	211.0	7.0			
	83				82.8	0.30	0.39
900°F/1 + 1 hours	83	238.0	210.0	6.5			
	84				101.3	0.30	0.48
1000°F/1 + 1 hours	84	223.0	204.0	8.0			
	79				117.0	0.30	0.575

TABLE 6

TENSILE AND CRACK PROPAGATION TEST RESULTS ON STEEL C5 IN AUSTEMPERED CONDITION, 0.07 INCH THICK, LONGITUDINAL DIRECTION. HEAT TREATMENT: 1700°F (SALT BATH) FOR 30 MINUTES, AIR COOL; 1600°F (SALT BATH) FOR 30 MINUTES, QUENCHED INTO SALT BATH AT 600°F, HELD ONE HOUR, AIR COOLED.

Test Temp (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
400	249.0	144.5	24.0			
400				186.0	1.00	1.29
240	227.0	153.5	10.5			
240				197.5	1.00	1.28
160				203.0	1.00	1.30
80	225.0	157.0	11.5			
80				199.0	1.00	1.27
80				198.0	1.00	1.27
40				187.0	1.00	1.18
0				170.0	1.00	1.07
-20				162.3	1.00	1.01
-40				152.5	0.50	0.955
-80	242.0	160.5	15.0			
-80				131.5	0.30	0.82
-200				71.5	0.15	0.43
-280	267.5	190.0	5.0			
-280				49.5	0	0.26

TABLE 7

TENSILE AND CRACK PROPAGATION TEST RESULTS ON STEEL C5 TEMPERED AT 600°F AS FUNCTION OF TESTING TEMPERATURE, 0.07 INCH THICK, LONGITUDINAL DIRECTION. HEAT TREATMENT: 1700°F (SALT BATH) FOR 30 MINUTES, AIR COOL; 1600°F (SALT BATH) FOR 30 MINUTES, OIL QUENCH; TEMPERED TWICE AT 600°F FOR 1 + 1 HOURS.

Test Temp (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
400	289.0	206.0	12.0			
400				111.0	1.00	0.54
240	292.0	221.5	5.0			
240				126.5	1.00	0.57
160				131.0	1.00	0.57
120				129.5	1.00	0.55
100				111.7	0.55	0.47
80	293.0	240.0	5.5			
80				106.5	0.55	0.445
0				120.0	0.50	0.48
-80	300.0	260.0				
-80				99.5	0.30	0.38
-200				69.0	0.20	0.24
-280	340.0	302.0	FOGL			
-280				39.0	0.10	0.13

FOGL = Fractured Outside Gage Length

TABLE 8

TENSILE AND CRACK PROPAGATION TEST RESULTS ON STEEL L2, 0.07 INCH THICK, LONGITUDINAL DIRECTION. HEAT TREATMENT: 1750°F (SALT-BATH) FOR 30 MIN., AIR COOL; THEN TEMPERED AS INDICATED. ONE SET OF SPECIMENS AUSTEMPERED BY TRANSFERING FROM 1750°F SALT-BATH TO 550°F FOR 1 HOUR, AIR COOL; (NOT TEMPERED)

Heat Treatment	Test Temp. (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. On 2 In. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
Air Hardened and Tempered at 400°F/1 hour	200				127.0	1.00	
	80	293.0	207.0	6.5			
	80				128.0	1.00	0.62
	0				118.5	0.45	
Air Hardened and Tempered at 500°F/1 hour	200				143.6	1.00	
	80	284.0	226.0	6.5			
	80				145.0	1.00	0.64
	0				132.0	0.55	
Air Hardened and Tempered at 600°F/1 hour	200				161.0	1.00	
	80	274.0	220.0	6.5			
	80				160.0	1.00	0.73
	0				141.0	0.60	
Air Hardened and Tempered at 700°F/1 hour	200				160.5	1.00	
	80	277.0	231.0	6.5			
	80				152.0	1.00	0.66
	0				133.0	0.40	
Air Hardened and Tempered at 800°F/1 hour	200				159.0	1.00	
	80	256.0	206.0	10.0			
	80				139.0	0.50	0.675
	0				115.5	0.35	
Air Hardened and Tempered at 900°F/1 hour	200				156.5	1.00	
	80	243.0	178.0	8.5			
	80				143.0	0.50	0.80
	0				105.0	0.20	
Austempered 550°F/1 hour	200				178.5	1.00	
	80	256.0	173.5	10.0			
	80				158.0	1.00	0.91
	0				129.5	0.30	

TABLE 9

TENSILE AND CRACK PROPAGATION TEST RESULTS ON STEEL P11, 0.09 INCH THICK, LONGITUDINAL DIRECTION. HEAT TREATMENT: 1575°F IN ARGON FOR 40 MINUTES, OIL QUENCHED; THEN TEMPERED AS INDICATED

Tempering Treatment	Test Temp. (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
As Quenched	80	306.5	194.0	7.5			
	80				97.0	0.35	0.50
	80				93.5	0.30	0.48
212°F/2 hours	200				107.0	0.55	
	80	303.0	194.5	8.0			
	80				113.5	0.45	0.585
	80				118.0	0.45	0.605
300°F/2 hours	200				127.0	1.00	
	80	297.0	211.0	8.0			
	80				144.0	0.80	0.68
	80				134.0	0.65	0.635
350°F/2 hours	200				150.0	1.00	
	80	283.0	221.0	8.0			
	80				161.0	1.00	0.73
400°F/2 hours	80	275.0	224.0	7.5			
	80				177.0	1.00	0.79
	80				177.0	1.00	0.79
	0				176.0	0.80	
500°F/2 hours	80	266.0	226.0	7.0			
	80				172.0	1.00	0.76
	80				177.0	1.00	0.785
	0				173.0	0.85	
600°F/2 hours	80	249.0	219.0	7.0			
	80				172.0	1.00	0.785
	80				177.6	1.00	0.81
	0				160.0	0.85	
700°F/2 hours	80	231.0	207.0	6.0			
	80				172.5	1.00	0.835
	0				177.5	0.95	
800°F/2 hours	80	208.5	192.5	8.5			
	80				179.0	1.00	0.93
	80				176.5	1.00	0.92
	0				182.0	1.00	
900°F/2 hours	80	196.0	184.5	7.5			
	80				173.0	1.00	0.94
	80				177.0	1.00	0.96
	0				185.5	1.00	
1000°F/2 hours	80	185.0	174.0	8.5			
	80				177.0	1.00	1.02
	80				175.0	1.00	1.01
	0				180.0	1.00	

TABLE 10

TENSILE AND CRACK PROPAGATION TEST RESULTS ON STEEL P11, 0.09 INCH THICK, TRANSVERSE DIRECTION. HEAT TREATMENT: 1575°F IN ARGON FOR 40 MINUTES, OIL QUENCHED; THEN TEMPERED AS INDICATED

Tempering Treatment	Test Temp. (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
As Quenched	200				75.5	0.55	
	80	317.0	202.0	6.5			
	80				105.0	0.35	0.52
	80				88.2	0.30	0.435
212°F/2 hours	200				78.5	0.40	
	80	323.0	210.0	7.0			
	80				83.4	0.20	0.40
	80				83.0	0.20	0.395
300°F/2 hours	200				131.0	1.00	
	80	307.0	234.0	7.0.			
	80				137.5	0.50	0.59
	80				122.0	0.55	0.52
400°F/2 hours	80	278.0	226.0	7.0			
	80				155.0	0.90	0.72
	80				164.0	1.00	
	0				166.0	0.95	
500°F/2 hours	80	267.0	225.0	6.0			
	80				170.0	1.00	0.755
	80				166.0	1.00	
	0				164.0	0.95	
600°F/2 hours	80	251.0	218.0	6.0			
	80				165.0	1.00	0.755
	80				167.0	1.00	
	0				164.0	0.95	
700°F/2 hours	80	231.0	208.0	6.5			
	80				168.0	1.00	0.81
	80				166.0	1.00	
	0				170.0	1.00	
800°F/2 hours	80	209.0	193.0	6.5			
	80				180.0	1.00	0.935
	80				182.0	1.00	0.945
	0				181.0	1.00	
900°F/2 hours	80	196.4	184.0	7.5			
	80				178.5	1.00	0.97
	80				181.0	1.00	0.985
	0				184.0	1.00	
1000°F/2 hours	80	188.5	176.0	9.0			
	80				181.0	1.00	1.03
	80				179.0	1.00	1.02
	0				184.0	1.00	

TABLE 11

TENSILE AND CRACK PROPAGATION TEST RESULTS ON STEEL X2 0.06 INCH THICK; LONGITUDINAL DIRECTION. HEAT-TREATMENT: 1650°F (SALT-BATH) FOR 45 MIN., AIR COOL; 1550°F (SALT-BATH) FOR 45 MIN., AIR COOL; THEN TEMPERED AS INDICATED

Tempering Treatment	Test Temp. (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
500°F/1 hour	200				161.0	1.00	
	80	244.0	206.0	5.5			
	80				160.0	1.00	0.78
	0				118.0	0.30	
600°F/1 hour	200				162.5	1.00	
	80	245.0	214.0	5.0			
	80				170.0	1.00	0.795
	0				113.5	0.35	
700°F/1 hour	200				171.0	1.00	
	80	229.0	203.0	5.5			
	80				176.5	1.00	0.84
	0				158.0	1.00	
800°F/1 hour	200				170.0	1.00	
	80	215.0	193.5	5.0			
	80				178.4	1.00	0.925
	0				177.0	1.00	
900°F/1 hour	200				173.0	1.00	
	80	204.0	183.5	8.0			
	80				176.0	1.00	0.96
	0				181.0	1.00	
1000°F/1 hour	200				174.0	1.00	
	80	200.0	179.5	8.5			
	80				175.0	1.00	0.975
	0				180.0	1.00	
1100°F/1 hour	200				173.0	1.00	
	80	179.0	162.5	10.0			
	80				175.0	1.00	1.075
	0				180.0	1.00	

TABLE 12

TENSILE AND CRACK PROPAGATION TEST RESULTS ON STEEL G6 0.07 INCH THICK, LONGITUDINAL DIRECTION. HEAT TREATMENT: 1850°F IN ARGON FOR 30 MINUTES, AIR COOLED OR OIL QUENCHED; THEN TEMPERED AS INDICATED

Heat Treatment	Test Temp (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
Air Hardened and Tempered	300				108.0	0.35	
900°F/2 + 2 + 2 hours	200				78.6	0.20	
	80	292.0	188.0	10.0			
	80	294.0	191.0	10.5			
	80				54.8	0.05	0.29
	0				45.5	0.00	
Air Hardened and Tempered	600				156.0	1.00	
950°F/2 hours	400				145.0	0.95	
	200				91.3	0.20	
	80	283.0	191.5	7.7			
	80	280.0	199.5	7.7			
	80				63.4	0.05	0.325
Oil Quenched and Tempered	600				193.0	1.00	
950°F/2 hours	400				161.0	0.95	
	200				100.0	0.25	
	80	271.0	198.0	7.7			
	80	271.0	203.0	8.0			
	80				55.6	0.02	0.275
Air Hardened and Tempered	600				194.0	1.00	
950°F/2 + 2 + 2 hours	400				158.0	0.95	
	200				86.5	0.30	
	80	278.0	215.0	4.3			
	80	285.0	211.0	7.7			
	80				75.5	0.05	0.355
Oil Quenched and Tempered	600				198.0	1.00	
950°F/2 + 2 + 2 hours	400				187.0	0.95	
	200				130.0	0.40	
	80	265.0	205.0	9.0			
	80	271.0	209.0	5.3			
	80				75.0	0.05	0.36
Air Hardened and Tempered	300				214.0	1.00	
1000°F/2 + 2 + 2 hours	200				101.5	0.30	
	80	273.0	220.0	6.5			
	80	270.0	220.0	7.0			
	80				113.0	0.10	0.515
	0				54.5	0.05	
Air Hardened and Tempered	200				217.0	1.00	
1050°F/2 + 2 + 2 hours	140				202.0	1.00	
	80	253.0	205.0	6.5			
	80	255.0	215.0	6.5			
	80				104.0	0.20	0.495
	0				91.0	0.10	
Air Hardened and Tempered	200				195.0	1.00	
1100°F/2 hours	140				128.0	0.33	
	80	262.0	208.0	6.5			
	80	261.0	210.0	7.0			
	80				89.4	0.15	0.425
	0				70.2	0.05	
Air Hardened and Tempered	200				208.0	1.00	
1100°F/2 + 2 hours	140				206.0	1.00	
	80	246.0	205.0	6.0			
	80	247.0	210.0	7.0			
	80				133.5	0.30	0.645
	0				117.5	0.15	
Air Hardened and Tempered	200				211.0	1.00	
1100°F/2 + 2 + 2 hours	80	231.0	197.0	6.0			
	80	229.0	192.0	7.5			
	80				203.0	1.00	1.04
	40				137.0	0.20	
	0				100.0	0.10	
Air Hardened and Tempered	200				174.0	1.00	
1150°F/2 + 2 + 2 hours	80	187.0	155.0	9.0			
	80	189.5	156.0	7.6			
	80				182.0	1.00	1.17
	0				183.0	1.00	

TABLE 13

TENSILE AND CRACK PROPAGATION TEST RESULTS ON STEEL G8 AS A FUNCTION OF THICKNESS, LONGITUDINAL DIRECTION. HEAT TREATMENT: 1850°F IN ARGON FOR 30 MINUTES, AIR COOL; TRIPLE TEMPERED 1050°F FOR 2 + 2 + 2 HOURS.

Thickness (in.)	Test Temp (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
0.129	80	269.0	227.0	5.5			
0.129	80	269.0	229.0	5.0			
0.126	80				111.5	0.10	0.49
0.127	80				121.5	0.10	0.53
0.100	80				112.0	0.20	0.51
0.100	80				112.0	0.20	0.51
0.080	80				112.4	0.20	0.51
0.0795	80				97.2	0.25	
0.060	80	263.0	222.0	9.0			
0.060	80	264.0	223.0	9.0			
0.061	80				130.0	0.40	0.585
0.059	80				127.5	0.45	0.57
0.040	80				193.0	1.00	0.87
0.040	80				197.0	1.00	0.895
0.020	80				171.5	1.00	0.77
0.0195	80				161.0	1.00	0.725

TABLE 14

TENSILE AND CRACK PROPAGATION TEST RESULTS ON STEEL G8 AS A FUNCTION OF TEMPERATURE AND THICKNESS, LONGITUDINAL DIRECTION. HEAT TREATMENT: 1850°F IN ARGON FOR 30 MINUTES, AIR COOL; DOUBLE TEMPERED 1100°F FOR 1 + 1 HOURS.

Thickness (in.)	Test Temp. (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
0.100	400	233.0	191.0	8.5			
	400				210.0	1.00	1.10
	200	243.0	208.0	6.7			
	200				209.0	1.00	1.005
	140				205.0	1.00	0.985
	110				180.0	1.00	0.855
	78	251.0	212.0	9.0			
	78	252.0	211.0	10.0			
	78				162.0	0.35	0.765
	40				125.0	0.25	0.575
	0	259.0	223.0	9.5			
	0				97.0	0.15	0.435
	-100	272.0	230.0	FOGL			
	-100				68.8	0.05	0.30
	-200	272.5	242.0	FOGL			
	-200				46.0	0.00	0.19
	-280	290.0	273.0	FOGL			
	-280				26.1	0.00	0.095
0.050	400				207.0	1.00	1.08
	200				210.0	1.00	1.01
	78				207.0	1.00	0.98
	60				209.0	1.00	0.97
	50				197.0	1.00	0.91
	40				135.6	0.40	0.62
	0				119.0	0.35	0.545
	-100				72.4	0.02	0.31
	-200				44.7	0.00	0.185
	-280				35.1	0.00	0.13

FOGL = Fractured Outside Gage Length

TABLE 15

TENSILE AND CRACK PROPAGATION TEST RESULTS ON STEEL W2 (RH 950 CONDITION) AS A FUNCTION OF TEMPERATURE. 0.10 INCH THICK, LONGITUDINAL DIRECTION. HEAT TREATMENT: 1750°F IN AIR FOR 10 MINUTES, AIR COOL; -100°F FOR 8 HOURS; 950°F FOR 1 HOUR

Test Temp (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
580	202.5	179.0	5.7			
580				162.0	1.00	0.905
400	214.0	191.0	5.5			
400				175.0	1.00	0.915
300				175.0	0.80	0.875
250				179.0	0.50	0.88
200	226.0	208.0	7.5			
200				184.0	0.35	0.885
190				169.0	0.30	0.815
180				132.0	0.30	0.63
160				119.0	0.25	0.57
140				110.0	0.15	0.52
80	234.0	216.0	7.5			
80				68.8	0.05	0.32
0				54.0	0.02	0.24

TABLE 16

TENSILE AND CRACK PROPAGATION TEST RESULTS ON STEEL W2 (TH 1050 CONDITION) AS A FUNCTION OF TEMPERATURE. 0.10 INCH THICK, LONGITUDINAL DIRECTION. HEAT TREATMENT: 1400°F IN AIR FOR 90 MINUTES, AIR COOL; 50°F FOR 30 MINUTES; 1050°F FOR 90 MINUTES

Test Temp (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
580	181.5	172.0	5.5			
580				170.0	1.00	0.99
400	190.0	187.0	5.5			
400				181.0	1.00	0.97
300				186.5	1.00	0.98
200	198.0	192.0	6.0			
200				188.0	0.70	0.98
140				193.5	0.50	0.99
75	206.0	198.5	7.5			
75				195.0	0.45	0.985
40				191.0	0.25	0.945
20				154.2	0.20	0.755
0				112.4	0.15	0.55
-100	220.0	215.0	5.0			
-100				76.8	0.05	0.355
-200				52.2	0.00	0.225
-280	256.0	252.0	5.7			
-280				39.0	0.00	0.155

TABLE 17

TENSILE AND CRACK PROPAGATION TEST RESULTS ON STEEL W2 (TH 1050 CONDITION) AS A FUNCTION OF THICKNESS AND TEMPERATURE. LONGITUDINAL DIRECTION. HEAT TREATMENT: 1400°F IN AIR FOR 90 MINUTES, AIR COOL; 50°F FOR 30 MINUTES; 1050°F FOR 90 MINUTES.

Thickness (in.)	Test Temp (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
0.100	300	197.0	191.0	5.0			
	300				186.0	1.00	0.975
	300				186.5	1.00	0.975
	80	212.0	205.0	6.0			
	80				191.0	0.35	0.93
	80				193.0	0.35	0.94
	-280	270.0	264.0	2.5			
	-280				32.4	0.00	0.12
	-280				34.0	0.00	0.13
0.080	300				184.5	1.00	0.97
	300				187.0	1.00	0.98
	80				193.5	0.50	0.94
	80				198.0	0.35	0.97
	-280				32.4	0.00	0.12
	-280				36.3	0.00	0.135
0.060	300	198.5	193.5	4.5			
	300	199.5	194.5	5.0			
	300				182.0	1.00	0.94
	300				184.0	1.00	0.95
	80	209.0	204.0	5.0			
	80	211.0	206.0	6.0			
	80				196.5	0.45	0.96
	80				186.0	0.50	0.91
	-280	262.0	262.0	FOGL			
	-280	269.0	242.0	6.5			
	-280				32.2	0.00	0.12
	-280				35.9	0.00	0.135
0.040	300				173.0	1.00	0.89
	300				180.0	1.00	0.93
	80				191.0	1.00	0.93
	80				186.6	1.00	0.91
	-280				38.7	0.00	0.145
	-280				36.5	0.00	0.14
0.020	300				174.0	1.00	0.89
	80				185.0	1.00	0.90
	80				171.0	1.00	0.835
	-280				37.8	0.00	0.145
	-280				37.5	0.00	0.145

FOGL = Fractured Outside Gage Length.

TABLE 18

TENSILE AND CRACK PROPAGATION TEST RESULTS ON STEEL K2 (SCT 850 CONDITION) AS A FUNCTION OF TEMPERATURE. 0.07 INCH THICK, LONGITUDINAL DIRECTION. HEAT TREATMENT: 1710°F (SALT BATH) FOR 10 MINUTES, AIR COOL; -100°F FOR 3 HOURS; 850°F FOR 3 HOURS, AIR COOL.

Test Temp (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
600	195.5	140.6	8.0			
600				159.5	1.00	1.13
430	189.5	146.0	9.5			
400				181.0	1.00	1.24
250	197.0	154.5	10.5			
200				178.5	1.00	1.11
150				187.0	1.00	1.12
78	210.0	177.0	16.5			
84				183.0	0.95	1.03
40				184.0	0.90	1.00
20				186.0	0.65	1.00
0				177.0	0.55	0.94
-50				146.0	0.35	0.745
-100	244.0	205.5	FOGL			
-100				119.0	0.25	0.58
-200				82.7	0.20	0.39
-300	283.0	241.0	FOGL			
-300				60.0	0.10	0.25

FOGL = Fractured Outside Gage Length.

TABLE 19

TENSILE AND CRACK PROPAGATION TEST RESULTS ON STEEL ACI, 0.065 INCH THICK. HEAT TREATMENT: 1750°F IN SALT FOR 30 MINUTES, WATER QUENCHED; -100°F FOR 6 HOURS; AGED 950°F FOR 90 MINUTES.

Specimen Orientation	Test Temp. (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elong. On 2 In. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
Longitudinal	80	243.0	226.0	5.0			
	80	244.0	227.0	5.5			
	80				53.8	0.20	0.24
	80				72.2	0.20	0.32
	200				155.4	0.75	
	300				164.0	1.00	
Transverse	80	246.0	230.0	5.0			
	80	245.0	226.0	6.0			
	80				82.0	0.20	0.36
	80				74.2	0.20	0.32
	200				125.0	0.50	
	300				127.0	0.90	

TABLE 20

TENSILE AND CRACK PROPAGATION TEST RESULTS ON STEEL AD1 (50% COLD ROLLED AND AGED CONDITION). 0.07 INCH THICK. TREATMENT: COLD ROLLED 50%; AGED 800°F FOR 3 HOURS.

Specimen Orientation	Test Temp (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
Longitudinal	0				194.0	0.50	
	90	291.0	225.0	18.0			
	90				220.0	1.00	0.98
	90				217.0	1.00	0.965
	200	267.0	247.0	7.5			
	200				231.0	1.00	0.935
Transverse	90	269.0	207.0	FOGL			
	90				147.0	0.50	0.71
	90				146.0	0.50	0.705
	200	278.0	249.0	7.5			
	200				149.0	0.70	0.60
	300				159.0	0.80	

FOGL = Fractured Outside Gage Length.

TABLE 21

MAXIMUM FRACTURE STRENGTH RATIOS OBTAINED FOR COMMERCIAL STEELS AT ROOM TEMPERATURE ACCORDING TO YIELD STRENGTH LEVEL.

	YIELD STRENGTH (KSI) NOT LESS THAN:								
IDENT.	180	190	200	210	220	230	240	250	260
C2		0.595	(0.475)	(0.475)	(0.475)	0.475			
C5			0.575	0.48	(0.455)	0.455	0.445		
L2	(0.73)	(0.73)	(0.73)	(0.73)	0.73	0.66			
P11	0.95	0.925	0.835	0.80	0.79				
P11(T)	0.98	0.94	0.81	0.755	0.755				
X2	0.96	0.925	0.84	0.795					
G6	(1.04)	1.04	0.645	(0.515)	0.515				
W2	(0.985)	0.985	0.935	0.32					
K2									
AC1					0.28				
AC1(T)					0.34				
AD1					0.97				
AD1(T)					0.71				

Notes: (a) Longitudinal test results except where (T) following identification symbol indicates transverse.

(b) A value in parentheses is the same as the value for the next highest yield strength and indicates that no higher value was obtained within the yield strength interval.

TABLE 22

TENSILE AND CRACK PROPAGATION TEST RESULTS ON TITANIUM-4Al-3Mo-1V SHEET Q1, 0.06 INCH THICK, LONGITUDINAL DIRECTION. HEAT TREATMENT: SOLUTION ANNEALED (AS RECEIVED) AND AGED IN VACUUM AS INDICATED.

Ageing Treatment	Test Temp (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
925°F/24 hours	85	192.5	163.0	6.0			
	83				42.6		0.26
	85				44.2		0.27
950°F/24 hours	85	176.0	156.0	7.5		Fracture appearance of this specimen not describable in terms of shear fracture	
	84				43.8		0.28
	86				84.7		0.54
975°F/12 hours	77	189.0	160.5	6.5			
	78				63.0		0.39
	0				50.0		
	-100				54.0		
1000°F/12 hours	77	180.0	156.0	FOGL			
	78				76.3		0.49
	0				56.8		
	-100				63.0		
1000°F/24 hours	86	173.0	153.5	FOGL			
	88				128.0		0.75
	81				113.0		0.65
1025°F/12 hours	80	174.0	153.0	7.5			
	76				85.4		0.49
	0				79.5		
	-100				77.6		
1050°F/12 hours	80	173.5	154.0	FOGL			
	77				98.8		0.575
	0				80.0		
	-100				79.0		
1050°F/24 hours	89	153.0	139.0	FOGL			
	84				135.0		0.97
	84				131.4		0.945
1075°F/12 hours	80	150.5	137.0	8.5			
	77				138.0	1.005	
	0				136.0		
	-100				128.0		
1100°F/24 hours	90	151.0	138.0	FOGL			
	85				117.0		0.85
	86				123.0		0.89

FOGL = Fractured Outside Gage Length

TABLE 23

TENSILE AND CRACK PROPAGATION TEST RESULTS ON TITANIUM-16V-2.5Al SHEET R1, 0.06 INCH THICK, LONGITUDINAL DIRECTION. HEAT TREATMENT: SOLUTION ANNEALED (AS RECEIVED) AND AGED IN VACUUM AS INDICATED.

Ageing Treatment	Test Temp (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 in. (%)	Nom. Net. Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
900°F/4 hours	200				56.2	0.40	
	80	210.0	*	*			
	80				54.3	0.30	0.26
	80				51.0	0.30	0.24
950°F/4 hours	200				78.0	0.70	
	80	217.0	210.0	3.5			
	80				72.8	0.35	0.34
	80				56.8	0.50	0.27
1000°F/4 hours	80	186.0	177.0	6.5			
	80				95.0	1.00	0.535
	80				99.5	1.00	0.56
	0				95.4	0.90	
1020°F/4 hours	80	175.0	165.5	7.0			
	80				96.3	1.00	0.58
	80				98.2	1.00	0.595
	0				95.0	1.00	
1050°F/4 hours	80	168.0	155.4	7.5			
	80				128.6	1.00	0.83
	80				120.0	1.00	0.77
	0				117.5	1.00	
1100°F/4 hours	80	158.0	146.0	8.5			
	80				130.5	1.00	0.895
	80				127.5	1.00	0.875
	0				130.0	1.00	

*Brittle Fracture Below Yield Strength.

TABLE 24

TENSILE AND CRACK PROPAGATION TEST RESULTS ON TITANIUM - 16V-2.5Al SHEET R1 AGED 4 HOURS AT 1020°F AS FUNCTION OF TESTING TEMPERATURE, 0.06 INCH THICK, LONGITUDINAL DIRECTION.

Test Temp (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 in. (%)	Nom. Net. Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
400	155.0	140.0	6.0			
400				128.4	1.00	0.92
200	166.0	153.5	7.0			
200				124.0	1.00	0.81
76	175.0	165.5	7.0			
79				96.3	1.00	0.58
79				98.2	1.00	0.595
0	185.0	177.0	7.0			
0				95.0	1.00	0.535
-100	202.5	194.0	6.0			
-100				88.3	1.00	0.455
-125				91.8	0.90	
-150				92.5	0.90	
-200				90.6	0.50	
-280				76.0	0.40	0.335
-300	224.0	227.0	6.0			

TABLE 25

TENSILE AND CRACK PROPAGATION TEST RESULTS ON TITANIUM-16V-2.5Al SHEET R1, 0.06 INCH THICK, TRANSVERSE DIRECTION. HEAT TREATMENT: SOLUTION ANNEALED (AS RECEIVED) AND AGED AS INDICATED.

Ageing Treatment	Test Temp (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	F_n/F_{ty}
900°F/4 hours	200				63.5	0.40	
	80	145.0	*	*			
	80				54.7	0.20	
	80				54.0	0.20	
950°F/4 hours	200				75.4	0.70	
	80	218.0	206.0	3.0			
	80				69.2	0.60	0.335
	80				67.6	0.50	0.33
1000°F/4 hours	80	187.0	177.4	5.5			
	80				111.0	1.00	0.625
	80				84.0	1.00	0.47
	0				75.0	0.90	
1020°F/4 hours	80	178.5	167.5	7.5			
	80				108.0	1.00	0.645
	80				111.0	1.00	0.665
	0				101.0	1.00	
1050°F/4 hours	80	167.0	155.5	9.0			
	80				123.0	1.00	0.79
	80				120.5	1.00	0.775
	0				107.5	1.00	
1100°F/4 hours	80	158.5	147.0	10.0			
	80				129.5	1.00	0.88
	80				130.0	1.00	0.885
	0				128.0	1.00	

*Brittle Fracture Below Yield Strength.

TABLE 26

TENSILE AND CRACK PROPAGATION TEST RESULTS ON TITANIUM - 6Al-4V SHEET S-4 IN THE MILL-ANNEALED CONDITION AS A FUNCTION OF TESTING TEMPERATURE, 0.06 INCH THICK.

Specimen Orientation	Test Temp (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	F_n/F_{ty}
Longitudinal	400	118.4	100.0	13.5		Fracture appearance of this specimen not describable in terms of shear fracture	
	400				110.3		1.10
	200	134.5	118.0	13.5			
	200				125.0		1.06
	80	146.5	133.0	13.5			
	80				121.4		0.915
	0				134.0		0.91
	-40	169.0	155.0	12.5			
	-100				134.5		0.83
	-150				138.5		0.825
	-160	183.0	170.0	FOGL			
	-200				149.0		0.83
	-240				142.5		0.75
	-280	220.0	196.5	FOGL			
	-280				138.5		0.705
Transverse	400	112.0	107.5	13.5			
	400				111.0		1.03
	240	127.0	122.5	13.5			
	200				130.5		1.06
	80	144.0	133.0	13.5			
	80				133.5		1.00
	0				125.4		0.89
	-100	174.0	174.0	11.5			
	-100				120.0		0.69
	-150				129.0		0.71
	-200				131.0		0.70
	-240				106.5		0.55
	-280	218.0	198.5	10.0			
	-280				89.0		0.45

FOGL = Fractured Outside Gage Length.

TABLE 27

TENSILE AND CRACK PROPAGATION TEST RESULTS ON TITANIUM-13V-11Cr-3Al SHEET T1, 0.06 INCH THICK, LONGITUDINAL DIRECTION. HEAT TREATMENT: SOLUTION ANNEALED (AS RECEIVED) AND AGED IN VACUUM AS INDICATED.

Ageing Treatment (Vacuum)	Test Temp (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
750°F/100 hours	75	197.0	182.0	5.5			
	78				70		0.385
	78				74.5		0.41
	78				69.2		0.38
800°F/100 hours	75	214.0	195.0	5.0			
	74				62.8		0.32
	75				63.6		0.325
	75				64.6		0.335
850°F/100 hours	81	219.0	198.0	5.0			
	83				60.4		0.305
	82				59.5		0.30
900°F/100 hours	82	189.0	178.0	2.5			
	80				71.6		0.40
950°F/100 hours	80	187.6	171.0	3.0			
	83				64.0		0.375
	81				70.6		0.41
	82				62.5		0.365
1000°F/100 hours	75	164.5	147.5	6.0			
	75				75.6		0.515
	76				79.4		0.54
	77				76.3		0.52
1050°F/100 hours	75	163.5	143.5	7.5			
	77				78.0		0.53
	77				83.0		0.565
	77				76.3		0.54
None	82	134.5	129.0	26.5			
	80				135.0		1.045

TABLE 28

TENSILE AND CRACK PROPAGATION TEST RESULTS ON TITANIUM-13V-11Cr-3Al SHEET T4 AS A FUNCTION OF THICKNESS, LONGITUDINAL DIRECTION. HEAT TREATMENT: SOLUTION ANNEALED (AS RECEIVED) AND AGED IN VACUUM FOR 48 HOURS AT 900°F, THEN 30 MINUTES AT 1100°F.

Thickness (in.)	Test Temp (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
0.130	80	182.0	164.0	7.5			
	78				76.0	0.10	0.46
	79				77.8	0.10	0.475
0.100	80	181.0	163.0	8.5			
	79				92.0	0.35	0.56
	80				86.7	0.30	0.53
0.080	81	176.5	159.5	8.5			
	78				94.5	0.30	0.59
	80				82.6	0.30	0.52
0.060	81	179.5	162.0	8.5			
	78				101.6	0.50	0.625
	81				99.5	0.40	0.61
0.040	81	170.5	154.0	8.5			
	81				92.8	0.80	0.60
	78				103.5	0.60	0.67
0.020	80	170.0	165.0	6.7			
	81				103.0	1.00	0.625
	80				106.5	1.00	0.645

TABLE 29

TENSILE AND CRACK PROPAGATION TEST RESULTS ON NRL HEAT 200. AUSROLLED AND TEMPERED COMPARED WITH OIL HARDENED AND TEMPERED. (A) AUSTENITIZED 1600°F (ARGON), AUSROLLED 75% AT 850-1050°F, OIL QUENCHED AND TEMPERED (B) HOT ROLLED 1800°F, NORMALIZED, AUSTENITIZED 1600°F (ARGON), OIL QUENCHED, AND TEMPERED.

Condition	Test Temp (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
(a) Ausrolled & Tempered	80	324.0	259.0	FOGL			
400°F/1 hour	80				106.0		0.41
Ausrolled & Tempered	80	317.0	288.0	FOGL			
600°F/1 hour	80				83.0		0.29
Ausrolled & Tempered	80	310.0	302.0	FOGL			
800°F/1 hour	80				86.5		0.285
Ausrolled & Tempered	80	314.0	268.0	FOGL			
900°F/1 hour	80				107.5		0.40
(b) Quenched & Tempered	80	300.0	258.0	FOGL			
400°F/1 hour	80				108.0	0.75	0.42
Quenched & Tempered	80	296.0	268.0	FOGL			
600°F/1 hour	80				116.0	0.70	0.435
Quenched & Tempered	80	280.0	257.0	FOGL			
800°F/1 hour	80				114.0	0.80	0.445
Quenched & Tempered	80	258.0	222.0	FOGL			
900°F/1 hour	80				120.0	0.50	0.54

FOGL = Fractured Outside Gage Length

TABLE 30

TENSILE AND CRACK PROPAGATION TEST RESULTS ON NRL HEAT 201. AUSROLLED AND TEMPERED COMPARED WITH OIL HARDENED AND TEMPERED. (a) AUSTENITIZED 1600°F (ARGON), AUSROLLED 75% AT 850-1050°F, OIL QUENCHED AND TEMPERED. (b) AUSROLLED, RE-AUSTENITIZED 1600°F (ARGON), OIL QUENCHED AND TEMPERED.

Condition	Test Temp. (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. On 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
(a) Ausrolled & Tempered 300°F/1 hour	80	260.0	235.0	5.5			
	80				142.0		0.605
Ausrolled & Tempered 400°F/1 hour	80	259.0	236.0	5.0			
	80				123.0		0.52
Ausrolled & Tempered 600°F/1 hour	80	242.0	240.0	FOGL			
	80				139.0		0.58
Ausrolled & Tempered 800°F/1 hour	80	250.0	245.0	5.0			
	80				154.0		0.63
Ausrolled & Tempered 900°F/1 hour	80	230.0	219.0	FOGL			
	80				147.0		0.67
(b) Re-Austenitized & Tempered 300°F/1 hour	80	264.0	215.0	5.0			
	80				104.6	0.30	0.485
Re-Austenitized & Tempered 400°F/1 hour	80	266.0	211.0	6.0			
	80				179.0	1.00	0.85
Re-Austenitized & Tempered 600°F/1 hour	80	245.0	201.0	6.0			
	80				187.0	1.00	0.93
Re-Austenitized & Tempered 800°F/1 hour	80	226.0	195.0	6.0			
	80				191.0	1.00	0.98
(b) Re-Austenitized & Tempered 900°F/1 hour	80	213.0	182.5	7.0			
	80				187.0	1.00	1.02

FOGL = Fractured Outside Gage Length

TABLE 31

TENSILE AND CRACK PROPAGATION TEST RESULTS ON NRL HEAT 202. AUSROLLED AND TEMPERED COMPARED WITH OIL HARDENED AND TEMPERED. (a) AUSTENITIZED 1600°F (ARGON), AUSROLLED 75% AT 850-1050°F, OIL QUENCHED AND TEMPERED. (b) AUSROLLED, RE-AUSTENITIZED 1600°F (ARGON), OIL QUENCHED AND TEMPERED.

Condition	Test Temp. (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elong. On 2 In. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
(a) Ausrolled & Tempered 300°F/1 hour	80 80	227.0	198.0	FOGL	166.0	1.00	0.84
Ausrolled & Tempered 400°F/1 hour	80 80	231.0	200.0	FOGL	180.0	1.00	0.90
Ausrolled & Tempered 600°F/1 hour	80 80	217.0	208.0	FOGL	185.0	1.00	0.89
Ausrolled & Tempered 800°F/1 hour	80 80	223.0	199.0	11.0	178.0	1.00	0.895
Ausrolled & Tempered 900°F/1 hour	80 80	212.0	192.5	FOGL	181.0	1.00	0.94
(b) Re-Austenitized & Tempered 300°F/1 hour	80 80	273.0	204.0	5.0	159.0	1.00	0.78
Re-Austenitized & Tempered 400°F/1 hour	80 80	247.0	198.0	FOGL	190.0	1.00	0.96
Re-Austenitized & Tempered 600°F/1 hour	80 80	229.0	191.5	FOGL	180.0	1.00	0.94
Re-Austenitized & Tempered 800°F/1 hour	80 80	208.0	176.0	7.0	172.0	1.00	0.98
Re-Austenitized & Tempered 900°F/1 hour	80 80	193.0	167.0	7.0	170.0	1.00	1.02

FOGL = Fractured Outside Gage Length.

TABLE 32

TENSILE AND CRACK PROPAGATION TEST RESULTS ON NRL HEAT 203, AUSROLLED AND TEMPERED COMPARED WITH OIL HARDENED AND TEMPERED. (a) AUSTENITIZED 1600°F (ARGON), AUSROLLED 75% AT 850-1050°F, OIL QUENCHED AND TEMPERED. (b) AUSROLLED, RE-AUSTENITIZED 1600°F (ARGON), OIL QUENCHED AND TEMPERED.

Condition	Test Temp (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
(a) Ausrolled & Tempered 300°F/1 hour	80 80	266.0	237.0	FOGL	131.0		0.555
Ausrolled & Tempered 400°F/1 hour	80 80	273.0	222.0	11.0	121.0		0.545
Ausrolled & Tempered 600°F/1 hour	80 80	272.0	271.0	5.5	127.5		0.47
Ausrolled & Tempered 800°F/1 hour	80 80	264.0	255.0	10.0	142.0		0.56
Ausrolled & Tempered 900°F/1 hour	80 80	258.0	236.0	10.0	145.0		0.62
(b) Re-Austenitized & Tempered 300°F/1 hour	80 80	279.0	229.0	6.0	90.2	0.60	0.39
Re-Austenitized & Tempered 400°F/1 hour	80 80	275.0	234.0	FOGL	179.0	1.00	0.765
Re-Austenitized & Tempered 600°F/1 hour	80 80	248.0	224.0	FOGL	196.0	1.00	0.875
Re-Austenitized & Tempered 800°F/1 hour	80 80	234.0	215.0	6.0	203.0	1.00	0.945
Re-Austenitized & Tempered 900°F/1 hour	80 80	217.0	198.0	6.0	196.5	1.00	0.99

FOGL = Fractured Outside Gage Length.

TABLE 33

TENSILE AND CRACK PROPAGATION TEST RESULTS ON NRL HEAT 204. AUSROLLED AND TEMPERED COMPARED WITH OIL HARDENED AND TEMPERED. (a) AUSTENITIZED 1600°F (ARGON), AUSROLLED 75% AT 850-1050°F, OIL QUENCHED AND TEMPERED. (b) AUSROLLED, RE-AUSTENITIZED 1600°F (ARGON), OIL QUENCHED AND TEMPERED.

Condition	Test Temp (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
(a) Ausrolled & Tempered 300°F/1 hour	80	274.0	270.0	FOGL			
	80				78.0	0.20	0.29
Ausrolled & Tempered 400°F/1 hour	80	266.0	265.0	1.5			
	80				88.3	0.20	0.33
Ausrolled & Tempered 600°F/1 hour	80	270.0	266.0	FOGL			
	80				70.4	0.25	0.265
Ausrolled & Tempered 800°F/1 hour	80	280.0	277.0	3.0			
	80				89.4	0.25	0.32
Ausrolled & Tempered 900°F/1 hour	80	261.0	258.0	4.5			
	80				100.5	0.35	0.385
(b) Re-Austenitized & Tempered 300°F/1 hour	80	285.0	218.0	5.0			
	80				80.6	0.20	0.37
Re-Austenitized & Tempered 400°F/1 hour	80	276.0	224.0	3.3			
	80				114.5	1.00	0.51
Re-Austenitized & Tempered 600°F/1 hour	80	252.0	215.0	3.1			
	80				166.0	1.00	0.77
Re-Austenitized & Tempered 800°F/1 hour	80	243.0	211.0	FOGL			
	80				178.5	1.00	0.845
Re-Austenitized & Tempered 900°F/1 hour	80	227.0	201.0	5.3			
	80				191.0	1.00	0.95

FOGL = Fractured Outside Gage Length.

TABLE 34

TENSILE AND CRACK PROPAGATION TEST RESULTS ON NRL HEAT 217, AUSROLLED AND TEMPERED COMPARED WITH AIR HARDENED AND TEMPERED. (a) AUSTENITIZED 1800°F (ARGON), AUSROLLED 75% AT 850-1050°F, OIL QUENCHED AND TEMPERED. (b) AUSROLLED, RE-AUSTENITIZED 1800°F (ARGON), OIL QUENCHED, AND TEMPERED.

Condition	Test Temp (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
(a) Ausrolled & Tempered 600°F/1 hour	80	274.0	240.0	5.0			
	80				201.0	1.00	0.84
Ausrolled & Tempered 700°F/1 hour	80	277.0	248.0	FOGL			
	80				220.0	1.00	0.89
Ausrolled & Tempered 800°F/1 hour	80	262.0	236.0	5.0			
	80				214.0	1.00	0.91
Ausrolled & Tempered 900°F/1 hour	80	259.0	228.0	8.0			
	80				208.0	1.00	0.915
Ausrolled & Tempered 1000°F/1 hour	80	166.0	135.0	7.0			
	80				157.0	1.00	1.16
(b) Re-Austenitized & Tempered 600°F/1 hour	80	196.0	153.0	8.5			
	80				183.0	1.00	1.20
Re-Austenitized & Tempered 700°F/1 hour	80	195.5	150.5	8.5			
	80				178.5	1.00	1.19
Re-Austenitized & Tempered 800°F/1 hour	80	194.5	150.5	9.0			
	80				177.5	1.00	1.18
Re-Austenitized & Tempered 900°F/1 hour	80	196.0	154.5	10.5			
	80				110.0	0.22	0.71
Re-Austenitized & Tempered 1000°F/1 hour	80	170.5	142.5	8.5			
	80				170.5	1.00	1.20

FOGL = Fractured Outside Gage Length.

TABLE 35

MAXIMUM FRACTURE STRENGTH RATIOS OBTAINED FOR EXPERIMENTAL AUSFORMING STEELS AT ROOM TEMPERATURE ACCORDING TO YIELD STRENGTH LEVEL.

	YIELD STRENGTH (KSI) NOT LESS THAN:									
	180	190	200	210	220	230	240	250	260	270
IDENT*										
200A								0.41	0.40	0.29
200R					0.54	(0.445)	(0.445)	0.445	0.435	
201A				0.67	(0.63)	(0.63)	0.63			
201R	1.02	0.98	0.93	0.85						
202A		0.94	0.90							
202R	(0.96)	0.96	0.78							
203A						0.62	(0.56)	0.56	(0.47)	0.47
203R		0.99	(0.945)	0.945	0.875	0.765				
204A								0.385	0.33	0.32
204R			0.95	0.845	0.51					
217A	(0.915)	(0.915)	(0.915)	(0.915)	0.915	0.91	0.89			
217R										

Note: A value in parenthesis is the same as the value for the next highest yield strength level and indicates that no higher value was obtained within that yield strength interval.

***A = Ausrolled and Tempered**
R = Re-austenitized, Quenched and Tempered.

TABLE 36

TENSILE AND CRACK PROPAGATION TEST RESULTS ON NRL HEAT 207: C 0.41%, UNALLOYED. BRINE QUENCHED FROM 1550°F, COOLED TO -100°F, THEN DOUBLE TEMPERED AS INDICATED.

Tempering Treatment	Test Temp (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
400°F/1 + 1 hours	200				57.8	0	
	80				57.5	0	
500°F/1 + 1 hours	200				81.5	0.10	
	80	174.0	151.0	2.5			
	80				71.3	0.05	0.47
600°F/1 + 1 hours	80	151.5	136.0	FOGL			
	80				156.0	0.60	1.15
	0				162.5	0.35	
700°F/1 + 1 hours	80	137.0	118.0	8.0			
	80				148.0	1.00	1.25
	0				152.0	1.00	

FOGL = Fractured Outside Gage Length.

TABLE 37

TENSILE AND CRACK PROPAGATION TEST RESULTS ON NRL HEAT 208: C 0.45%, Si 1.52%. BRINE QUENCHED FROM 1550°F, COOLED TO -100°F, THAN DOUBLE TEMPERED AS INDICATED.

Tempering Treatment	Test Temp (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
400°F/1 + 1 hours	200				43.1	0.13	
	78	258.0	227.0	2.0			
	76				45.0	0.13	0.20
500°F/1 + 1 hours	200				45.5	0.22	
	79	241.0	216.0	3.0			
	76				52.2	0.13	0.24
600°F/1 + 1 hours	200				57.3	0.26	
	79	219.0	217.0	FOGL			
	76				58.0	0.29	0.27
700°F/1 + 1 hours	80	191.5	185.0	FOGL			
	77				165.0	1.00	0.89
	0				184.0	1.00	
800°F/1 + 1 hours	80	166.0	151.5	FOGL			
	77				166.0	1.00	1.10
	0				166.0	1.00	
1000°F/1 + 1 hours	80	118.5	104.5	20.0			
	77				117.0	1.00	1.12
	0				120.0	1.00	

FOGL = Fractured Outside Gage Length

TABLE 38

TENSILE AND CRACK PROPAGATION TEST RESULTS ON NRL HEAT 209: C 0.42%,Mn 0.97%. OIL QUENCHED FROM 1550°F, COOLED TO -100°F, THEN DOUBLE TEMPERED AS INDICATED.

Tempering Treatment	Test Temp (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elogn. on 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
400°F/1 + 1 hours	200				146.0	1.00	
	80	283.0	225.0	6.0			
	80				109.3	0.45	0.485
500°F/1 + 1 hours	80	241.0	208.0	FOGL			
	80				169.0	1.00	0.815
	0				167.0	1.00	
600°F/1 + 1 hours	80	213.0	190.0	4.5			
	80				179.0	1.00	0.94
	0				182.0	1.00	
700°F/1 + 1 hours	80	197.5	178.0	7.0			
	80				178.0	1.00	1.00
	0				184.5	1.00	
800°F/1 + 1 hours	80	166.0	147.0	6.5			
	80				156.5	1.00	1.06
	0				160.5	1.00	
1000°F/1 + 1 hours	80	120.0	105.0	12.5			
	80				118.5	1.00	1.13
	0				123.0	1.00	

FOGL = Fractured outside gage length.

TABLE 39

TENSILE AND CRACK PROPAGATION TEST RESULTS ON NRL HEAT 213: C 0.41%, Ni 2.13%. OIL QUENCHED FROM 1550°F, COOLED TO -100°F, THEN DOUBLE TEMPERED AS INDICATED.

Tempering Treatment	Test Temp. (%)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
400°F/1 + 1 hours	200				48.4	0.10	
	80				48.4	0.10	
500°F/1 + 1 hours	200				66.3	0.15	
	80	230.0	199.5	2.5			
	80				58.5	0.15	0.29
600°F/1 + 1 hours	200				141.0	0.30	
	80	194.5	173.0	6.5			
	80				116.6	0.20	0.675
700°F/1 + 1 hours	80	171.5	150.0	8.0			
	80				158.6	1.00	1.06
	0				167.5	1.00	
800°F/1 +1 hours	80	134.0	126.0	10.0			
	80				130.0	1.00	1.03
	0				132.5	1.00	
1000°F/1 + 1 hours	80	93.3	81.0	15.0			
	80				89.8	1.00	1.11
	0				95.0	1.00	

TABLE 40

TENSILE AND CRACK PROPAGATION TEST RESULTS ON NRL HEAT 234: C 0.43%, Cr 1.95%. OIL QUENCHED FROM 1550°F, COOLED TO -100°F, THEN DOUBLE TEMPERED AS INDICATED.

Tempering Treatment	Test Temp. (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. On 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
400°F/1 + 1 hours	80	275.0	218.0	6.0			
	80				91.5	1.00	0.42
	0				108.6	0.45	
500°F/1 + 1 hours	80	251.0	211.0	5.0			
	80				151.0	1.00	0.715
	0				118.5	0.40	
600°F/1 + 1 hours	80	242.0	210.0	6.0			
	80				165.5	1.00	0.79
	0				157.0	1.00	
700°F/1 + 1 hours	80	226.0	202.0	5.0			
	80				180.0	1.00	0.89
	0				171.0	1.00	
800°F/1 + 1 hours	80	204.0	183.0	4.3			
	80				183.5	1.00	1.00
	0				188.0	1.00	
1000°F/1 + 1 hours	80	146.6	129.0	9.5			
	80				145.0	1.00	1.12
	0				150.0	1.00	

TABLE 41

TENSILE AND CRACK PROPAGATION TEST RESULTS ON NRL HEAT 214: 0.43% C, 2.15% Co. BRINE QUENCHED FROM 1550°F, COOLED TO-100°F, THEN DOUBLE TEMPERED AS INDICATED.

Tempering Treatment	Test Temp. (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elogn. on 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
400°F/1 + 1 hours	80	150.5		FOGL			
500°F/1 + 1 hours	200				131.0	1.00	
	80	161.5	127.0	2.7			
	80				116.4	0.25	0.92
600°F/1 + 1 hours	80	148.0	121.0	5.0			
	80				131.6	1.00	1.09
	0				129.4	1.00	
700°F/1 + 1 hours	80	133.5	105.5	7.5			
	80				120.0	1.00	1.14
	0				129.0	1.00	
800°F/1 + 1 hours	80	117.5	102.0	FOGL			
	80				109.0	1.00	1.07
	0				116.5	1.00	
1000°F/1 + 1 hours	80	82.2	72.2	FOGL			
	80				80.2	1.00	1.12
	0				57.7	1.00	

Note: No useful results obtained for tempering at 400°F because of extreme brittleness of material.

FOGL = Fractured Outside Gage Length.

TABLE 42

TENSILE AND CRACK PROPAGATION TEST RESULTS ON NRL HEAT 215: C 0.41%, Mo 0.56%. OIL QUENCHED FROM 1550°F, COOLED TO -100°F, THEN DOUBLE TEMPERED AS INDICATED.

Tempering Treatment	Test Temp. (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 in. (KSI)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
400°F/1 + 1 hours	200				73.2	0.30	
	80	205.0	164.0	3.0			
	80				55.8	0.20	0.34
500°F/1 + 1 hours	200				130.0	1.00	
	80	225.0	203.5	FOGL			
	80				79.0	0.35	0.39
600°F/1 + 1 hours	80	220.0	198.5	5.5			
	80				193.0	1.00	0.97
	0				198.0	1.00	
700°F/1 + 1 hours	80	197.0	186.0	FOGL			
	80				189.0	1.00	1.015
	0				196.5	1.00	
800°F/1 + 1 hours	80	187.5	175.0	5.5			
	80				181.0	1.00	1.03
	0				184.0	1.00	
1000°F/1 + 1 hours	80	155.0	148.0	7.0			
	80				155.0	1.00	1.04
	0				155.0	1.00	

FOGL = Fractured Outside Gage Length

TABLE 43

TENSILE AND CRACK PROPAGATION TEST RESULTS ON NRL HEAT 216: C 0.45%, V 0.10%. OIL QUENCHED FROM 1550°F, COOLED TO -100°F, THEN DOUBLE TEMPERED AS INDICATED.

Tempering Treatment	Test Temp. (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
400°F/1 + 1 hours	80	125.6	87.3	9.0			
	80				129.5	1.00	1.48
	0				131.5	1.00	
500°F/1 + 1 hours	80	116.6	91.2	FOGL			
	80				126.5	1.00	1.39
	0				142.0	1.00	
600°F/1 + 1 hours	80	132.0	98.6	FOGL			
	80				123.0	1.00	1.25
	0				132.5	1.00	
700°F/1 + 1 hours	80	131.0	96.2	10.0			
	80				126.0	1.00	1.32
	0				129.0	1.00	
800°F/1 + 1 hours	80	129.0	103.5	FOGL			
	80				111.0	1.00	1.07
	0				134.0	1.00	
1000°F/1 + 1 hours	80	93.3	69.0	17.5			
	80				93.0	1.00	1.35
	0				95.6	1.00	

FOGL = Fractured Outside Gage Length

TABLE 44

TENSILE AND CRACK PROPAGATION TEST RESULTS ON NRL HEAT 219: C 0.42%, Ni 0.99%, Cr 1.04%. OIL QUENCHED FROM 1550°F, COOLED TO -100°F, THEN DOUBLE TEMPERED AS INDICATED.

Tempering Treatment	Test Temp. (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
400°F/1 + 1 hours	200				134.5	1.00	
	80	258.0	199.0	5.0			
	80				79.0	?	0.395
500°F/1 + 1 hours	200				155.2	1.00	
	80	253.0	202.0	4.5			
	80				78.5	0.30	0.390
600°F/1 + 1 hours	200				189.0	1.00	
	80	248.0	209.0	5.0			
	80				110.5	0.40	0.53
700°F/1 + 1 hours	80	223.0	195.0	4.0			
	80				181.0	1.00	0.93
	0				113.6	0.30	
800°F/1 + 1 hours	80	193.5	175.0	6.5			
	80				188.0	1.00	1.075
	0				185.5	1.00	
1000°F/1 + 1 hours	80	143.0	124.0	FOGL			
	80				140.2	1.00	1.13
	0				149.0	1.00	

FOGL = Fractured Outside Gage Length

TABLE 45

TENSILE AND CRACK PROPAGATION TEST RESULTS ON NRL HEAT 220: C **0.43%**, Cr 1.00%, Mo 0.48%. OIL QUENCHED FROM 1550°F, COOLED TO -100°F, THEN DOUBLE TEMPERED AS INDICATED.

Tempering Treatment	Test Temp. (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
400°F/1 + 1 hours	80	290.0	234.0	4.3			
	80				131.0	1.00	0.56
500°F/1 + 1 hours	80	274.0	227.0	5.0			
	80				149.0	1.00	0.655
	0				98.2	0.35	
600°F/1 + 1 hours	80	253.0	214.0	5.0			
	80				168.0	1.00	0.785
700°F/1 + 1 hours	80	251.5	216.0	6.0			
	80				197.0	1.00	0.915
	0				154.0	1.00	
800°F/1 + 1 hours	80	216.0	193.0	6.0			
	80				186.0	1.00	0.965
	0				192.0	1.00	
1000°F/1 + 1 hours	80	181.0	163.0	9.0			
	80				169.0	1.00	1.035
	0				179.0	1.00	

TABLE 46

TENSILE AND CRACK PROPAGATION TEST RESULTS ON NRL HEAT 221: **C 0.41%**, Cr 0.87%, V 0.19%. OIL QUENCHED FROM 1550°F, COOLED TO -100°F, THEN DOUBLE TEMPERED AS INDICATED.

Tempering Treatment	Test Temp. (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
400°F/1 + 1 hours	80	284.0	248.0	5.0			
	80				123.0	1.00	0.495
	0				110.0	0.70	
500°F/1 + 1 hours	80	269.0	251.0	5.0			
	80				178.0	1.00	0.71
	0				126.0	0.55	
600°F/1 + 1 hours	80	251.0	239.0	5.0			
	80				193.0	1.00	0.81
	0				155.0	1.00	
700°F/1 + 1 hours	80	227.0	220.0	5.0			
	80				199.0	1.00	0.905
	0				206.0	1.00	
800°F/1 + 1 hours	80	199.0	193.0	6.0			
	80				197.0	1.00	1.02
	0				203.0	1.00	
1000°F/1 + 1 hours	80	157.0	150.0	10.0			
	80				157.0	1.00	1.045
	0				162.0	1.00	

TABLE 47

TENSILE AND CRACK PROPAGATION TEST RESULTS ON NRL HEAT 222: C 0.41%, Cr 0.87%, Co 0.78%. Oil QUENCHED FROM 1550°F, COOLED TO -100°F, THEN DOUBLE TEMPERED AS INDICATED.

Tempering Treatment	Test Temp. (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
400°F/1 + 1 hours	80	271.0	216.0	6.0			
	80				138.0	1.00	0.64
	0				128.0	1.00	
500°F/1 + 1 hours	80	256.0	213.0	5.0			
	80				125.0	1.00	0.585
	0				98.7	0.30	
600°F/1 + 1 hours	80	241.0	203.0	FOGL			
	80				144.0	1.00	0.71
	0				100.0	0.25	
700°F/1 + 1 hours	80	219.0	192.0	FOGL			
	80				180.0	1.00	0.94
	0				175.0	1.00	
800°F/1 + 1 hours	80	193.0	174.0	6.5			
	80				174.0	1.00	1.00
	0				181.0	1.00	
1000°F/1 + 1 hours	80	135.0	113.0	FOGL			
	80				131.0	1.00	1.16
	0				138.0	1.00	

FOGL = Fractured Outside Gage Length

TABLE 48

TENSILE AND CRACK PROPAGATION TEST RESULTS ON NRL HEAT 223: C 0.42%, Ni 0.99%, Mo 0.49%. OIL QUENCHED FROM 1550°F, COOLED TO -100°F, THEN DOUBLE TEMPERED AS INDICATED.

Tempering Treatment	Test Temp. (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
400°F/1 + 1 hours	80	267.0	220.0	6.0			
	80				183.6	1.00	0.835
	0				181.0	1.00	
500°F/1 + 1 hours	80	251.0	217.0	4.5			
	80				183.0	1.00	0.845
	0				187.0	1.00	
600°F/1 + 1 hours	80	226.0	201.0	5.0			
	80				200.0	1.00	0.995
	0				197.0	1.00	
700°F/1 + 1 hours	80	208.0	186.0	6.5			
	80				189.5	1.00	1.02
	0				198.0	1.00	
800°F/1 + 1 hours	80	182.0	167.0	6.5			
	80				177.0	1.00	1.06
	0				183.0	1.00	
1000°F/1 + 1 hours	80	161.0	149.0	10.5			
	80				160.0	1.00	1.07
	0				181.0	1.00	

TABLE 49

TENSILE AND CRACK PROPAGATION TEST RESULTS ON NRL HEAT 224: C 0.43%, Co 0.86%, Mo 0.54%. OIL QUENCHED FROM 1550°F, COOLED TO -100°F, THEN DOUBLE TEMPERED AS INDICATED.

Tempering Treatment	Test Temp. (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
400°F/1 + 1 hours	80	260.0	218.0	5.5			
	80				136.0	1.00	0.625
	0				118.0	1.00	
500°F/1 + 1 hours	80	240.0	212.0	5.0			
	80				166.0	1.00	0.785
	0				124.0	0.67	
600°F/1 + 1 hours	80	241.0	219.0	4.0			
	80				181.0	1.00	0.825
	0				174.0	1.00	
700°F/1 + 1 hours	80	218.0	196.5	4.0			
	80				185.5	1.00	0.945
	0				188.5	1.00	
800°F/1 + 1 hours	80	191.6	181.5	5.5			
	80				173.5	1.00	0.955
	0				178.0	1.00	
1000°F/1 + 1 hours	80	162.0	150.0	9.0			
	80				155.0	1.00	1.03
	0				158.5	1.00	

TABLE 50

TENSILE AND CRACK PROPAGATION TEST RESULTS ON NRL HEAT 225: C 0.43%, Ni 1.03%, Cr 0.98%, Mo 0.59%. OIL QUENCHED FROM 1550°F, COOLED TO -100°F, THEN DOUBLE TEMPERED AS INDICATED.

Tempering Treatment	Test Temp. (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
400°F/1 + 1 hours	80	290.0	236.0	5.0			
	80				122.0	1.00	0.52
	0				101.0	0.50	
500°F/1 + 1 hours	80	276.0	231.0	5.0			
	80				155.0	1.00	0.67
	0				111.5	0.40	
600°F/1 + 1 hours	80	266.0	230.0	5.0			
	80				183.0	1.00	0.795
	0				116.0	0.45	
700°F/1 + 1 hours	80	248.0	215.0	6.0			
	80				195.5	1.00	0.91
	0				147.0	0.65	
800°F/1 + 1 hours	80	227.0	204.0	6.0			
	80				188.0	1.00	0.92
	0				171.4	1.00	
1000°F/1 + 1 hours	80	193.0	178.0	6.0			
	80				183.0	1.00	1.03
	0				187.0	1.00	

TABLE 51

TENSILE AND CRACK PROPAGATION RESULTS ON NRL HEAT 226: C 0.43, Cr 0.98, Co 0.99, Mo 0.62. OIL QUENCHED FROM 1550°F, COOLED TO -100°F, THEN DOUBLE TEMPERED AS INDICATED.

Tempering Treatment	Test Temp. (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
400°F/1 + 1 hours	200				92.4	1.00	
	80	256.0	241.0	0.5			
	80				77.8	0.40	0.32
500°F/1 + 1 hours	200				123.5	1.00	
	80	267.0	237.0	FOGL			
	80				76.0	0.35	0.32
600°F/1 + 1 hours	200				165.0	1.00	
	80	*	*	*			
	80				77.4	0.40	(0.34)
700°F/1 + 1 hours	80	235.0	218.0	2.0			
	80				150.0	1.00	0.69
	0				93.4	0.20	
800°F/1 + 1 hours	80	227.0	207.0	6.0			
	80				186.0	1.00	0.90
	0				126.4	0.50	
1000°F/1 + 1 hours	80	192.0	176.5	8.5			
	80				181.0	1.00	1.02
	0				189.0	1.00	

* Specimen broke prematurely at quench crack

FOGL = Fractured Outside Gage Length

TABLE 52

TENSILE AND CRACK PROPAGATION TEST RESULTS ON NRL HEAT 227: C 0.45%, Cr 1.03%, Mo 0.53%, V 0.18%. OIL QUENCHED FROM 1550°F, COOLED TO -100°F, THEN DOUBLE TEMPERED AS INDICATED.

Tempering Treatment	Test Temp. (°F)	TENSILE STRENGTH (KSI)	Yield Strength (KSI)	Elongn. on 2 In. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
400°F/1 + 1 hours	80	290.0	249.0	4.5			
	80				106.0	1.00	0.425
	0				94.4	0.45	
500°F/1 + 1 hours	80	284.0	253.0	5.0			
	80				143.5	1.00	0.57
	0				106.0	0.45	
600°F/1 + 1 hours	80	270.0	249.0	5.0			
	80				179.0	1.00	0.72
	0				103.6	0.40	
700°F/1 + 1 hours	80	249.0	233.0	5.0			
	80				190.0	1.00	0.815
	0				124.0	0.90	
800°F/1 + 1 hours	80	231.0	220.0	5.0			
	80				196.5	1.00	0.895
	0				192.0	1.00	
1000°F/1 + 1 hours	80	205.0	191.5	7.0			
	80				189.0	1.00	0.99
	0				193.0	1.00	

TABLE 53

TENSILE AND CRACK PROPAGATION TEST RESULTS ON NRL HEAT 228: C 0.45%, Ni 1.00%, Cr 1.02%, Mo 0.53%, V 0.18%. OIL QUENCHED FROM 1550°F, COOLED TO -100°F, THEN DOUBLE TEMPERED AS INDICATED.

Tempering Treatment	Test Temp. (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
400°F/1 + 1 hours	90	296.0	248.0	3.5			
	90				126.0	1.00	0.51
	0				103.0	0.60	
500°F/1 + 1 hours	90	286.0	247.0	5.0			
	90				160.5	1.00	0.65
	0				119.0	0.75	
600°F/1 + 1 hours	85	276.0	251.0	5.0			
	90				202.0	1.00	0.805
	0				127.0	0.80	
700°F/1 + 1 hours	85	253.0	228.0	5.0			
	90				179.0	1.00	0.785
	0				159.0	1.00	
800°F/1 + 1 hours	85	236.0	221.0	5.0			
	90				198.0	1.00	0.895
	0				186.0	1.00	
1000°F/1 + 1 hours	85	205.0	191.0	8.0			
	90				194.0	1.00	1.015
	0				197.0	1.00	
Supplemental							
650°F/1 + 1 hours	80	264.0	238.0	6.0			
	80				183.0	1.00	0.77
	80				180.0	1.00	0.755
700°F/1 + 1 hours	80	251.0	234.0	6.0			
	80				209.0	1.00	0.895
	80				189.5	1.00	0.81
750°F/1 + 1 hours	80	241.0	224.0	6.0			
	80				190.5	1.00	0.85
	80				183.0	1.00	0.82

TABLE 54

TENSILE AND CRACK PROPAGATION TESTS ON NRL HEAT 233: C 0.48%, Ni 1.03%, Cr 0.99%, Mo 0.52%, V 0.19%. OIL QUENCHED FROM 1550°F, COOLED TO -100°F, THEN DOUBLE TEMPERED AS INDICATED.

Tempering Treatment	Test Temp. (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
400°F/1 + 1 hours	200				84.2	1.00	
	80	308.0	267.0	5.0			
	80				83.8	0.65	0.31
500°F/1 + 1 hours	200				98.4	1.00	
	80	302.0	267.0	5.0			
	80				82.0	0.50	0.305
600°F/1 + 1 hours	80	281.0	261.0	5.0			
	80				120.0	1.00	0.46
	0				95.4	0.30	
700°F/1 + 1 hours	80	264.0	248.0	5.0			
	80				167.0	1.00	0.675
	0				115.0	1.00	
800°F/1 + 1 hours	80	238.0	229.0	6.0			
	80				186.0	1.00	0.81
	0				159.0	1.00	
1000°F/1 + 1 hours	80	206.0	197.5	7.5			
	80				198.0	1.00	1.00

TABLE 55

MAXIMUM FRACTURE STRENGTH RATIOS OBTAINED FOR EXPERIMENTAL LOW ALLOY STEELS AT ROOM TEMPERATURE ACCORDING TO YIELD STRENGTH LEVEL.

	YIELD STRENGTH (KSI) NOT LESS THAN:									
IDENT.	180	190	200	210	220	230	240	250	260	TYPE
207										Plain C
208	0.89	(0.27)	(0.27)	0.27	0.20					Si
209	(0.94)	0.94	0.815	(0.485)	0.485					Mn
213	(0.29)	0.29								Ni
234	1.00	(0.89)	0.89	0.79						Cr
214										Co
215	1.015	0.97	0.39							Mo
216										V
219	(0.93)	0.93	0.53							Ni-Cr
220	(0.965)	0.965	(0.915)	0.915	0.655	0.56				Cr-Mo
221	(1.02)	1.02	(0.905)	(0.905)	0.905	0.81	(0.71)	0.71		Cr-V
222	(0.94)	0.94	0.71	0.64						Co-Cr
223	1.02	(0.995)	0.995	0.845	0.835					Ni-Mo
224	0.955	0.945	(0.825)	0.825						Co-Mo
225	(0.92)	(0.92)	0.92	0.91	(0.795)	0.795				Ni-Cr-Mo
226	(0.90)	(0.90)	0.90	0.69	(0.32)	(0.32)	0.32			Co-Cr-Mo
227	(0.99)	0.99	(0.895)	(0.895)	0.895	0.815	0.72	0.57		Cr-Mo-V
228	(1.015)	1.015	(0.895)	(0.895)	0.895	(0.805)	(0.805)	0.805		Ni-Cr-Mo-V
233	(1.00)	1.00	(0.81)	(0181)	0.81	(0.675)	0.675	(0.46)	0.46	Ni-Cr-Mo-V

Note: A value in parentheses is the same as the value for the next highest yield strength level and indicates that no higher value was obtained within the yield strength interval.

TABLE 56

TENSILE AND CRACK PROPAGATION TEST RESULTS ON NRL HEAT 240: Ni 19.5%, Ti 1.60%, Cb 0.32%, Al 0.27%. SOLUTION TREATED 1500°F FOR 1 HOUR, AIRCOOLED. AGED 950°F FOR 1 HOUR.

Test Temp. (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
580	220.0	212.0	FOGL			
580				150.0	1.00	0.71
400	234.0	226.0	2.7			
400				200.0	1.00	0.885
300				217.0	1.00	0.94
250				184.5	1.00	0.79
200	248.0	238.0	3.5			
200				203.0	1.00	0.845
170				158.0	1.00	0.655
140				78.6	0.40	0.32
76	265.0	254.0	3.0			
77				65.6	0.25	0.26
0	276.0	270.0	2.7			
0				57.3	0.15	0.21
-100				47.2	0.10	0.175
-200				39.8	0.05	0.135
-280	324.0	316.0	2.0			
-280				36.0	0.00	0.115

FOGL = Fractured Outside Gage Length

TABLE 57

TENSILE AND CRACK PROPAGATION TEST RESULTS ON NRL HEAT 241: Ni 24.3%, Ti 1.60%, Cb 0.52%, Al 0.23%. SOLUTION TREATED 1500°F FOR 1 HOUR, AIRCOOLED. 1300°F FOR 4 HOURS, AIRCOOLED. AGED 950°F FOR 1 HOUR.

Test Temp. (°F)	Tensile Strength (KSI)	Yield Strength (KSI)	Elongn. on 2 in. (%)	Nom. Net Strength (KSI)	Fracture Shear Fraction	$\frac{F_n}{F_{ty}}$
580	212.0	179.0	4.5			
580				172.0	1.00	0.96
400	226.0	191.0	4.0			
400				172.0	1.00	0.90
300				174.0	1.00	0.855
250				166.5	1.00	0.785
200	236.0	219.0	5.5			
200				162.0	1.00	0.74
140				139.5	1.00	0.65
120				110.5	0.80	0.525
100				109.0	0.60	0.525
77	247.0	207.0	7.0			
76				103.6	0.55	0.50
0	260.0	209.0	7.0			
-100				65.0	0.20	0.29
-200				53.4	0.15	0.22
-280	305.0	263.0	5.3			
-280				49.0	0.10	0.185