INFLUENCING PARAMETERS ON THE FRACTURE RESISTANCE OF ALUMINUM ALLOYS

by

JOSEPH SIGMUND WYCECH
B.C.E., University of Detroit
(1970)

Submitted in partial fulfillment

of the requirements for the degree of

Master of Science
in Civil Engineering

at the

Massachusetts Institute of Technology

(September 1971)

Signature of Author ....
Department of Civil Engineering, (September 1971)

Certified by ....
Thesis Supervisor

Accepted by ....
Chairman, Departmental Committee on Graduate Students

ABSTRACT

INFLUENCING PARAMETERS ON THE FRACTURE

RESISTANCE OF ALUMINUM ALLOYS

by

JOSEPH SIGMUND WYCECH

Submitted to the Department of Civil Engineering on July 7, 1971, in partial fulfillment of the requirements for the degree of Master of Science.

Due to the recent use of high strength aluminum alloys fracture resistance could no longer be ignored, and aluminum alloys had to be categorized on the basis of strength and toughness. To rate the alloys, three accepted measures of toughness; the unit propagation energy, the notch-yield ratio, and the stress intensity factor have been used over such values because they indicate correctly yielding before fracture and can be used to scale very ductile or tough alloys. Also in scaling one alloy against the other, testing variable such as notch geometry, notch root radius, the ratio of crack length to specimen width, and the type of test must be considered with regard to their similarity.

Not only have recent investigations been involved in the merit rating of alloys but also in the understanding of the effects of compositional changes or the employment of working or heat treatment. Generally increasing alloy additions increases strength and reduces toughness, and working and heat treatment have the same effect depending upon their degree. Along with compositional changes and variable processes, quality control can prove to be beneficial since a large amount of defects such as inclusions, intermetallic compounds, and porosity can create notch effects which result in a brittle state and lower toughness. The above is applicable to both wrought and cast alloys.

In design, toughness must be considered and can be critical if thick plate is used in welded or unwelded redundant members. Usually thick plate exhibits a fracture resistance which varies through the thickness and in the longitudinal direction with the short transverse direction having a toughness which is the least for plate but may be greater than that of

the weld metal for a given yield strength. Concerning welds, the toughness of the filler metal can be greater than that of the parent plate, but the yield strength may be less; that is, the toughness of the filler can be greater than that of the parent plate on the basis of fracture resistance alone.

Thesis Supervisor: Frederick J. McGarry

Title: Professor

ACKNOWLEDGMENT

The author wishes to thank Professor Frederick J. McGarry for his support in conducting this survey of published material on the fracture resistance of aluminum alloys. And gratitude must be given to the people affiliated with Project Intrex at the Institute for their help in securing pertinent articles.

TABLE OF CONTENTS

Introduction:

The rapidly advancing aerospace technology in the last ten years has produced an increased demand for materials which can sustain a high working load under a variety of severe conditions such as cryogenic temperatures and corrosive environments. One particular group of these materials are the aluminum alloys. The aluminum alloys have an advantage over many other materials for aerospace applications in that they have adequate strength with a minimum of weight and that the majority of the alloys do not exhibit brittle behavior with decreasing temperature. But even with these inherent characteristics there was some uncertainty of the behavior of these alloys under the newly imposed conditions such as uses for large support members in aircraft or space vehicles or for storage vessels containing cryogenic liquids. Consequently research was renewed to disclose the characteristics of these alloys under their new operating conditions and also possibly to quantitatively measure the fracture resistance of these alloys for use in design.

This latter objective of the research is of interest because the aluminum alloys were thought of as being so tough that fracture resistance need not be considered. But this was no longer true, for the alloys were being used in greater thicknesses and higher strengths which inherently meant a decrease in resistance. Thus the fracture resistance of these alloys was seen from a new perspective, and a consequence of this was an overall view of the effects of temperature, composition, processes, and quality control on the toughness of the aluminum alloys.

The nature of the research was such that it was conducted on both a qualitative and quantitative level with the majority of the work being done by the Aluminum Company of America and Battelle Memorial Institute. Thus a disclosure of what has been discovered would be beneficial to both the designer and materials specialist.

The discussion will proceed with the research that has been done in establishing proper testing methods and quantitative data related to the fracture resistance of these alloys and then will lead into the analysis of the effects of the previously mentioned factors on the fracture resistance of aluminum alloys.

Measuring Fracture Resistance

The simplest means of determining fracture resistance under static loading is by measuring the elongation or reduction in cross secional area of a sample. But both these characteristics or properties have their drawbacks in that elongation is a measure of a material to deform locally and uniformly which can be misleading since it is a combination of two properties, and both elongation and reduction in area are not correlated to strength which is critical for considering fracture resistance. Pure aluminum can be used as an example where it deforms greatly but has little fracture resistance because of its low strength.

To demonstrate the inability of elongation and reduction in area in measuring fracture resistance, a comparison between these two properties and the accepted qualitative measures of resistance can be shown (1). These accepted measures are the notched tensile strength to tensile yield ratio (NYR) and the unit propagation energy (UPE), but before discussing the above correlation between the two properties and the accepted measures, the measures must be defined and discussed to demonstrate their acceptance as standards.

The notch-yield ratio (NYR) as obtained from tensile tests on notched round specimens measures the ability of a material to deform plastically in the presence of a severe stress raiser (2,3,4), and it categorizes a material on a relative basis as having a high notch toughness if it has a notch-yield ratio greater than 1 since this shows that extensive plastic deformation has taken place in the presence of a notch before the specimen fails in tension. Thus on this basis the NYR is of value for qualitatively comparing alloys for their fracture resistance provided the test specimens for the alloys considered are the same since variability in test specimens can result in an invalid comparison.

The NYR not only has the advantage of measuring the possibility of a material to yield locally but also is a better qualitative measure of toughness with a variety of specimen notch geometries than the notch tensile strength (5). Figures 1 and 2 show the notch strength and notch yield ratio as a function of various alloys and notch geometries. The plot for the notch strength as a function of the above shows a considerable spread for a particular alloy, but in the plot for the NYR, the spread is minimized somewhat and there appears to be a better correlation. There is less confusion in the relative ratings of notch toughness for any alloy in Figure 2 over the relative rating of the same alloy in Figure 1. This is illustrated in Figure 2 by the fact that a deep round notch may have a higher NYR than a deep sharp notch for a number of alloys, but this does not hold for the notch strength as shown in Figure 1. Thus in this way the NYR is a valid means of qualitatively rating aluminum alloys if the notch geometry is taken into account.

The unit propagation energy (UPE) is the energy to propagate a crack in a Kahn type tear specimen. And it is simply measured as the area under the load deformation curve divided by the net sample area (see Figure 7 for specimen and proper area of load deformation curve). The acceptance of the UPE is derived from the fact that it is inexpensive to determine by means of a tear test, it can be determined for any specimen orientation within any alloy product, and can compare brittle and very ductile alloys on a quantitative basis because there are no test restraints on the mode of fracture, level of applied stress, nor amount of plastic deformation. Another reason that it is a primary measure of fracture resistance is that it takes into account both strength and ductility of the alloy and is a measure of conditions for crack instability.

Other tests similar to the tear test are the drop weight tear test and the explosive tear test (6), but their use in determining a propagation energy has not been widely accepted possibly because the explosive tear test requires abnormal testing conditions and the drop weight tear test is basically a charpy impact test. The results of such a test should not be used for merit rating or comparing aluminum alloys (7) because the purpose of a charpy test is to determine a transition temperature. But for aluminum alloys a transition temperature does not exist.

Thus using the NYR as obtained from tensile tests on a one-half inch diameter specimen, Figures 3 and 4 show that there is a broad and not very useful relationship between elongation, reduction in area, and the NYR. Figures 5 and 6 show the same correlation for elongation and reduction in area except that the UPE is used as a basis for a comparison. The last two figures again show a wide scattering of data points. Thus from these experimental results, the elongation and reduction in area of an aluminum alloy cannot be considered as valid measures of fracture resistance.

As stated previously, the NYR and UPE are accepted as valid measures of fracture resistance on a qualitative or relative basis. But there is a third measure, namely, fracture toughness which is a design related, quantitative index of fracture resistance. The first two measures have for their purpose the screening and comparison of aluminum alloys during development, and the third measure of resistance has been developed for use in design based upon fracture strength rather than a yield or plastic failure criterion.

Fracture toughness is measured in a variety of ways. Usually tensile tests of center slotted and single edge notched bend tests are employed. Figure 8 shows typical specimen configurations for these tests (3,8,11). For all these test specimens, a "thickness" and "deviation" criterion must be met to insure measurement of toughness under plain strain conditions in the specimen (9). That is, the thickness of the specimen must be 2.5 times the ratio $\left(k_{I_c}/\sigma_{YS}\right)^2$, and any deviation from linearity in the load deformation curve prior to the load used for the k_{I_c} calculation must primarily represent crack extension as indicated by:

secant offset at $0.8P_5 \leq \frac{1}{4}$ secant offset at P_5.

where P_5 is the load at secant offset of 5 per cent.

This can also be restated that the horizontal displacement of the load deformation curve from the initial slope at a load of 80 per cent of that at the 5 per cent secant offset intercept shall not be more than $\frac{1}{4}$ of the displacement at the 5 per cent secant offset intercept (8,9).

Typical load deformation curves for center notched, single edge notched, and notch bend specimens are shown in Figures 9, 10, and 11. And if the load deformation curves conform to the deviation criterion and the specimen to the thickness requirement, plane strain conditions will be exhibited, and the fracture toughness will be of its lowest or most critical value. But two other criteria must also be checked before the above statement is completely valid; they are: rapid fracture must take place at nominally elastic stresses and that all specimens must be fatigue cracked because machine notching can lead to higher or unconservative values of k_{I_c} or fracture toughness. Only after all the above constraints are satisfied will the value of k_{I_c} be a reliable parameter useful in design.

Recently work has been done in correlating the three measures of fracture resistance. The reason for this is that in the tests determining fracture toughness not all the criteria can be met especially maintaining a nominal elastic stress.

Thus the reported values of toughness for very ductile and tough alloys are invalid, and if a general relationship can be determined between the results of qualitative tests and fracture toughness, a true plain strain intensity factor for a tough alloy can be determined by simply evaluating the UPE and then using the general relationship. This has been done (2,4,10), and Figures 12, 13, and 14 show a direct correlation between the critical stress intensity factor (Fig.12), the plain strain intensity factor (Figs.13 and 14) and the unit propagation energy. Note that the values for the stress intensity factors are obtained by means of tensile tests on center slotted panels and the propagation energies by Kahn type tear tests. Also Figure 13 is essentially the same as Figure 14 except that it employs a linear approximation whereas the latter uses a parabola to fit the data. But neither curves differ by an appreciable amount. Thus it is possible to determine the toughness, k_c or k_{I_c}, of an aluminum alloy no matter how tough it is by means of a tear test, and it offers confidence in the merit rating of the various alloys on the basis of tear test results.

Similar conclusions using the notch yield ratio correlated with fracture toughness have evolved (4), but the correlation is only good for rating or comparing alloys since no direct relationship between the NYR and k_c or k_{I_c} can be established with confidence because the NYR is dependent upon specimen geometry and is thus variable. Also Figure 15 shows a lack of correlation between the NYR and UPE with a NYR greater than 1. Thus if a linear relationship holds for the fracture toughness and UPE, there cannot be a direct correlation between the NYR and toughness with a NYR greater than 1, and this seems to be validated by Figure 16 except the linear extrapolation for higher ratios is in some doubt. Thus the qualitative tests are valuable in comparing or merit rating of alloys, but the results of the tear test also have a direct relationship with fracture toughness and is thus more valuable especially in considering the inexpensiveness of the test and the few constraints upon the testing conditions.

For merit rating of aluminum alloys and also for their measurement of toughness, there is an important consideration, namely, that alloys cannot be compared on the results of tests using different specimens and/or notch configurations or similar stress concentration factors with different specimens or notches. Table 1 and Figure 18 show the results of tensile tests taken on the specimens in Figure 17. Figure 18 clearly shows a dependence of the notch yield ratio upon the theoretical stress concentration factor which implies that data for one alloy cannot be compared to data for another alloy with the same concentration factors without specifying the notch geometry (12). From Table 1 one can see that the large specimens have an appreciable variation in notch-tensile strength or, in turn, a notch-yield ratio with different notch-tip radii and corresponding concentration factors, but for other specimens with a shallow notch the implied NYR varies slightly (12). Also the toughness of the alloy is different for various configurations of stress concentrators. For a particular k_t, one type of specimen provides a lower notch yield ratio over another specimen, but for varying concentration factors, the reverse is shown in Table 1. Note that even specimen thickness has a varied influence on the toughness of the alloy for various stress raisers, and this is also validated by Table 2 which shows the NYR as a function of specimen thickness as derived from the results of a slow bend test (13). Table 2 shows the ratio decreasing with increasing thickness thus inferring a possible constraint on local yielding. But the calculated plane strain concentration factor does not vary with specimen thickness thus showing a lack of correlation between the NYR and k_{I_c} and definite plane strain conditions. But the point is that for merit rating of aluminum alloys based upon the NYR or toughness, all the above factors embedded in specimen design must be considered in order to validly scale one alloy against another for their fracture resistance.

Other specimen parameters which must be considered when comparing alloys are the widths and crack lengths of the test specimens. Figures 19 and 20 show that the calculated toughness k_c using the notch strength analysis[1] of a center notch panel increases with increasing specimen width for a given crack length and conversely decreases with decreasing crack length of given panel width. For the particular alloy plate tested, the ratio of highest to lowest k_c is about 1.5 for a crack length to width ratio of 0.33 indicating a toughness-specimen configuration dependence (14). This is invalidated somewhat by Figure 21 which shows the variation in crack extension force, G_c, which is directly realted to toughness with specimen width at 78°F. and -320°F. But these plots are obtained by surface flow specimens which no doubt show a different response to tensile fracture than a through crack specimen. The discrepancy could probably be due to the difference in crack propagation system and variability of grain structure through which the crack is propagating. Note that for the material of higher fracture toughness in Figure 21, the specimen width affects the toughness more; thus for materials of low toughness, narrower specimens can be used to obtain satisfactory results (15). But the toughness is relatively invariant with specimen width for the 2014-T6 alloy which does not necessarily imply the same result for other alloys (15). One cannot compare different alloys with various specimen widths assuming other parameters to be the same and say that there is a relative correlation of fracture data between alloys. The specimen width should be taken into account along with the crack length to width ratio despite the results of Figure 21.

But as stated before, the lack of correlation between Figures 20 and 21 could be based upon the differences in the testing method and not necessarily the response of the material.

Another important aspect of comparing alloys based on their fracture resistance is the specimen type and crack propagation system as used in a fracture toughness test. As mentioned previously, the fracture toughness of an alloy can be obtained by means of tests on center slotted panels, single notched edge tensile tests, or notch bend tests (3,16,17). Tests using surface flaw or wedge opening specimens have been run (16) and have shown data for the critical stress intensity factor which varies considerably from the values obtained from the other above named tests. Figure 22 shows the crack propagation systems employed in the notch bend and wedge opening tests and illustrates the propagation direction with respect to R, the rolling direction, W, the plate width direction, and T, the direction of plate thickness. These coordinate systems are also used in the other specimen types. Figures 23 and 24 show the results of toughness tests using the above specimen types and various propagation systems. From Figure 23 the SF specimen always yields the highest toughness for the same crack propagation system followed by, in decreasing order, the SB, SEN, and CN specimens with the same crack propagation system. Also for the same material and crack system the SEN and CN results are very close, and the toughness as determined by the SB and SF specimens give results which are higher than SEN and CN specimens. The reason for the high results of the SB and SF specimens could be due to the friction effects of the three point loading in the SB test and the sensitivity to material variability in the SF specimens (16). However, the overall variability of the results in Figure 23 is probably due to the differences in specimen type. Figure 23 also shows that there is no strong crack orientation effect; that is, the results from different crack propagation directions as derived with the same specimen types are about the same even though the crack orientation is different. But one can see from Figure 23 that there is a strong crack propagation direction influence upon toughness of a particular alloy with the same specimen type. Thus the crack propagation direction is important when a comparison of alloys is made using any of the above tests. Figure 24 also shows results which validate this point.

Shown in Figure 25 is the effect of the root radius upon the toughness with a "corner" characteristic of the various graphs as the $\sqrt{\rho}$, $\sqrt{\text{the root radius}}$, reaches a value of 0.1 $\sqrt{\text{inches}}$ (18). Thus if the artificially produced cracks in the test specimens do not have small enough radii, inaccuracies in predicting the toughness will result. Table 3 shows the results of tests using three accepted toughness test specimens of various alloys. The data is obtained using the "thickness" and "deviation" criteria as pointed out previously, and the results show that the apparent toughness as determined from the initial departure from linearity or the 5 per cent secant offset in the SEN and NB tests agree quite well. But the results of the CN tests are higher by 5 to 20 per cent. This is due to the fact that the notch in the CN specimen is not fatigue cracked and possibly the "corner" on the toughness-root radius curve is not achieved by machine notching. Thus the importance of achieving a sharp enough crack is the acquisition of valid values of toughness, k_{I_c}, and recent ASTM requirements state that all toughness specimens must be fatigue cracked so that a root radius of less than 0.001 inches be achieved.

Thus when making a comparison of various aluminum alloys based upon their fracture resistance as measured by the notch yield ratio or fracture toughness several points must be considered:

1. The notch geometry of the various specimens along with theoretical stress concentration factors should be the same for the various alloy specimens.
2. The notch tip radius for each specimen notch should be less than 0.001 inches.
3. The thickness should be greater than 2.5 $(k_{I_c}/\sigma_{yS})^2$ when measuring the plain strain intensity factor.
4. The same crack length to specimen width ratio should be maintained for each alloy specimen tested.

5. Any correlation of fracture resistance of one alloy versus another must be based upon the same specimen type; that is, if toughness of various alloys are to be compared, results from tests on CN specimens should only be used and not results from a variety of tests.[2]

6. The crack propagation direction is important in making a comparison of various alloy plate because they show a considerable degree of anisotropy,and the same propagation directions should be maintained for all alloy specimens so as to make a comparison based upon toughness in a specified direction.

Thus when only considering the above qualifications is one prepared to make a comparison of aluminum alloys based upon tests to determine their fracture resistance.

Compositional Effects

For an aluminum alloy, the amount and relative proportion of alloying elements not only have a substantial effect upon yield strength and elongation but also fracture resistance. Figure 26 and 27 clearly validate this by showing a probable band of fracture toughness as given by the plane strain stress intensity factor and the unit propagation energy versus the yield strength. The band is given for two series of high strength alloys; the 2xxx series which has copper as its main alloying component and the 7xxx series which is alloyed with zinc, magnesium, and copper. For a given yield strength, the 7xxx series has a higher toughness than the 2xxx series which is due to the different alloying elements. And for a given series, the toughness varies inversely with the yield strength due to the amount and type of alloying element and the heat treatment employed. Both series of alloys derive their strength from the precipitates which are formed with the hardening elements of zinc, magnesium, and copper. And the type and amount of precipitate that is formed is dependent upon, again, the heat treatment. But for this present discussion only the effects of the addition of different alloying elements will be revealed and the effects of heat treatment will subsequently follow.

Most of the investigation into the effects of aluminum alloy composition on the room temperature fracture resistance has been directed towards the high strength series, 2xxx and 7xxx. This is no doubt due to the fact that the 5xxx and 6xxx series and their various tempers are relatively tough but have a lower yield strength. Thus the aim of the investigation is to determine the possibility of obtaining high strength alloys with a high level of toughness. But Figures 26 and 27 illustrate that an optimum appears to be unlikely at this time. What is important though in the results of the investigation is that likely reasons for the inherent inverse relationship between toughness and strength of the two series have been published and are worthwhile discussing.

The results of the investigation show that the nature and number of precipitates in the above heat treatable alloys is the key to explaining how adequate fracture resistance is derived. But first, the source and formation of these hardening particles must be disclosed for a better understanding of a possible approach to achieving the above objective of optimizing toughness and strength.

For both series the source of yield and fracture strength is the amount of precipitate which is formed from the alloying elements. If the solid solution[3] becomes supersaturated with alloying elements at high temperatures, the solute formed by the alloying elements will precipitate out of the solid solution and form concentrations of solute as the temperature decreases. Naturally, the amount of precipitate depends upon the degree of supersaturation and the decreasing temperature. If the quenching is rapid, the supersaturated state is maintained to room temperature after which the precipitates begin to form until equilibrium is reached as the alloy ages naturally or artificially.[4] If the quenching is slow, the precipitates will form without aging along grain boundaries rather than in the matrix and will cause a reduced strength of the alloy. The precipitates can be either found in the matrix itself or along grain boundaries. But in general, the precipitates will form or nucleate at sites of greater disorder and higher energy such as grain boundaries, subgrain boundaries, and dislocations[5] (19).

Close to room temperature or during the initial stages of artificial aging at moderate temperatures, solute in the solid solution begins to cluster together and form Guinier-Preston (GP) zones within the matrix. This concentration of solute causes a distorted lattice within the matrix and the zone itself. Thus this distortion causes an impedence to dislocation movement which means a reduced fracture resistance by creating local stress concentrations and an increased yield strength. As aging continues, these zones form a transition precipitate which is different from the crystal structure of the solid solution and the final equilibrium phase. The transition precipitate is continuous with the solid solution; that is, both precipitate and solvent exhibit the same strain and are not separated by any phase boundary. In other words, the two are coherent.

After prolonged aging the transition particles grow in size and reach the equilibrium condition which means that coherency with the solid solution is lost because the bond between the two is broken. This results in a decrease of yield strength and demonstrates that the heat treatable alloys have a peak yield strength which is dependent upon the aging conditions.

One of the major reasons that the GP zones form at low temperatures is that there are vacancies in lattice positions of the crystal. That is, atoms do not occupy all the expected positions. Consequently, at these locations the solute can concentrate to form the strengthening zones. Another inherent characteristic of the vacancies is that they can migrate just as dislocations and one result of this is that sites for nucleation of precipitates adjacent to grain boundaries do not exist because vacancies and dislocations have moved to the boundaries leaving a depleted zone of nucleation sites. This depleted zone is distinct from the grain interior and the boundary and is termed the precipitate free zone (PFZ). The PFZ is thought of as being a major influence upon the fracture resistance of these high strength alloys, and it usually varies directly with the amount of grain boundary precipitate. But grain structures are possible with a narrow PFZ and large amount of boundary precipitate depending upon the heat treatment employed.

With this general introduction of the critical parameters which control the mechanical properties of high strength aluminum alloys, the differences in fracture resistance as a function of yield strength for the two series can be given.

Basically, for the 2xxx series which is formed with alloying elements of copper alone or copper and magnesium, the precipitates are in the form of flat platelets which are aligned with the [100] crystal planes. For the 7xxx series, GP zones are the main dislocation barriers and are usually spherical in shape (20). For example, two alloys of the same yield strength, 2024-T86 and 7075-T6, have a considerable difference in toughness as given by Figure 26. The reason for this is that the larger precipitate platelets (500 to 1000 Å) of the 7075-T6 alloy cause dislocation pile-up which result in an increase in local stress, possible void nucleation, and reduced toughness. Thus the difference in size of the precipitates in the two alloy series is the reason for the great difference in toughness.

The principle hardening elements of both series have generally the same result upon toughness as measured by the UPE as their amount added to the solution is increased. Figures 28, 29, and 30 show the effects of the addition of zinc plus magnesium, zinc, and the saturation ratio of zinc, magnesium, and copper.[6] Note that the UPE steadily decreases with increasing saturation ratio or per cent addition until a near zero value is reached (actually the constant UPE is estimated at 50 in $-lb/in^2$). This is no doubt due to the increased number and size of precipitates which are formed and interfere with dislocation mobility.

In Figure 26, the 2020-T6 alloy has the lowest toughness and highest yield strength with compositions of 4.5 pct. copper, 0.2 pct. cadium, and 1.3 pct. lithium. The last two alloying elements cause a greater number of precipitates to be formed and thus reduce the toughness even though the copper content is the same as in the 2024 alloy without the additions. Thus the slight additions of these alloying elements must create mucleation sites for further precipitate growth (20,21).

Similar to the effect of additions of cadium and lithium to the aluminum copper series is that of manganese. Manganese when added to aluminum copper alloys containing 3 and 4.5 pct. copper causes nucleation of precipitates. But the distinction is that these precipitates do not form along grain boundaries but rather at alternate sites in the matrix afforded by fine particles of manganese containing compounds. The result is transcrystalline fracture which implies an increased fracture resistance (22). This has been validated up to manganese contents of 2 pct., and any increase in this percentage will cause grain boundary precipitation to increase and thus reduce fracture resistance. But no results of tests using increased manganese contents above 2 pct. have been reported and the above conclusion is speculative but does seem reasonable in the light of the previous discussion of the increasing additions of the other alloying elements.

Note that this discussion has been restricted to the effects of the amounts and sizes of the precipitates, and only a brief mention has been made with respect to the effects of the location of precipitates.

As implied previously, the grain boundaries appear to be the preferred locations for the nucleation of precipitates. And since increases in the number of precipitates means a decreased fracture resistance and intercrystalline fracture, a discussion of the nature of cracking along the grain boundary is necessary for an understanding of the fracture process.

Figure 31 shows the effect of the area fraction of the grain boundary precipitates upon the plane strain stress intensity factor and the mode of fracture. This figure shows that toughness is reduced and cracking is primarily confined to the grain boundaries as the amount of precipitates increases. For predominantly intercrystalline fracture and low toughness, the amount of precipitate has little effect.

The reason that the grain boundary precipitate confines the cracking to the boundary is that with higher concentrations there is a difference in strength between the boundary areas and the interior and that larger precipitates reduce the strain necessary to create voids ahead of the advancing boundary crack meaning that the cracks will not go into the grains but follow the boundaries even though going through the former is the most direct route (23). Also if the matrix yield strength is higher than the grain boundary region, localized strains will occur in the region thus increasing the likelihood of grain boundary cracking with larger precipitates. Thus grain boundary precipitates play a definite role in the fracture resistance of the high strength alloys and even more so than the matrix precipitates. This can be seen in Figure 30 which is derived by using heat treatments which keep the number and size of matrix precipitates constant and in using fine manganese particles in an aluminum copper alloy which results in transgranular fracture rather than intergranular simply because the number of boundary precipitates is reduced assuming that a maximum of 2 pct. manganese is employed.

Also as part of the study performed on the effects of boundary precipitates are the effects of varying the PFZ width (23). Originally large PFZ widths were thought of as being necessary to provide adequate fracture resistance because they could relieve local stresses by yielding in the PFZ when narrow bands of dislocations impinged upon the grain boundary in concentrated alloys. This interaction of the wide PFZ and impinging dislocations could possibly reduce the susceptability to boundary cracking especially if larger precipitates are present

(24). But this has not been validated by any test measuring the accepted quantities of fracture resistance as a function of the PFZ width. What actually has been discovered and can be seen in Figure 32 for an Al - 6% Zn - 3% Mg alloy is that the plane strain stress intensity factor along with the mode of fracture is invariant with the PFZ width for a given yield stress. This is contrary to what has been formerly proposed, but it illustrates the uncertainty of what role the PFZ width actually plays in the fracture resistance of high strength aluminum alloys. This test which is performed by means of a notch bend test on a variety of samples under a variety of heat treatments is a step in the right direction. But more data has to be gathered before one can positively say that there is no correlation between the PFZ width and fracture resistance.

An important thing to recognize in the above discussion on the effects of grain boundary precipitates and PFZ width on fracture resistance is that sufficient resistance can only be obtained if cracks are prevented from nucleating and propagating in the grain boundaries. And this is the key to obtaining high yield strengths with high toughness.

One effect of increasing additions of alloying elements to the solid solution has been decreased toughness with cryogenic temperatures. The reason for this is that the substitutional atoms usually have varying atomic radii which are quite different from the parent aluminum atoms. This difference in radii between atoms causes a distortion in the lattice, and this distortion becomes effective at very low temperatures (-320°F. to -423°F.) because the thermal energy of the atoms decreases and restricts the movement of atoms or molecules. Thus the substitutional elements are frozen in place and can act as dislocation barriers which may lead to local stress concentrations, and reduced toughness. At higher temperatures (-320°F. to +100°F.) the atoms or molecules can move about and cannot impede dislocation movement. Thus the alloy remains relatively tough. But note that this is only valid for a specific composition of alloy within a temperature range, but if more or less of the alloying elements are added, the toughness will decrease or increase accordingly as pointed out previously. Thus the distorted lattice approach to decreased toughness is only applicable for the -320°F. to -423°F. range.

Figure 33 shows the effect of using substitutional elements of varying radii upon the notched to unnotched tensile strength which isn't a valid measure of resistance but should suffice to show the effect of varying the substitutional atomic radii upon the fracture resistance of a variety of aluminum alloys. Note that the horizontal axis of the figure gives the summation of the product of the atom per cent and the difference in effective atomic radii between the elements and aluminum (26). And for increasing abscissa values, the measure of resistance decreases at the temperature of -320°F. The reason for this is the dislocation pile-up due to larger substitutional particles which are rather immobile.

A typical example of the effect of different radii can be seen in Figure 34[7] by comparing the notched to unnotched tensile ratios of 2024-T3 and 6061-T4 alloys at -423°F. Both of these alloys have a drastic reduction of fracture resistance as compared to the 2014-T6, 2024-T4, 2219-T81, 2219-T3, and 6061-T6 alloys because 2024-T3 has a highly strained lattice due to increased alloying additions up to 8.7 pct. as compared to 6.5 pct. for the other alloys in the 2xxx series and 6061-T4 has larger amounts in solid solution of Fe, Mg, and Si which have large effective atomic radii as compared to aluminum. Note that the alloy composition of 6061 is 2.7 pct. as compared to 6.5 pct. of 2014 or 2219 (see Table 4) implying an increased toughness at -423°F., but this is not so due to the large effective radii of Fe, Mg, and Si which cause local stress concentrations and reduced toughness.

Similarly other alloys can exhibit this decreased toughness down to -423°F. Specifically some of the common alloys in the 5xxx and 7xxx series have been tested for their toughness down to -423°F. and show that the toughness decreases considerably from -320°F. to -423°F. for the 5xxx series alloys and from room temperature to -423°F. for the 7xxx series alloys (27,28). This can be seen in Figures 35 and 36 which show the toughness as measured by the notch to unnotched tensile ratio as a function of temperature for several alloys in both series. Note that the toughness of the 7xxx series alloys decreases faster with decreasing temperature than the 5xxx series. This is the result of the 7xxx series alloys having a larger percentage of alloying elements than the 5xxx series which is validated by Tables 5 and 6 which show the composition of the various alloys tested in the two series. For the 7xxx series, the amount of Zn, Mg, and Cu added is crucial to the temperature-toughness dependence because in Figure 36 the 7079 sheet alloy has a greater toughness than the 7075, x7275, and 7178 alloys along with the least amount of zinc. Apparently, of all the alloying elements, zinc seems to have the greatest effect upon the toughness-temperature interaction since the total amount of alloy in the above 7xxx series is the same even though the amount of zinc is less (see Table 6). No doubt this is a reflection of the difference in atomic size between zinc and aluminum and also shows again the importance of alloy addition. Consequently, both the amount and demension of the alloying element must be considered when an aluminum alloy is required to have a high fracture resistance in a low temperature range.

In summary, reported results show that:

1. The number and size of matrix precipitates are the key to high yield strength of an aluminum alloy rather than grain boundary precipitates because dislocations are more readily found in the interior of the grains

rather than along the boundaries.

2. Increase in the number and size of grain boundary precipitates decreases the toughness of the alloy.
3. Matrix precipitates have little effect upon toughness as compared to grain boundary precipitates.
4. The amount and size of the precipitates are dependent upon the degree of saturation of the alloy meaning that increasing the amount of alloying elements increases the amount of precipitate and reduces toughness over a large range of temperatures.
5. The size of the alloying element relative to the aluminum atom is critical for cryogenic applications, -320°F. to -423°F. since it causes lattice distortion which impedes dislocation mobility.
6. The PFZ width may or may not have an effect upon fracture resistance but the small amount of data to date seems to indicate that the width of the PFZ has little effect upon toughness.

Thus all of the above must be considered when trying to optimize both strength and toughness. And in the future, present questions such as the effect of the PFZ may be answered giving further information towards attaining the above objectives.

Compositional Effects upon Void Nucleation and Coalescence

For the commercially available aluminum alloys, the predominant mode of fracture is a normal or dimpled rupture. This type of failure is the result of the formation and coalescence of voids formed by the cracking of inclusions or precipitates or by the separation of these particles from the matrix. Microscopically the fracture surface is pocked marked with craters called dimples which reflect the amount of necking of bridges between voids and also the level of toughness where shallow dimples represent low toughness and deep dimples high toughness. From the macroscopic point of view, the fracture surface is normal to the direction of applied stress and appears to be quite similar to the fracture surface of the cleavage mode.

For this particular type of failure of the aluminum alloys, the plane strain condition is critical since plastic deformation is localized around the crack tip, the stress for plastic flow is higher, and the stresses are of a triaxial nature which have a definite effect as will be discussed subsequently (30). Under plain strain conditions, the critical crack opening displacement for instability, V*(c), is reduced for a given notch length and tip radius, and the strain to fracture is less. This is completely opposite to the plane stress condition where the crack displacement can be large and the alloy can exhibit considerable strain and distortion before fracture.

As mentioned before, this type of failure, normal rupture, is composed of void formation and coalescence. And both of these components of failure link up to form the critical crack opening displacement, V*(c), associated with instability. V*(c) is the sum of the displacements at the crack tip derived from void formation and the joining of the voids and tip together by plastic tearing (29). Thus if there is a large strain required to create and coalesce voids, the value of V*(c) will be large and the alloy will be tough.

For void formation the tensile stress in the alloy must be such as to crack the hardening particles or separate them from the matrix. The magnitude of stress to accomplish this is given by (29):

$$\sigma_\rho \cong \frac{M\gamma}{nb}$$

where M is a numerical constant

γ is the work to crack a particle

n is the number of dislocations piled up

b is the Burger's vector[8]

Note that σ_ρ will be lower for larger and more numerous particles and higher for smaller and less numerous particles. (30,31,32), and if cross slipping can occur, a screw dislocation will follow the path of least resistance and dodge obstacles in its way by sliding over the plane on which the obstruction lies, and consequently dislocation pile up and n will be reduced and

σ_ρ increased. Concerning the work of cracking, it is higher for precipitates than for inclusions of the same configuration since the precipitate is strongly bound to the matrix. Thus one can expect aluminum alloys with a considerable amount of inclusions to be less tough than the same alloys without them.

In Figure 37a the magnitude of particle fracture strength is given as a function of the particle size and shows the inverse relationship between the two. Figure 38 shows the results of notched tensile tests on an aluminum 6.1 pct. Si alloy and demonstrates that for a given failure stress the size of the particles fractured is greater than the size of unfractured particles thus validating the theoretical prediction of Figure 37a that increasing particle size reduces the particle fracture strength and one is likely to find a greater number of broken particles. Note that in a typical alloy there will be a distribution of strengthening particle sizes as given by Figure 37b and that the distribution of the average distance, λ, between these particles will be inversely proportional to N, the number of particles having a size either greater or less than the given particle size, p*. Thus if the particles are smaller or larger than p*, the average distance between them will increase, and for particles larger than p*, the particle fracture strength, σ_ρ, will decrease as average spacing between particles increases, Figure 37d. This relationship between σ_ρ and λ can be taken as the fracture strength of a particle at a distance, r, from the crack tip in an aluminum alloy because the order of spacing between the crack tip and the particle is the same as the distance between particles (31) (see Figure 39).

In a notched condition, the longitudinal stress builds up from the yield point at the crack tip to a higher value of $\frac{1}{\beta}(\sigma_y)$[9] at a distance from the tip and then remains constant within the plastic zone (31). In Figure 39 high yield strength means that small particles are cracked and voids are easily nucleated at distances close to the crack tip. This means that the plastic zone size is smaller and that the crack opening displacement, V(c), is less.[10] Thus there will be little plastic deformation in the area of the crack tip, and the alloy can exhibit reduced toughness. Note that this is only the effect of adding elements resulting in precipitates and inclusions and that in plane strain conditions toughness can be even reduced further since there is also a material restraint to the growth and formation of the plastic zone at the head of the crack.

Basically the above discussion has only been concerned with the effects of alloy additions to void formation and consequent toughness. But what about the effects of alloy additions to void coalescence since it is also critical in attaining adequate toughness?

To answer this, the critical crack opening displacement, V*(c), can be used to demonstrate the variability of toughness with composition. The value of V*(c) is made up of two components, crack displacement due to void formation and void coalescence, and it is a suitable measure of toughness[11] since a small value of V*(c) means that the crack has to open very little before instability occurs, and a high value of V*(c) indicates that considerable crack blunting and large plastic deformation is occurring at the crack tip resulting in slow crack growth and high toughness. The void coalescence displacement contribution to V*(c) is a measure of the work necessary to join the first void ahead of the crack and the crack tip. If there is considerable displacement, the material between the first void and the tip will stretch and yield and the void will appear as a deep dimple on the fracture surface. Thus the alloy will be relatively tough. And if the displacement is small the opposite is true. Usually this mechanism of failure occurs by a shearing process (33) and is proportional to $\varepsilon_f'(c)\rho$ where ρ is the crack tip radius and $\varepsilon_f'(c)$ is the ductile fracture strain which is given by:

$$\varepsilon_f'(c) = \frac{(1-n)\,\ell n(\ell_0/2b_0)}{\sin h\ [(1-n)(\sigma_{xx}+\sigma_{yy})\ /\ 2\bar{\sigma}/\sqrt{3}}$$

where b_0 is the initial radius of a cylindrical hole

ℓ_0 is the spacing between holes

n is the strain hardening exponent whose stress-strain curve is $\sigma_0 \varepsilon^n$

and

$\sigma_{xx}/\overline{\sigma}$ and $\sigma_{yy}/\overline{\sigma}$ are constant transverse stress ratios under a plain strain condition.

Thus if ℓ_0 is taken as the spacing between the first void and the crack tip, the fracture strain and the crack opening displacement will decrease as the void spacing, $\ell_0 = r$, decreases resulting in a lower work required for void coalescence and consequent reduced toughness. Note also that the fracture strain will be reduced if the density of voids is increased which is due to increasing the number and size of hardening inclusions or precipitates. Thus the same parameters, σ_y and σ_ρ, which governed void formation also affect the ability to coalesce voids in the same manner; that is, increasing the yield strength or reducing the particle fracture strength, reduces the work to coalesce the voids formed by the particles and reduces toughness.

For the comparison of the 2xxx and 7xxx series alloys, the reason that the 7xxx series has a higher toughness than the 2xxx series is the increased size of the precipitates in the aluminum copper series. In the light of the above discussion, the larger precipitates increase dislocation pile up, reduce the particle fracture strength, σ_ρ, and increase the yield strength. Consequently void formation and coalescence is easily accomplished without a large fracture strain, and a reduced toughness results relative to that of the 7xxx series. Within each of the series, the toughness decreases as the amount of alloying elements increases because the amount of precipitate increases. This again results in the same mechanism of reduced dislocation mobility, etc. and diminishes fracture resistance. Under cryogenic conditions the 5xxx and some of the 2xxx and 6xxx series alloys have a reduced toughness at temperatures below-320°F. and 7xxx series alloys below ambient temperature. The reason for this is that cross slipping of dislocations can no longer occur and pile up intensifies and reduces the particle cracking strength, σ_{ρ}, which leads to increased void occurrence and reduced toughness. At higher temperatures these alloys allow for cross slipping of dislocations because the atoms of the inclusions or precipitates are vibrating and moving in a random fashion allowing dislocation movement. The result is that the particle fracture strength is higher and the toughness greater. Note that this is an analysis of an alloy of given composition within a series under lower temperatures. Naturally, alloys of the same series will have different toughness-temperature characteristics because their compositions will differ.

Thus the above discussion illustrates how the composition of an alloy based upon the amount and size of inclusions or precipitates which are formed by alloying additions can affect the fracture resistance of aluminum alloys by influencing the ease at which void formation and coalescence can occur.

Processing Effects

In trying to determine the best strength-toughness combinations, a metallurgist has several methods available which he can use. These methods are basically changing the composition, employing heat treatment, or simply working the alloy to achieve desired results. But as discussed previously, just adding more alloying elements tends to increase strength and decrease toughness due to the increased number and size of zones and/or precipitates which are created during heat treatment or working. To achieve good toughness and strength both compositional and processing effects must be considered. But for the present discussion only the effects of processing will be disclosed.

The processes available are annealing, strain aging, solution heat treatment, quenching, and natural or artificial aging. These treatments along with their combinations have been given special designations which are listed in Tables 7 and 8, and a distinction is made between heat treatable (2xxx, 6xxx,7xxx) and non-heat treatable alloys (1xxx,3xxx,4xxx,5xxx). The reason for this is the fact that the non-heat treatable alloys do not respond to the solution and aging treatments of the heat treatable series; that is, their mechanical properties are not enchaced by these processes.

Note in Table 7 for the strain hardened alloys, the variation in mechanical properties is dependent upon the working and/or the heating of the alloy at low temperatures or to a partial anneal. But in Table 8 the variation of properties of the 2xxx, 6xxx, and 7xxx series alloys depends upon heating practices such as annealing or solution treatment, the amount of cold work, and the type of aging. Thus these processes are quite different for the two types of alloys and have a pronounced effect upon both strength and toughness as will be discussed subsequently.

In order to understand what the effects of these various processes are, one must examine each process separately with regard to how it affects the microstructure of the heat treatable or non-heat treatable alloys.

For the non-heat treatable alloys, there are basically three types of processes that can be used: the as fabricated temper, F, in which the wrought alloy is worked without considering the degree of working, the annealed temper, O, in which the alloy is recrystallized and no effects of working remain, and the strain hardened tempers, H1 through H3, in which controlled amounts of working are used to obtain desired strengths. Of the commercial wrought alloys available, the F temper is seldom used and only the O and H tempers are employed depending upon the desired mechanical properties.

In the strain hardened tempers,[12] the alloy is stretched varying amounts, and consequently the yield strength varies directly with the amount of deformation since increased working means an increased fragmentation of grains and density of dislocations. These increased dislocations cause immobility and interactive stresses which result in a higher yield but lower toughness. Thus if the alloy is worked extensively,[13] the toughness will decrease due to increasing the number of dislocations which can pile up on dispersed hardening particles and/or inclusions and reduce the fracture stress.

To alleviate the effects of strain hardening, annealing can be used to recrystallize the alloy and remove all dislocation concentrations either within the grain or along their boundaries. This is done by heating the alloy to high temperatures (450°F. to 775°F.) for a length of time depending upon the composition and the degree of prior cold or hot working. Usually greater degrees of cold work result in a lower temperature and less time of annealing for recrystallization, and hot working and greater composition result in an opposite effect. Thus if the alloy is quenched from the annealed state, one can expect higher toughness and lower strength than the strain hardened condition.

This is validated by Figure 40 which shows the UPE and yield strength as a function of temper. Note the annealed state gives the highest toughness and lowest strength and H38 just the opposite with H38 being representative of 75 pct. reduction in area of the alloy and a high degree of strain hardening. Also from this figure one can see that increasing the amount of work decreases the toughness by examining the UPE for the intermediate tempers.

Some work has been done examining various alloys of the 5xxx series for their variation in toughness with decreasing temperatures(34 through 38). Generally the result of all this work is the same as shown in Figure 40; namely, the toughness of the annealed or slightly worked alloy is greater than the highly worked or hardened alloy over a range of temperatures. Table 9 illustrates this using the notch yield ratio as a measure of toughness for several 5xxx alloys for temperatures down to -452°F. And Table 10 shows the same relationship for different alloys of the 5xxx series but using the propagation energy as a measure. Based on both of these tabulated measures of toughness, the fracture resistance of the 5xxx series alloys is very high and is about the same at -452°F. as it is at room temperature indicating good application for designs where fracture is critical rather than strength.

Some other work has been published (38,39) on the toughness of the 5xxx series alloys, but the results are erroneous because the inconsistent measure of toughness, the notched to unnotched tensile strength ratio, is used. The results can be seen in Figure 41 and show that for the same alloy the fracture resistance over a range of temperatures can be greater for the strain hardened rather than the annealed temper. This is invalid not only because of the use of the notched to unnotched ratio but also for the fact that the strain hardened state has been previously shown to be less tough than the annealed condition because of the increased number of dislocations and possible dispersed particles. Thus, in general, the non-heat treatable alloys will exhibit a decrease in toughness as the amount of cold working of the alloy increases but will exhibit a greater yield strength. And this can be reversed if the alloy is subjected to a partial or full anneal after strain hardening.

Concerning the heat treatable alloys, working of the alloy can be performed either prior to (T3 or T8 temper) or following (T9 temper) aging. Usually work prior to either natural or artificial aging of the 2xxx series alloys has a pronounced effect because precipitation hardening is accelerated due to the increased number of nucleation sites generated by the more numerous dislocations. Thus for the 2024 and 2219 alloys, the yield strength increases with cold working before aging by increasing the amount and reducing the size of the transition precipitates which result during aging (40). But for this process and these alloys, toughness is sacrificed.

In the 2xxx series alloys there is one exception, 2021. The alloy 2021 will have reduced yield strength with prior cold work because working the alloy coarsens the coherent transition particles which results in a lower hardening ability (40). But this inherently means an increased toughness which may not be so detrimental.

For the remainder of the heat treatable alloys, only the 2xxx series shows a detectable response to cold working because the alloys of the 7xxx series fail to show any effect of working due to their strengthening structure which mainly consists of zones rather than precipitates (40).

A recent investigation has been made into the relationship between the strain hardening exponent and toughness (41) of the heat treatable alloys. Note that the strain hardening exponent, n, as given in the equation for true stress and strain, $\sigma = \sigma_0 \varepsilon^n$, does not vary with the amount of cold work but rather with the initial yield strength of the alloy. This is seen in Figure 42 which indicates a constant slope, n,[14] as the reduction in thickness increases. Note the decrease in slope of these lines as the initial yield strength increases. Thus the strain hardening exponent only relates to the initial yield strength and not to the degree of working.

From Figure 42 there appears to be a correlation between the strain hardening exponent and the yield strength, and to try to explain the two alloy properties as being separate influences on alloy fracture resistance may not be valid. But work has been done with this objective in mind(41).

Specifically, the results of the investigation are illustrated in Figure 43 and show that the fracture toughness of a 2024-T62 alloy of 49 ksi yield strength is greater than that of a 2024-T851 alloy with a yield strength of 65 ksi over a range of thicknesses. For the tougher alloy the strain hardening exponent is given as 0.09, and for the less ductile alloy, the exponent is 0.045. Thus this implies that a higher exponent for a particular alloy means a higher toughness under both plane stress and strain conditions. But this may be masked by the effect of the lower yield strength of the tougher alloy.

Previously, the strain hardening exponent was found to have an effect upon the ductile fracture strain, ε_f', (see previous section) in which higher values meant an increasing crack opening displacement and toughness. Physically this can be interpreted as being the amount of work necessary to stretch and pull the material between voids of an alloy which fractures by void formation and coalescence. If the strain hardening exponent is low, only a small increase in stress is necessary to plastically pull or tear apart the bridge material between the crack tip and the first void as the alloy is placed under tension. On the other hand, if the exponent is large, more stress is required to tear the bridge material as the material is stretched more and more. Thus the advantage of having a material with a good strain hardening ability is evident if the alloy fails in the particular type of mode discussed above. Also the yield strength must be considered, for a material with a high yield strength usually will have a low toughness irregardless of what the strain hardening exponent might be because the alloy will fracture even before the yield stress or hardening region is reached.

Besides showing some correlation of toughness of the heat treatable alloys to the strain hardening exponent, recent investigations show that various heat treatments can alter the strength and toughness relationships for the high strength alloys. With these treatments there appears to be more flexibility in the final mechanical properties of an alloy as is indicated by the number of tempers listed in Table 8. Basically these treatments consist of a solution treatment and/or some form of aging.

The purpose of solution heat treating is to raise the temperature of the solid solution just below melting so that the greatest degree of saturation by the hardening elements of Zn, Si, Cu, and Mg is obtained. Consequently the strength will be higher, and the amount of precipitation after quenching can be controlled more readily so that the final desirable properties are obtained.

Usually the steps in processing a heat treatable alloy consist of solution heat treatment, quenching, and some form of aging. For this series of procedures, the quenching should be rapid so that the solid solution retains its supersaturated state down to room temperature from which the alloy can age naturally for a short period of time or else the alloy can be subjected to a moderate temperature rise within a range of 200 to 400°F. for a period of time.

This latter form of aging is called artificial aging and produces the highest yield strengths and lowest toughness. The reason for this is simple since in the artificially aged state a large number of precipitates is formed which inhibit dislocation movement and cause dislocation pile up as the alloy is stressed in the working condition after treatment.

The temper designations for the solution heat treatments employing wither natural or artificial aging are T4 and T6 respectively. And the T6 temper is known as the hardest temper or the heating process which can produce the highest strength alloys with the lowest toughness.[15] The T4 temper generally yields the opposite effect which can be seen in Figure 44 for a 2014 alloy. For the 2014 alloy, if the aging continues beyond the T6 condition, the alloy becomes soft and its yield strength is lowered as its toughness is increased. But this increase in toughness is not up to the level of the underaged condition (20,42).

This is the result of annihilation of the coherent phases which provide the hardening effects and their replacement with equilibrium precipitates which are formed by solutes leaving the solid solution as aging continues. Thus if tough alloys are desired, the best method of treatment after solution heating appears to be either underaging or severe overaging.

If the alloy is aged at higher temperatures and longer periods of time (severe overaging), the toughness at room temperature will increase (44). This has been indicated by the measurement of the UPE for a 2020-T6 alloy at temperatures ranging between 400 to 600°F. for $\frac{1}{2}$ to 24 hours. The results are shown in Table 11 and Figure 45, and note in the figure that the toughness for the various 2xxx series alloys increases above the room temperature toughness as the temperature increases. This could possibly mean that the alloy is coming close to the annealed state where recrystallization occurs with a reduced number of dislocations.

Another way in which the effects of aging upon precipitation can be augmented is by a quench interruption. As the solid solution is being cooled from the solution heat treatment temperature, the cooling is held constant for a period of time and then quenching resumes. The result of this interrupted quench is the control of the number and size of grain boundary precipitates in the 7xxx series alloys. And this has been shown to have a definite effect upon toughness in the above series.

Figure 46 shows that increasing the interruption time at 330°F. increases the PFZ width and also the number of boundary precipitates. Thus to reduce the number of boundary precipitates formed during final aging, the quench interruption treatment should not be employed unless increasing the PFZ width is found to be beneficial for increasing toughness (24).

But this has yet to be proved.

Thus aging of a heat treatable aluminum alloy plays a major role in the fracture resistance of the alloy and can be used in combination with other processes to achieve desired results as will be seen in the following section.

Note that in Table 8 of the tempers for the heat treatable alloys the number of treatments that can be performed varies from just the T4 and T6 tempers. Cold work can be applied before or after the two types of aging and an alloy can be stress relieved to obtain desired results. Note also that the two forms of aging do not have to follow solution heat treatment (T1 and T5 tempers) and that cold work does not necessarily have to accompany solution heat treatment either (T10 temper). The effects of these various treatments on toughness and strength have been reported and the conclusions drawn are worthwhile discussing.

Generally for the high strength alloys of the 7xxx series, the alloys employing a T73 or T76 temper have a lower toughness than those alloys with a T6 type temper. The reason for this is that even though the T73 and T76 tempers imply a higher toughness due to overaging after solution heat treatment, their toughness is reduced due to the amount of cold work that is performed which is indicated by the second digits of the temper, 3 and 6 (3).

For the 2xxx series, the T3 or T4 alloys have a higher toughness than those in the T6 or T8 type tempers (3,8) because the artificial aging sequence of the latter two tempers results in a higher yield strength as explained before. Even though the cold working prior to natural aging in the T3 temper results in a substantial yield strength, the increase is not as great as the T6 or T8 temper, and thus the toughness is not reduced as much.

There is even a difference in toughness resulting from the use of the T3 or T4 type temper in the 2xxx series alloys. The cause of this is the intermediate cold working of the T3 tempered alloy which increases the yield strength. This difference can be seen in Tables 12 and 13 for a 2024 alloy if one examines the ratio of the notched tensile strength to the yield strength. Note that this ratio is larger for the T4 temper over a wide range of temperatures which is reasonable in the light of the prior discussion on the effects of working and that the effect of the difference in specimen thickness (0.025 in. to 0.032 in.) is negligible when considering the relative values of the ratios.

The degree of working the alloy as indicated by the temper also has a definite influence on the outcome of an alloy's toughness. For example, Tables 14 and 15 show the results of tensile tests taken on notched specimens of a 2219 alloy in the T81 and T87 temper. The T81 temper indicates a minimal amount of strain aging between solution heat treatment and artificial aging while the T87 temper indicates a substantial degree of strain aging between the same two treatments. Consequently one could expect a higher yield strength and lower toughness for the latter temper, and this is actually what appears for a range of temperatures in Tables 14 and 15 if one examines the notch tensile strength to yield ratio.

In summary, the application of cold work and/or artificial aging to a heat treatable alloy will increase the yield strength and decrease the toughness, and if no cold work is performed and only natural aging is employed, the opposite will be true. This does not necessarily apply to the 7xxx series alloys because the time for these alloys to form strengthening precipitates is long and natural aging periods are inefficient. Thus only the tempers consisting of some form of artificial aging are used on the 7xxx series alloys.

In trying to achieve both maximum strength and toughness, investigations have been made upon the combined effects of compositional changes and variable heat treatments. In particular, the effect of varying the magnesium content and temper of a non-heat treatable aluminum - magnesium - manganese alloy has been published and is shown in Figure 47. Naturally the O temper will have the highest toughness and lowest yield strength because of the lack of lattice imperfections or dislocations. But note that there is a maximum value of toughness for a given magnesium content indicating that the increase in alloying element causes a decrease in the energy to propagate a crack. This might be due to the relative ease in forming and coalescing voids due to the decreased spacing between hardening elements or else dislocation mobility is greatly reduced as may be the case in the H34 temper. But this illustrates that both composition and heat treatment can be used together to optimize the yield-toughness requirement.

Since combining changes in composition and heat treatment or cold working seems to be a viable direction in which to proceed in order to optimize toughness and yield strength, research has been done in this area but little has resulted so far except for the results published on a 7175 alloy. These results can be seen in Figure 48 and show that the 7175 alloy in the T66 or T736 temper[16] has a greater toughness and strength over the conventionally used 7075 alloy in either the T6 or T73 temper (3).

Besides varying the composition and heat treatment, other miscellaneous processes have recently evolved which show a slight increase in yield strength along with a substantial increase in toughness. Specifically these processes are the premium extrusion of the 7075 alloy in the T73510 and T76510 tempers and Alcoa Process 417. For the premium extrusion, the graphical results are given in Figure 49 and demonstrate that this process increases toughness substantially over the conventional extrusion of the 7075 alloy. The results of the Alcoa

417 Process are shown in Figure 50 and again show the same achievement of increasing toughness but for a 2024-T851 alloy. The Alcoa 417 Process appears to be propriety because no information is available concerning the actual metallurgical steps in the new process (3).

Also what has been found to improve toughness is the use of pre-forging of plate before rolling (3). Notch bend tests of specimens taken from various orientations within a pre-forged and standard rolled plate have been performed (45,46) and show that the plane strain stress intensity factor is greater for the pre-forged plate rather than the standard rolled. This published result can be seen in Table 16 or in Figures 51 through 53. And the reason for the increased toughness is speculated to be due to the texture and the smaller and more uniform grain structure in the pre-forged plate (45). This texturing effect will be discussed in a following section on anisotropy and its relation to fracture resistance of aluminum alloys.

From the above discussion the role of processing in the variability of the fracture resistance of aluminum alloys is great and can be summarized as follows:

1. For the non-heat treatable alloys strain aging or work hardening will decrease toughness but increase the yield strength because the number of lattice distortions and dislocations increase.

2. Annealing will have the opposite effect of strain hardening because alloy recrystallization occurs with the removal of dislocation concentrations.

3. For the heat treatable alloys, cold working prior to or following aging will increase the yield strength but decrease toughness. And this process has more effect upon the 2xxx series alloys than any alloy in the 7xxx series because the hardening structure of the former series is due to precipitation rather than zone formation in the latter series. The increased number of dislocations offer more nucleation sites for precipitation in the 2xxx series alloys.

4. From some recent work done on the toughness-thickness relationship for two 2xxx series alloys, the strain hardening exponent might be an indication of toughness. A higher value of the exponent implies a tough alloy and a low value, a low toughness under both plane stress and strain conditions.

5. Artificial aging of an alloy up to peak strength yields the lowest toughness and natural aging yields just the contrary. If the alloy is aged beyond the peak strength, the yield strength decreases and the toughness increases but not up to the level of underaging.

6. Severe overaging can result in greater toughness of an alloy at room temperature.

7. Control of an interrupted quench is a means of controlling the amount of grain boundary precipitate in a 7xxx series alloy. And this has been shown to be related to toughness.

8. Combinations of cold working and artificial aging yield the lowest toughness and highest yield strengths. And alloys in the annealed state or only naturally aged yield lower strengths but the greatest toughness. Employing cold work with natural aging results in intermediate levels of yield strengths and toughness.

9. Combining compositional changes and various processes appears to be the best direction in optimizing strength and toughness. And this has been done for a 7175 alloy in the T66 and T736 tempers.

10. Various mechanical processes have shown that their use can increase toughness:

    a.) Premium extrusion of 7075 in specialized tempers has increased toughness and strength over conventional extrusions of the same alloy in the same temper.

    b.) Alcoa Process 417 increases toughness and strength for a 2024-T851 plate alloy.

    c.) Pre-forging and then rolling increases toughness over standard rolling of plate because of textural differences.

Thus in seeking to optimize yield strength and toughness, the metallurgist has a variety of options available when considering the effects of compositional changes, heat treatment, and mechanical working.

Fracture Resistance of Alloy Castings

In the prior discussion on compositional and processing effects, the investigation was centered on the fracrure resistance of wrought alloys, and no mention was made about this property of alloy castings. But their fracture resistance must be considered because they are frequently used as structural members which must sustain a load under a variety of conditions. The reason for their increased popularity is that they can be fabricated in a variety of configurations which are impossible or impractical in the wrought condition. Even though the castings have this ability to take on any shape, they do not possess the same fracture resistance as the wrought alloys.

This is made evident in Figure 54 which shows the toughness and yield strength at four temperatures of several castings. Note that all the differently fabricated alloys fall below the band for the wrought alloys at all temperatures and only the premium strength castings[17] are close to the strength-toughness relationships of the wrought alloys. The difference in strength and toughness of the two types of alloys is based upon their microstructure with wrought alloys having a solid solution matrix composed of grains and cast alloys, a matrix composed of solid solution dendrites[18] with complex binary and ternary eutectics[19] filling the spaces between the dendrites. Also for the castings the amount of hardening elements (Si, Cu, Mg, Zn) is greater than in the wrought alloys with the exception of the silicon bearing alloys where the amount of alloying element is about the same. This implies a reduced toughness for the cast alloys if one accepts the conclusion drawn from the discussion on compositional effects.

Even though the cast alloys don't meet up to the toughness and strength levels of the wrought alloys, they appear to show the same effects of compositional and processing variations on fracture resistance as for the wrought alloys (47).

The effects of both composition and processing variables can be seen in Tables 17 through 19 and Figures 55 and 56. Table 17 gives the tabulated compositions of the 100, 200, 300, and 600 series castings along with their manner of fabrication whether it be sand, permanent mold, or premium cast. Note that the alloying percentages of the cast alloys are greater than those of the wrought alloys of Tables 20 and 21 and that the same heat treatments can be used for both types of alloy as indicated by the tempers in Table 17.

With the help of these charted compositions and the tabulated results (Table 19) of notched tensile tests of specimens taken from permanent mold castings, one can see that increasing the composition does decrease the toughness by examining the notch yield ratio for the various alloys. For example, alloys 359-T62 and 354-T62 have the lowest notch toughness and highest strengths, and alloys A344-F and A356-T7 have the highest toughness but lowest yield strength with the latter two alloys using the smallest percentage of alloying elements and the former two, the largest percentage of the entire 3xx series tested. Thus this seems to indicate the same trend in the effects of compositional changes as seen in the reported test results for the wrought alloys.

Concerning heat treatments, the cast alloys can be subjected to solutionizing, annealing, and artificial or natural aging just as the wrought alloys with basically the same results. This can be verified by examining the 356-T6 or T7 alloy in the sand or permanent mold casting process (Tables 18 and 19). For both methods of fabrication, the 356 alloy shows a higher notch toughness and lower yield strength for the solution heat treated and overaged temper (T7) than for the T6 temper which indicates solutionizing and artificial aging up to peak strength. This is the same result which occurs when both tempers are applied to a wrought alloy.

Even if cold work is applied between solution heat treating and aging of a cast alloy, the result is the same as for a wrought alloy. This is shown in the test results (Table 19) for an A356 and A357 alloy fabricated in the permanent mold or premium cast state and using the T61 and T62 tempers. From Table 19 the T62 temper of both alloys in both casting forms derives a higher yield strength and lower toughness due to the increased amount of work over the T61 temper. And again this result can be similarly found for the wrought alloys (note results of Tables 13 and 14).

Another trend of parallelism between the two types of alloys is the effect of mechanical processing. Just as pre-forging before rolling of plate increases toughness of the wrought alloys, premium or permanent mold casting does the same for the cast alloys. The alloy, A356-T61, which exhibits about the same yield strength in the premium cast state as in the permanent mold condition has a higher notch toughness due to the controlled foundry practices of premium casting. If one were to rank the different type of castings according to the increase in fracture resistance, sand casting would be first and premium casting last. And this illustrates the role of mechanical processing even in cast alloys.

One other similarity between the two available types of alloys is that additions of zinc to both produce high strength but low toughness with decreasing temperatures (see Figures 29 and 56). This, no doubt, is due to the same response to heat treatment for both the cast and wrought alloys.

With all these similarities between the compositional and processing variables in the cast and wrought alloys, there are a few differences.

One of the differences between the behavior of the two types of alloys in the presence of a stress raiser is the variation of toughness with temperature for the 1xx series cast-ings and the 2xxx series wrought alloys. As seen in Figure 56, the

1xx series alloys lose their notch toughness as the temperature falls[20] while in Figure 34 the notch toughness remains about the same for the majority of the 2xxx series alloys even though both series in the cast and wrought condition have the same principle hardeners of Cu and Mg. Another difference between the two series is that the 1xx alloys generally have a lower yield strength relative to the other castings while the 2xxx alloys are considered to be high strength alloys.

For the cast alloys, the 3xx series has a substantial yield strength relative to the 1xx and 2xx series in the sand cast form. Thus one could term the alloys in this series as being high strength cast alloys using silicon as the main hardener. In the wrought alloys, the 4xxx series also uses silicon for its source of strength, but the yield strength derived by strain hardening is in the medium range relative to the other wrought alloys. Thus the difference between the two series in the available conditions lies in their relative rating based upon their yield strengths.

Note that the toughness of the 3xx series cast alloys maintain approximately their room temperature toughness down to -320°F. (see Figure 56). And if the 4xxx series behaves under the presence of a stress raiser like the other medium strength wrought alloys (5xxx series),[21] the silicon bearing wrought alloys will not decrease in toughness with temperature either. Thus in this way the 3xx series and the 4xxx series of the cast and wrought alloys respectively are alike.

Basically, these are the differences and similarities between the two available types of aluminum alloys. And the reason for the parallelism appears to be in their similar response to heating and mechanical treatments while their differences must be derived by the divergence in microstructure.

Concerning the cast alloys themselves, the 1xx, 2xx, and 6xx series alloys will exhibit a lowering of toughness as the temperature drops while the 3xx series alloys will have an invariant toughness. And the 3xx and 6xx series alloys will have the highest yield strengths but possibly the lowest toughness depending on the casting process and heat treatment. Note also that permanent mold and premium strength casting improves toughness while not reducing strength and that sand casting tends to produce the lowest toughness due to the coarse grain structure formed during the process.

Quality Control

To attain a substantial level of toughness and strength, the amount of porosity and the number of oxide inclusions and intermetallic compounds should be controlled for both the production of wrought and cast alloys. And any of the above processing variables can reduce fracture resistance substantially if present in large amounts.

Porosity is derived by the amount of hydrogen that is dissolved in the liquid metal. And if the concentration of hydrogen reaches a threshhold value, the porosity will be greatly increased. At values less than the threshhold value the resulting porosity can be eliminated by mechanical working if the alloy is to be in the wrought condition (48). Otherwise, large gas contents result in an inevitable porosity which will produce microscopic notch effects that can lead to reduced toughness.

The strength reducing action of the pores can be seen in Figure 57. And note that initially the pore assumes a circular shape, but as the alloy is subjected to an applied stress, the pore will try to assimilate a lenticular form. The degree of sharpness of the newly formed pore will depend upon the migration of atoms from the pore surface to the corner or tip or the vacancies from the tip to the other portions of material surrounding the pore. If the migration of atoms to the tip is high, the tip will be blunted and the pore cannot offer any local stress multiplication (49). But if the vacancy migration from the tip is greater than the atom accumulation, a stress concentration will occur, and the continued nucleation of vacancies will form a void which reduces the alloy's fracture strength as it coalesces with other voids. Thus the action of the pore is similar to the action of adding hardening elements which create voids by their fracture due to the stress concentration built up by dislocation pile up.

To illustrate the effect of porosity on toughness, Figure 58 shows the variation of impact energy of an Izod type test with hydrogen level. Note that the impact energy is not a valid measure of toughness but can be used here to demonstrate the effect porosity can have. Thus the figure shows that increasing hydrogen level up to 0.18 cc/100 g of alloy decreases toughness for both the 2 and 3 inch plate, as expected. But for hydrogen levels greater than 0.18 cc/100 g (threshhold value), the toughness maintains a constant low level indicating that increased porosity cannot reduce toughness anymore because the alloy has achieved essentially a brittle state. Also in this figure, the threshhold value is the same for both plates illustrating that rolling to greater thinness cannot reduce the notch effects of the pores. Thus the only way to prevent reduced toughness due to porosity is to maintain hydrogen levels below the threshhold value by controlling the casting procedure which can be seen in Table 22.[22]

The second processing variable mentioned previously is the amount of oxide inclusions. These defects are formed by an oxide film which develops on the surface of the molten metal as it remains in the holding furnace and passes into ingot molds. This oxide film is submerged in the metal as the molten mass flows turbulently into the ingots and forms the inclusions which can reduce toughness by creating lattice defects in the final crystalline state. These defects reduce dislocation mobility and can reduce toughness if present in substantial amounts.[23] As in the case for controlling porosity, the amount of defects due to oxide inclusions can be reduced by controlling the casting operation. If the molten metal is poured through a glass screen into the ingot, the number of inclusions will be minimal (48). Note how much this filtering reduces the amount of inclusions by examining the last column of Table 22 for the appropriate casting conditions.

The third processing variable, namely, the amount and distribution of intermetallic compounds has the same effect as increasing porosity or oxide inclusions. That is, increasing their frequency in the matrix results in reduced toughness. Evidence of this can be seen in Tables 23 and 24 where the notch-yield ratio of the x7275 alloy is greater than the same ratio for the 7075 alloy over a range of temperatures. Note that x7275 is a pure form of 7075 with the amount of iron and silicon reduced as is illustrated in Table 25. Iron and silicon are the main elements which compose the intermetallic compounds, and thus reducing their amount will generally increase toughness assuming the alloy tempers to be the same. This conclusion can be extrapolated to other series of wrought alloys as can be verified in the results of notched tensile tests taken on a few of the 5xxx series alloys.

For this latter series, the alloys 5083 and 5456 can be compared. Alloy 5083 has about the same alloying content as 5456 but lower amounts of iron and silicon and a harder temper (H38 as compared to H343) than the 5456 alloy. The result is as expected; the 5083 alloy has a higher toughness as measured by the notched to unnotched tensile ratio over a range of temperatures. This result is shown previously in Figure 35 and the outcome of the testing may be questionable on the basis of the appropriate measure of toughness that is used but should suffice for a comparison since the results are parallel to those of adding the same elements to the x7275 alloy.

The mechanism of failure for the alloys with a large amount of intermetallic compounds is the same as for those alloys strengthened by hardening elements. The compounds act as nucleation sites for voids since the dislocations which are blocked at compounds increase the local stress which can, in turn, crack the compound particle or separate the particle-matrix interface. These compounds if present in substantial numbers can also reduce not only the work necessary to create voids but also to coalesce them. Just as increasing the amount of hardening elements reduces the fracture strain and crack tip opening displacement, the addition of intermetallic compounds will naturally lead to relative ease of tearing the bridge material between the crack tip and voids within the plastic zone. This latter mechanism of failure is also dependent upon how many voids are nucleated initially. If the number of voids nucleated is large due to the presence of numerous intermetallic compounds, the energy to coalesce voids will be small and the fracture strain less because there will be less bridge material to tear between the voids.

For alloys which fail in the ductile rupture mode, the nucleated voids can coalesce to form dimples which are seen microscopically on the fracture surface. The size of these dimples indicates the toughness of an alloy. If the dimple is deep and wide, a substantial degree of material tearing is occurring before the alloy fractures, and consequently the fracture resistance is adequate or at least greater than the fracture resistance of an alloy characterized by shallow and narrow dimples. Shallow and small dimples indicate a reduced amount of work necessary to tear the bridge material.

Metallographic studies have shown this above analysis to be true and indicate a dimple size for a high purity 7075-T6 alloy to be 0.4×10^{-4} inches as compared to the commercial 7075-T6 alloy which has a dimple size of 0.2×10^{-4} inches (50).

The corresponding toughness for both alloy types is given in Figure 59 for various specimen orientations, and indeed the toughness of the pure alloy is greater than the commercial alloy. Thus the correlation between dimple size and toughness appears to be valid along with the explanation of the probable effects of increasing compound additions.

The intermetallic compounds not only have an effect upon the fracture resistance of wrought alloys but also cast alloys. Iron in combination with aluminum will form hard particles which precipitate between dendrites and dendrite arms causing embrittlement and reduced toughness (51). The only way to alleviate this problem is to maintain quality control, for heat treatment cannot dissolve the iron - aluminum compounds(51). The action of these impurity elements in the cast alloys is believed to be very similar to that of the iron and silicon impurities in the wrought alloys; that is, the iron - aluminum compounds of the castings promote fracture by confining the fracture to areas where the concentration of impurities is the greatest.

In searching for better alloys to resist fracture, quality control during the foundry process plays a definite role. If the porosity of an alloy becomes considerable due to uncontrolled hydrogen levels in the molten metal, the toughness of the resulting alloy will be less. If careful control is not used in screening the molten metal before pouring into the ingots for cooling, oxide inclusions will result which can reduce toughness. And finally, if more precision is used in the amount of alloying additions, especially iron and silicon, the toughness of the final product can be increased by minimizing the amount of elements that form intermetallic compounds. The reason that the above processing variables reduce toughness if present in substantial amounts is that they act as internal notches which produce local stress concentrations. This increase in local stress allows for relative ease of void nucleation which in turn influences the final coalescence of the voids to fracture.

Directional Properties of Alloy Plate

One of the critical properties of alloy plate is the anisotropy of its fracture resistance. In the as cast condition, the aluminum alloy is basically isotropic; that is, its fracture resistance is the same in the three principal directions. But as mechanical working is performed such as rolling, pressing, or extruding, there is a reorientation of the grains according to the direction of working. This reorientation of the grains is the basic reason for the anisotropy of plate toughness as will be discussed subsequently.

To identify the anisotropy of alloy plate, three principal directions have been categorized according to the geometry and direction of deformation of the plate. These three directions are called the longitudinal, long-transverse, and short transverse directions respectively.

To assist in determining the relative positions of these directions, Figure 60 shows that the longitudinal direction refers to a crack propagation system in which the applied tensile stress is in the direction of rolling and the crack propagates at right angles to the rolling direction. The long-transverse direction differs from the longitudinal in that the applied stress is normal to the rolling direction but is similar with respect to the crack propagation direction which is essentially against the grain. Note that the short transverse direction differs from both of the above directions with respect to the direction of applied stress and crack propagation.

There has been some refinement in defining the various crack propagation systems of an alloy plate (45). These systems are given in Figure 61 which identifies each system by two indices. The first index gives the direction of the normal to the crack plane and the second, the direction of crack propagation whether it be R, the rolling direction, W, the direction of width, or T, the direction of thickness. Note that the previous directions designated as longitudinal, long-transverse, and short transverse can be identified as the RW, WR, or TR directions respectively.

What truly makes these crack propagation systems different is the grain direction. The mechanical working of the alloy whether it be any of those previously mentioned causes flat, elongated grains to be formed in the direction of working. If a section were cut parallel to the rolling plane, microscopically elongated grains would appear which are actually platelets that are stacked one on top of the other. In section these grains would appear as lines which have grain boundaries lying parallel to the rolling surface and short boundaries which are perpendicular to this surface and are not aligned from crystal to crystal (52). Typical micrographs of the three principal sections can be seen in Figure 62 and note that the transverse section shows the least amount of grain boundary which can inhibit crack propagation. Thus what really distinguishes the three propagation systems is the manner in which the crack moves relative to the grain structure.

For the longitudinal specimen, the crack has to separate and tear elongated grains; whereas, for the long-transverse direction, the crack will have a tendency to separate the grain boundaries in the direction of grain elongation. The short transverse direction is quite different in its manner of crack movement from the other two directions in that the propagating crack will separate the granular platelets themselves like splitting wood in the direction of the grain. In other words, what is critical for the various directions is the amount of resolved normal stress relative to the face of the grains (58,60). If the grain boundaries are parallel to the applied stress such as in a longitudinal specimen, the toughness will be high because the fracture is incurred by tangential stresses rather than normal stresses (58). But if the grain boundaries are normal to the applied stress, the toughness will be low because the full normal stress which is mobilized on the individual grains can pull the platelets apart with relative ease in the presence of a notch. Thus one can expect the toughness to be the least in the short transverse direction.

With these various means of designating the propagation systems, tabulated results of toughness testing have been recorded accordingly (3,17,25,45,53-55). Tables 3,9,12-16, and 26-28 show these results and rank toughness according to increasing values as given by the short transverse, long-transverse, and longitudinal directions respectively. One exception to this ranking is the data reported in Tables 26 and 28 which show the plane strain stress intensity factor and the notch-yield ratio for the long-transverse direction being comparable to the same measures for the longitudinal direction in a rolled plate. On the other hand, extruding an alloy appears to make the toughness in the two above directions diverse but equalizes the toughness in the two transverse directions (see Table 26). Apparently this difference in rolling and extruding is based upon the final grain structure which differs.

For the extruded bar, the final grain structure consists of a fiber texture[24] with a (111)[25] direction parallel to the product axis and random crystal directions perpendicular to this axis (56). The rolled plate differs from this in that the texture is composed of three ideal textures (110)(112), (112)(111), and (123)(211). But the texture of a rolled plate can be described by the (111) planes which have a variable orientation as disclosed by x-ray analysis (56).

A typical x-ray pole figure[26] of two alloys is shown in Figure 63. And the shaded areas represent the density of normals or poles[27] of the crystallographic (111) planes with reference to a random orientation of normals, R, in an ideal metal sample. The variability of the shaded areas indicates a preferred orientation of the (111) planes which is different from that of an extruded alloy with its texture being defined by a (111) direction parallel to the product axis. Thus this may be the reason for the discrepancy in toughness for the two types of fabricated alloys in the two plate directions.

For the majority of the recorded data, the notch-yield ratio or the stress intensity factor is determined for the longitudinal and transverse direction with no distinction being made between the transverse directions. Thus one cannot immediately say that the recorded transverse fracture data is the most critical condition since the transverse direction is not specified. If one were to assume the results of Table 26 to hold for all rolled plates, the recorded transverse toughnesses would be critical if their values were considerably less than the fracture resistance of rolled plate in the longitudinal direction. This assumes that the plate toughness is about the same in the longitudinal and long-transverse directions. Thus one has to examine the tabulated results carefully before being able to say that a particular alloy has a specified minimum of toughness in a definite direction.

Concerning other plate directions, work has been performed to show the variation of specimen toughness with directions other than the three principal directions (57,58). Basically what has been done is the impact testing of notch specimens at various angles measured from the rolling direction. This is illustrated in Figure 64 and shows two types of specimens cut from the plate at 15° intervals from the rolling direction, OX. The results of these tests for an aluminum alloy, V95,[28] are given in Table 29 and show that the maximum impact energy occurs when the specimen is oriented in the longitudinal direction. Note that as the specimens are rotated in the plane of rolling and away from the rolling direction, the toughness decreases to a minimum which corresponds to a transverse fracture resistance. All of the oblique specimens have a toughness which is intermediate of the maximum and minimum of the longitudinal and transverse directions respectively. Thus the transverse direction is critical when specifying the fracture resistance of alloy plate.

The major reason for the above anisotropy of alloy plate is the final texture of the product. This texture is caused by slip in specific crystal directions (110) on restricted planes (111). And deformation on the (111) planes causes rotation of grains and grain fragments into preferred orientations relative to the working direction and the surface of the product. Thus what results are long plate-like grains in the direction of rolling which are stacked on top of each other.

Not only does rolling produce reorientation of grains but also extruding as mentioned before. Extruding of liquid metal along a parting plane of a forging die cavity can cause the grain to flow in a direction normal to the forging surface rather than parallel to it. The result is a transverse grain structure along the parting plane which can cause a toughness variation (59). This is schematically shown in Figure 65 for two crack propagation systems, and note that for the low toughness specimen, the fracture surface exhibits a relatively smooth topography; whereas, the high toughness specimen is characterized by a dimpled rupture (see micrographs in Figure 65). Note that the difference in fracture mode is due to the crack propagating through the diverse grain structure. For the low toughness direction, the crack is splitting the platelets apart; whereas, the crack in the tougher alloy is confined to either grain boundaries or else actually tears grains apart in a ductile fashion. Thus this indicates that all forms of mechanical working can result in some sort of anisotropy which must be considered.

One other problem that may arise in the anisotropy of alloy plate is the degree of anisotropy with plate thickness. Results of tensile tests of center slotted panels show that increasing plate thickness also implies a greater degree of anisotropy (60). Table 30 and Figure 66 verify this by showing the variation of G_c[29] with thickness ranging from $\frac{1}{16}$

to 1 in. Note that as the thickness increases the degree of anisotropy changes from 82 pct. in the $\frac{1}{16}$ in. gage to 25 pct. in the 1 in. gage. This is easily identified in Figure 66.

The reason for this characteristic is that the greater degree of working in the thinner gages causes a finer distribution and size of insoluables and an equiaxed grain structure (60). This equiaxed structure is derived from the fact that the grains are also stretched in width as in length, and the degree of width stretching determines the amount of anisotropy (58).

As the thickness increases, the grains lose their equiaxed structure and become elongated in the direction of working. Similar to the elongation of the grains is the elongation of the intermetallic compounds (60). These intermetallics act as nuclei for voids and process zone cells ahead of a crack tip. And tests have shown that if the inclusions are spread further apart and/or are fewer in number, the process zone cell will be larger along with the plastic zone ahead of the crack tip (46,50). Usually under cases of tensile instability the process zone size is given by:

$$k_{I_c} = E\,(2\pi r)^{\frac{1}{2}}\,n$$

where E is the modulus of elasticity

n is the strain hardening exponent

k_{I_c} is the plane strain stress intensity factor

and

r is the process zone size as measured from the crack tip.

Reported data shows that for a high purity alloy, 7075-T6, the intermetallics are fewer in number and more widely spread. Consequently, the process zone size has been evaluated as 5.49×10^{-4} in. as compared to 2.85×10^{-4} in. for the commercially available 7075-T6 with a greater number of intermetallics. In lieu with this increase in process zone size has been an increase in toughness(50).

Specifically, the way this relates to the anisotropy of the thicker plate is that the elongated inclusions exhibit a greater spacing for crack directions parallel or cross grain, and thus the process zone size is larger in this direction resulting in greater toughness (46). In the short transverse direction, the crack propagating parallel to the rolling plane sees a greater number of inclusions of smaller size and at a shorter distance. Thus the process zone size is reduced in this direction along with toughness. As the material gets even thicker, the difference in grain and inclusion orientation is extenuated even more, and thus the degree of anisotropy is increased.

In examining the anisotropy of alloy plate, another characteristic was found of thick plate. This other property is that the toughness of specimens taken from the center of the plate differ from the toughness of the same specimens taken from near the surface. But this is only true for the longitudinal direction because the specimens in the transverse directions exhibit a uniform toughness with thickness (see Figures 51 through 53). The reason for the variation of toughness with thickness in the longitudinal direction is believed to be due to the preferred orientation of the (110) planes in the center of the plate (46). For the center of the preforged plate, there is a five times random count, 5R, for the (110) plane as given by a pole figure, and for the hot rolled plate, the center random count is 7 times, 7R.

The above random count at the center of both plates means that the number of slip systems as determined by the density of the (110) planes is large. And thus a crack has a greater chance of being blunted because there is a plastic deformation capability due to absorbtion of resolved shear stresses on the available slip systems (46). Thus in this way texture variation can affect the toughness of the plate and may be a means of controlling the toughness variation with thickness.

In summary, the anisotropy of aluminum alloy plate as determined by the longitudinal and transverse directions must be considered in design. And for critical applications the toughness of the plate must be based upon its minimum fracture resistance which is in the short transverse direction or could possibly be near the surface of the plate.

The reason that the short transverse cracking is critical is that propagation occurs with normal stresses on the grains being maximum and with the splitting of the platelets along their parallel faces rather than cross grain or along longitudinal grain boundaries.

Recent investigations have also found that different mechanical processes affect the degree of anisotropy in different ways. Rolling to plate usually results in longitudinal and long-transverse toughnesses being about the same and higher than the short transverse toughness. Extruding has a different effect in that the long-transverse and short transverse toughness is comparable but considerably less than the longitudinal toughness. This difference in the resulting product fabricated by the two processes is believed to be due to the variability in texture as determined by the preferred orientation of the (111) slip planes.

Another characteristic of alloy plate is that the degree of anistropy increases with plate thickness. This property is derived from the increasing elongation and orientation with respect to the rolling direction of the grains and inclusions as thickness increases. The elongated inclusions are more effective in reducing the plastic deformation capability in the short transverse direction.

Basically these are the results of the investigation in in the anisotropy of alloy plate, and they must be considered before making use of plate material for practical applications.

Fracture Resistance of Welds

With the fabrication of large redundant members and pressure vessels, welding of aluminum alloys has become critical from the view point of fracture resistance. In the application of welds there are basically two types; the gas metal-arc or metal inert gas process (MIG)[30] and the gas tungsten-arc or tungsten inert gas process (TIG).[31] These two processes are readily used for the welding of the pressure vessel alloys of the 1xxx, 3xxx, 5xxx, 6xxx series and one alloy of the 2xxx series, 2219. Usually these processes are not used for the high strength aircraft alloys of the 2xxx and 7xxx series because undue cracking results. The alloys used for aerospace applications require special techniques which make them more expensive and inconvenient.

Both types of welded alloys have substantial toughness especially those of the 5xxx series. But work has been done in determining the relative rating of these welds and also to possibly quantize their toughness for use in design.

The first type of welded alloys investigated, namely, those of pressure vessel application have been shown to be rather tough. In fact, both tear and notch tensile tests show that the toughness as given by the UPE and the notch-yield ratio is greater for the weld metal than the parent plate in various cold worked tempers (4,8). This is shown in the bar chart of Figures 67 and 68 and Table 31. Note that the toughness of the filler alloy (see Table 35 for filler designation and composition) is generally greater than the toughness of the parent plate of the same composition with various tempers (Figures 67 and 68) and that even the toughness of welded plate or sheet material with a different filler metal is tougher than that of the base metal (see Table 31). The reason for this fracture property is that welding anneals the material surrounding the weld by the heat and energy required to fuse the plate and filler metal. Thus the toughness of the weld will be indicative of an annealed state, and as pointed out before in the section on processing effects, annealing causes a recrystallization which leads to the maximum possible toughness available in an alloy. Note that even though welding does anneal the weld area, the toughness of the weld is not comparable to that of the annealed parent plate (4,8,61). Figures 67 and 68 and Table 31 again validate this point but also show that the toughness of the weld is generally between that of the annealed parent plate and the base metal in the various tempers. The values of toughness for the various welds in the above table and figures is determined at the center of the weld which is critical since the fracture resistance appears to be the lowest there (see Table 32).

One exception to the fracture strength of an alloy weld of this type relative to the parent plate is that of the welded alloy 6061-T6 with 4043 filler metal. Welds used in this combination of plate and filler material usually develop fracture resistances which are between $\frac{1}{2}$ and $\frac{2}{3}$ of the parent plate in the T-6 temper (4,62 through 64).

Figure 69 shows that the plain strain stress intensity factor for the above weld is about 16 ksi $\sqrt{\text{in.}}$. These values were taken from a surface flawed specimen which might give higher than normal values of toughness but can be used to show the relative rating of the weld and plate. With the same type of flawed specimen, the relative results of plate and weld toughness for the above combination are given in Figure 70 and show on the basis of fracture strength that the toughness of the 4043 weld is considerably lower than that of the 6061-T6 plate.

One important characteristic of all these pressure vessel welds is that their toughness does not decrease with decreasing temperatures (4,65). This is tabulated in Tables 32 through 34 and shows that at a minimum the UPE (Table 32) and the notch-yield ratio (Tables 33 and 34) remain nearly constant down to -320°F., and below -320°F. the measures of toughness decrease slightly (Table 34). Overall, the welded alloys are very tough at lower temperatures as indicated by the notch-yield ratio which is above 1 throughout the range and the UPE which is generally larger than 700 in -lb./in.^{2} (4,65) except for the 2319 and 4043 filler alloys.

Work has been done to quantize the toughness of these welds for use in design. And with the aid of Figure 71, filler alloys with a UPE of 700 should have a plane strain stress intensity factor of 40 ksi $\sqrt{\text{in.}}$ which is beyond the realm of practical lab testing (4). But for the 2319 and 4043 alloys, the lower UPE anticipates a k_I value of possibly 30 ksi $\sqrt{\text{in.}}$ And both values of these stress intensity factors indicate a very substantial toughness which is so large that unstable crack propagation in these welds is not critical.

On the basis of toughness alone, the above pressure vessel type alloys are superior to that of the parent plate. But if strength is considered in lieu with toughness, one finds that for a given level of sttength, the toughness of the welded alloy is less than that of the plate in the wrought condition (65). This is shown in Figure 71 and note that very few of the welded alloys approach the band of the toughness-strength relationships for the wrought alloys. Thus one has to decide on an optimum level of toughness and strength when welded alloys are to be employed, and the toughness of welds must be considered in design just as in plate in order to produce a safe and durable product.

In summary, the welds of the pressure vessel alloys appear to have a characteristic toughness which is greater over a range of temperatures than that of the parent plate available in the various cold worked tempers. This means that unstable crack propagation in these welds generally need not be considered in fracture design unless the 2319 and 4043 filler alloys are used in the as welded or heat treatable condition respectively. But even this fracture problem with these latter filler alloys can be alleviated either by heat treatment as in the case of 2319 or the use of another filler as in the case of 4043.

Even though the toughness of the above welds may be greater than that of the parent plate, their strength levels may be less and thus may be critical in design based upon optimizing both toughness and strength. Usually the welded alloys have a smaller toughness than the wrought alloys for a given strength. Thus, again the problem of utilizing best either toughness or strength of the welds becomes an issue of design just as for plate or other alloy configurations as pointed out previously. This is a dilemma one has to face in designing for both fracture and strength and may not be solved until future research uncovers a new material which does not sacrifice strength for toughness.

As mentioned previously, the high strength alloys of the 2xxx and 7xxx series have welds which require more control over the welding procedure due to cracking. Even though the welds of these series have this undesirable characteristic, their toughness is greater than that of the base metal (4,66). Figure 72 and Table 36 validate this by showing the notch yield ratio as being greater for the filler alloy than the parent plate in a variety of heat treatments (4,66). The reason for this is again due to the annealing effect of the weld.

Note that for the three filler metals employed in the aerospace alloys of Figure 72, the welds in the 7xxx series alloys are less tough than those of the 2xxx and 5xxx alloys and that even the 4043 weld of the 6061-T6 alloy is tougher than the welds of the 7xxx series alloys using different filler metals. As indicated in Figure 72, the aircraft alloys of the 5xxx series appear to have the greatest toughness when used with magnesium filler metals; this appears to be the same result as indicated in the figures for the pressure vessel alloys of the same series.

Concerning temperature variation, the NYR of welds in the high strength alloys in all but the 7xxx series generally remain constant with decreasing temperatures down to -320°F. (4,40). Table 37 shows this to be true but also indicates that there appears to be more embrittlement of the 7xxx series alloys when used with 4043 filler metal than any other of the 5xxx series filler metals. The variation of the high strength series welds with temperature is very similar to the temperature variation of toughness of the parent plate even though the filler material of the welds is not the same as that of the welded plate or sheet (see Figure 73). Thus the welds of the 7xxx series alloys are expected to decrease with decreasing temperature just as for the plate and sheet material of the same series (see Figure 36). And the welds of the 5xxx and 2xxx series will exhibit an invariant toughness with decreasing temperature down to -320°F., and thereafter toughness decreases slightly (see Figures 34 and 35).

No doubt the welded alloys used in aerospace applications will have the inherent inverse relationship between toughness and strength as shown in Figure 71 for the pressure vessel alloys, and they will probably have a weld toughness which is less than the parent plate for a given strength. Thus even for the high strength alloy welds, the designer is faced with a choice between high toughness and medium strength or medium toughness and high strength. Thus again the trade off must be made.

One possible direction in which optimization of weld toughness and strength can be achieved is in the use of filler material. For example, 2319 filler material when used with the 2219 alloy (4,67) can be made to be tougher if aging or solution quenching and artificial aging occur after welding (see Figure 67). Note that the latter form of treatment improves toughness even more than just aging after welding (see Figure 67 and Table 38) and that for the

2219 alloy the strength of the weld is increased also. This can be seen in Figure 71 which shows the heat treated and artificially aged weld as having a strength and toughness which is greater than that of the 2219 welded alloy in just the aged condition.

Other ways in which toughness can be improved is by means of changing the filler alloys. Usually the filler metals containing silicon such as 4043 and 718 (see Table 35) will have a lower toughness than those of the 5xxx series which have magnesium as their predominant alloying element (54,62,63). Tables 39 and 40 show that any of the 5xxx series filler alloys will improve toughness over the silicon bearing fillers for a 6061-T6 alloy, and that even for the high strength alloys of the 7xxx series, the magnesium fillers appear to enhance toughness (see Figure 72). One drawback to both the silicon and magnesium bearing alloys is that they will not respond to heat treatment like the 2319 filler alloy. That is, their toughness will remain the same, at most, after treatment (54,65). Usually there is a decrease in toughness with treatment for these filler alloys (see Table 40 and Figures 67 and 72).

Concerning the use of the 5xxx filler alloys, there is no real advantage of one filler type over another on the basis of toughness (65). The only discernable difference that may appear is that there is a slight decrease in toughness with increasing magnesium content at room temperature as shown in Figure 67. Otherwise, one may find an overlapping of toughness as measured by the NYR over a range of temperatures for alloys using 5xxx series fillers (see Figure 73).

Thus the possibility of improving welded joint toughness and strength may lie within the confines of the choice of filler material combined with the parent plate.

As mentioned before, the critical section of the weld for minimum toughness is at the center (4). But there are basically four regions in the weld area which have a variable toughness. These four sections are the weld metal itself, the transtion zone where filler metal combines with the alloy plate, the heat of fusion or heat affected zone, and the plate itself. In the order of increasing toughness, the weld metal comes first, secondly, the transition or heat of fusion zone, and finally the plate alone (4,62,64,68). For the relative toughness of the center of the various welds with respect to the heat affected zone (HAZ), Table 31 shows a higher UPE for the HAZ than the weld center, and Figure 74 indicates the above order of toughness relative to the weld centerline for a 6061-T6 plate. Note that for the two types of welds in Figure 74, the plane strain stress intensity factor is smaller at the center of the weld and remains essentially constant at various distances within the HAZ, as expected. Note also that the toughness of the transition zone adjacent to the filler-plate boundary is either equal to or slightly greater than the toughness at the plate centerline and that the toughness of any of these locations within the respective zones in Figure 74 is less than that of the 6061-T6 plate which is measured as 24 ksi $\sqrt{\text{in.}}$ using a surface flaw specimen. Thus this demonstrates that in measuring toughness of welded joints, the notch should be located within the filler metal itself and that the toughness of the weld may be lower than that of the base metal, but this may not always be the case as seen in Figures 67 and 72.

One of the toughness characteristics of welds recently studied is the dependence of toughness upon welding procedure. Recent investigations show that welding procedure and/or position has no effect upon fracture resistance of welds in 5083-0 plates and extrusions which were applied in the flat, vertical, and horizontal positions by either an automatic or semi-automatic process. Note that the 5183 welds in the 5083-0 plate have a toughness which is basically invariant with both the welding position or method at all temperatures. Thus the only way in which the welding process appears to affect toughness is in the overall quality or soundness of the weld. Variability in welding is of no consequence (68).[32]

A new type of welding process has been developed which appears to have an advantage over the conventional MIG or TIG welds. This new welding method is called election beam welding (69) and employs a low travel speed, low voltage, high amperage, and over powered type settings to achieve desirable welds (69).

An investigation has been made into the properties of welds using this new process. Specifically, notch tensile specimens of 2219 sheet have been used to derive a plane strain stress intensity factor for both the conventional tungsten-inert gas weld and the election beam weld. Table 42 gives the results of these tests and shows that the TIG weld produces a toughness at the weld centerline which is less than the base metal. For this type of weld, 2319 is used as a filler, and the results of Table 42 agree with those of Figure 67 for the same filler and plate combination except in the figure the UPE is the measure of toughness.

For the plate welded by the election beam process, the as welded toughness is larger than the conventional TIG weld. But note that a different temper is used for the plate, T6 E46. T6 E46 is a modified T6 temper derived by under aging and results in higher tensile properties than the T-6 temper due to the reduced aging time and temperature (69). If this temper is employed after election beam welding of 2219-T42 sheet, the toughness and tensile properties are increased even more (see Table 42). But quenching and distortion problems may occur in this post weld heat treatment process. If these problems do not exist or can be alleviated, employing the new welding process can result in tensile strengths equivalent to base plate and a toughness greater than that of the conventional TIG method. If the post aging process cannot be used, the toughness will still be larger than that of the conventional weld, and there will be a 20 pct. increase in tensile properties over the TIG weld. Thus this newly developed weld process has a definite advantage over the various gas arc welds.

Of definite importance concerning welding procedures, is the effect on toughness of repair welds. Studies of surface flaw specimens indicate that repair welding can reduce toughness considerably, by as much as 50 pct. according to Figure 75. This may be of serious concern since the objective of a repair weld is to alleviate the possibility of failure not to increase it. But according to the study, this latter effect is the result of repair welding. Thus there are three possible plans of action; make sure the initial weld is sound and adequate, have better control over the repair welding process, or finally test more repair weld specimens using accepted testing procedures rather than surface flaw specimens which can possibly predict erroneous results. Only after more tests are completed, can one say that repair welding reduces toughness. But for now, more control should be used in the application of repair welding as possible failure remedies.

One area of weld behavior that has been overlooked so far is that of welded alloy castings. Reported results of notched tensile tests show that, as in the welds of the wrought alloys, the welded castings exhibit a large toughness which is invariant with temperature (65). This is shown in Figure 73 and Table 43 and note that cast alloys can be welded to wrought alloys. The welds of the two types of alloys will develop adequate notch toughness only if the properties of the base metals are about the same. Otherwise, the welded joint will assume the properties of the weaker base metal (65). Thus when welding the two types of alloys, both base metals should have the same tensile properties and toughness.

In comparison to the welded wrought alloys, the welds of cast alloys have a comparable strength but lower toughness. This is evident in Figure 71 and shows the inherent inverse relationship of toughness and strength for the welds in the wrought and cast alloys. Note the lower toughness for the cast alloy welds as compared to the welded and unwelded wrought alloys for a given strength. This relative position of the welded cast and wrought alloys is the same as that of the unwelded alloys. Thus welds in cast alloys do not appear to be optimum on the basis of strength and toughness.

With this disclosure of the toughness of welds in pressure vessel and aircraft alloys, several points must be considered before these alloys are to be used in fabricated welds:

1. On the basis of toughness, weld zone toughness is greater than that of the base metal in either the cold worked or heat treated tempers but is less than the toughness of the annealed base plate. An exception to this is the 6061-T6 plate and 4043 filler alloy and the 2219 plate with 2319 filler alloy as welded.
2. The plane strain stress intensity factor for the pressure vessel alloys is estimated to be 40 ksi $\sqrt{in.}$ with the exception of as welded 2219 and 6061-T6 plate using 2319 and 4043 filler alloys respectively. The k_I value is anticipated at 30 ksi $\sqrt{in.}$ for these latter two welds.
3. Concerning decreasing temperatures, the toughness of the high strength welds of the 7xxx series alloys drops with decreasing temperatures but the toughness of the other types of alloy welds remains essentially the same down to -320°F. thereafter decreasing slightly.

4. On the basis of both strength and toughness, the toughness of welds in wrought and cast alloys is less than that of the wrought base metal for a given strength.
5. Usually filler alloys containing silicon (4043) and magnesium (5xxx series) do not have increased toughness as heat treatment and/or aging is employed.
6. Magnesium base filler alloys provide greater toughness than the silicon bearing filler alloys. And increasing the magnesium content of the filler usually results in a slight reduction in toughness although the toughness of the 5xxx filler alloys is about the same with no big spread of values.
7. Heat treating and/or aging of 2319 filler alloy results in an appreciable increase in fracture resistance.
8. Concerning the toughness with respect to the weld geometry, the toughness increases in the following order according to the distance or location of the zone from the weld centerline; the weld metal proper, the transition zone, the heat affected zone, and the base metal itself.
9. Welding position or procedure (automatic or semi-automatic) has no effect on weld toughness.

10. Electron beam welding increases toughness over the conventional weld processes.
11. Just as in wrought alloys, welded alloy castings exhibit high notch yield ratios which decrease only slightly with temperature below -320°F.; otherwise, the ratio is constant with decreasing temperatures.
12. Welds in alloy castings are lower than welds in wrought alloys based upon toughness and exhibit the inherent inverse relationship between toughness and strength which is lower than the same relationship for wrought alloy plate.
13. Casting alloys when welded to wrought alloys should have comparable base metal properties so as to insure adequate weld toughness.

With these basic discoveries outlined, the understanding of weld behavior and properties should be close at hand. And the critical parameters necessary for safe design are available so that the possibility of weld failure is minimized if taken into account during design.

Conclusion

For the optimization of fracture and yield strength design, few alloys like x7005-T6351 of Figure 76 have been developed. But the above discussion should aid in a possible direction for continued study to achieve the desired objective.

In pursuit of the above objective, many parameters must be taken into account for the optimal alloy such as the effect of alloying additions on the type and location of the hardening zones or precepitates. Even mechanical and thermal processes can alter both toughness and strength. And for optimization of strength and fracture resistance, a combination of compositional change and variable processes must be used. This not only applies to the wrought alloys but also for the alloy castings which respond similarly to increasing alloy additions and foundry processes.

One part of the final alloy microstructure which is critical for alloy optimization is the amount of porosity, oxide inclusions, or intermetallic compounds. These inherent defects if present in appreciable numbers can create notch effects in both wrought and cast alloys which can induce a brittle state of low fracture resistance. Even if the composition and foundry process is closely controlled, the quality of an alloy can make a difference in its final properties. Thus precautionary measures must be taken during the initial stages of alloy fabrication in order to achieve a high quality alloy with the required properties.

Basically, there appears to be three influences or variables which the metallurgist can employ to achieve optimum alloys. These are the composition of the alloy, the process employed to achieve desirable properties, and the control of the melting and casting of the alloy. Only if the combination of these three is taken into account, can an optimum be obtained.

With this intention of combining the three parameters, recent work has shown that this synthesis achieves desirable combinations of fracture resistance and strength and should be pursued even further to obtain alloys superior to that of x7005-T6351.

In trying to discover the reasons for the parameter influences upon fracture resistance and strength, a better understanding of natural occurrences within the microstructure has been attained. But many questions have yet to be answered such as the effect of the precipitate free zone on toughness or strength and will not be answered until more testing and research has been undertaken. But what has been presently discovered or theorized such as the role of dislocations and hardening zones or precipitates is a new step in recognizing the action of properties of the microstructure on the strength and fracture resistance of aluminum alloys. The understanding of how the microstructure is formed or changes with composition or foundry process and how it reacts under service load is the key to developing tougher and stronger alloys.

The explanation of the dependence of toughness and strength on the microstructure is not the only development in aluminum alloys within the recent years, for work has been done to quantize this fracture resistance so as to facilitate design based upon fracture rather than yield strength.

Published results as indicated in Figure 27 and Table 3 show a band and listing of plane strain stress intensity factors for a given yield strength of commercially available high strength alloys. The intention is to be able to design for either a critical crack size which must be detected by various inspection techniques or for a nominal fracture stress given a specific crack size. Note that for a yield criterion, a specified yield point is given,and a designer can choose the appropriate alloy within the band of Figure 27 and find the appropriate fracture resistance. This resistance as read off the plot for various alloy tempers and forms can be used to determine what the safe service load can be before fracture instability accurs. And if this nominal fracture load is less than the service load which causes overall yielding, the fracture stress will govern; otherwise, the load causing large plastic flow in the member will be critical because fracture is very unlikely to occur. Thus the quantized fracture resistance of the high strength alloys allows for a two fold design and eliminates the possibility of unstable or catastrophic failure if a careful design is undertaken.

Other factors which are critical in design and have been disclosed in the above discussion are the variability of toughness in the three principal directions of alloy plate and the toughness of welds. Both must be considered if large, thick plate is to be fabricated into spar components of aircraft or any other multiple redundant structure. And usually the miniumu fracture resistance of the plate must be taken into account. But for plate of a given yield strength, either the toughness in the short transverse direction or weld toughness can be critical. Predominantly the latter governs because the toughness of the weld is less than that of the parent plate for a given yield strength. Thus for a safe design, the minimum toughness must be taken into account even though cracks may be found predominantly outside of the weld metal or else in the longitudinal or long transverse direction of a plate where the fracture resistance can be larger.

By showing what is critical in designing for fracture and what data is available concerning the description of an alloy's ability to retard catastrophic fracture, the objective of this discussion is complete. And by demonstrating what properties of the microstructure are critical for adequate toughness, the designer and/or metallurgist has greater insight into the role of each variable and the possible direction in which further research can be pursued. Even though this study was unable to survey all the available material, the highlights of published data on the fracture resistance of aluminum alloys appears to be covered and should be adequate to show the value of aluminum alloys as a structural material based upon both strength and fracture resistance.

Bibliography*

1. Holt, M. and Kaufman, J.G., "Indices of Fracture Characteristics of Aluminum Alloys under Different Types of Loading," Material Technology - An Inter-American Approach, 1968, p. 426-436.
2. Kaufman, J.G. and Hunsicker, H.Y., "Fracture Toughness Testing at Alcoa Research Laboratories," Fracture Toughness Testing and Its Applications, American Society of Testing Materials, 1965, p. 290-309.
3. Kaufman, J.G., Schilling, P.E., and Nelson, F.G., Fracture Toughness of Aluminum Alloys, American Society of Metals Congress 1968, Detroit, Mich.
4. Kaufman, J.G., Nelson, F.G., and Holt, M., "Fracture Characteristics of Welds in Aluminum Alloys," Welding Journal, July 1966, p. 321S-329S.
5. Kaufman, J.G. and Johnson, E.W., "Notch Sensitivity of Aluminum Alloy Sheet and Plate at -320°F. Based upon Notch: Yield Ratio," Advances in Cryogenic Engineering, Plenum Press, N.Y., 1963, vol. 8, p. 678-685.
6. Goode, R.J., "Evaluation of Materials for Hydrospace Applications," Weld Imperfections, 1968, p. 393-426.
7. Van Horn, K.R., Aluminum, American Society for Metals, Metals Park, Ohio, 1967, vol. 2, p. 95-97.

8. Kaufman, J.G., "Fracture Toughness of Aluminum Alloy Plate from Tension Tests of Large Center Slotted Panels," Journal of Materials, Dec. 1967, 2,(4), p. 889-914.

9. American Society of Testing Materials Recommended Practice for Plane Strain Fracture Toughness Testing of High Strength Metallic Materials Using a Fatigue Cracked Bend Specimen, Letter Ballot Draft, 1967.

10. Kaufman, J.G. and Holt, M., "Evaluation of Fracture Characteristics of Aluminum Alloys at Cryogenic Temperatures," Advances in Cryogenic Engineering, Plenum Press, N.Y., 1965, vol. 10, p. 77-85.

11. Davis, R.A. and Quist, W.E., "Fracture Toughness," Materials Design Engineering, vol. 62, no. 5, Nov. 1965, p. 91-97.

12. Kaufman, J.G., "Role of Theoretical Stress Concentration Factor in Evaluating Notch Toughness," Materials Research and Standards, vol. 7, no. 5, May 1967, p. 189-193.

13. Chu, H.P., "The Notch-Bend Strength of Titanium, Aluminum and Copper Base Alloys in Heavy Sections," Journal of Basic Engineering, Dec. 1969, 91, (4), p. 830-840.

14. Kuhn, P., "Strength Calculations for Sheet Metal Parts with Cracks," Materials Research and Standards, Sept. 1968, 8, (9), p. 21-25.

15. Christian, J.L., Witzell, W.E., and Hurlich, A., "Evaluation of the Effects of Specimen Configuration and Testing Variables on Crack Propagation Properties," Advances in Cryogenic Engineering, Plenum Press, N.Y., 1965, vol. 10, p. 86-101.

16. Davis, S.O., Tupper, N.G., and Niemi, R.M., "Effect of Specimen Type and Crack Orientation on Fracture Toughness," Engineering Fracture Mechanics, June 1968, 1, (1), p. 213-233.

17. Kaufman, J.G., Nelson, F.G., and Holt, M., "Fracture Toughness of Aluminum Alloy Plate Determined with Center Notch Tension, Single Edge Notch Tension, and Notch Bend Tests," Engineering Fracture Mechanics, Aug. 1968, 1, (2), p. 259-274.

18. Mulherin, J.H., Armiento, D.F., and Markus, H., "The Relationship between Fracture Toughness and Stress Concentration Factors for Several High Strength Aluminum Alloys," Journal of Basic Engineering, vol. 86, no. 4, Dec. 1964, p. 709-717.

19. Van Horn, K.R., Aluminum, American Society for Metals, Metals Park, Ohio, 1967, vol. 1, p. 114-126.

20. Nock Jr., J.A. and Hunsicker, H.Y., "High Strength Aluminum Alloys," Journal of Metals, 1963, 15, (3), p. 216-224.

21. Knoll, A.H. and Kaufman, J.G., "Lithium in Aluminum 2020," Metal Progress, vol. 77, June 1960, p.80-82.

22. Ghate, V.B. and West, F.D.R., "The Effect of Manganese on the Tensile Properties of Quenched and Aged Aluminum Copper Alloys Containing 3 and 4.5 Wt. % Copper," Metallurgia, vol. 63, June 1961, p. 269-272.

23. Unwin, P.N.T. and Smith, G.C., "The Microstructure and Mechanical Properties of Al - 6% Zn - 3% Mg," Journal of the Institute of Metals, vol. 97 (9), Sept. 1969, p. 299-310.

24. Ryum, N., Haegland, B., and Lindtveit, F., "Brittleness and Microstructure of Some Al - Mg - Zn Alloys," Zeitschrift Metallkunde, vol. 58, no. 1, Jan. 1967, p. 28-31.

25. Christian, J.L. and Watson, J.F., "Mechanical Properties of Several 2000 and 6000 Series Aluminum Alloys," Advances in Cryogenic Engineering, Plenum Press, N.Y., 1965, vol. 10, p. 63-76.

26. Christian, J.L., "Effects of Chemistry and Processing on the Mechanical Properties of Engineering Alloys at Cryogenic Temperatures," Metals Engineering Quarterly, 1964, 4, (3), p. 53-63.

27. Christian, J.L. and Watson, J.F., "Mechanical Properties of Several 5000 Series Aluminum Alloys at Cryogenic Temperatures," Advances in Cryogenic Engineering, Plenum

Press, N.Y., 1962, vol. 7, p. 490-497.

28. Christian, J.L. and Watson, J.F., "Properties of 7000 Series Aluminum Alloys at Cryogenic Temperatures," Advances in Cryogenic Engineering, Plenum Press, N.Y., 1961, vol. 6, p. 605-621.

29. Tetelman, A.S. and McEvily Jr., A.J., Fracture of Structural Materials, John Wiley & Sons,Inc., N.Y., 1967, p. 217, 329-333.

30. Low Jr., J.R., "Effects of Microstructure on Fracture Toughness of High Strength Alloys," Engineering Fracture Mechanics, June 1968, 1, (1), p. 47-53.

31. Tetelman, A.S., "Fundamental Aspects of Fracture with Reference to the Cracking of Steel Weldments," Weld Imperfections, 1968, p. 249-275.

32. Gangulee, A. and Gurland, J., "On the Fracture of Silicon Particles in Aluminum Silicon Alloys," Transactions of the Metallurgical Society of the American Institute of Mining Engineers, vol. 239, no. 2, Feb. 1967, p. 269-272.

33. Brock, D., "Some Considerations on Slow Crack Growth in a Commercial Aluminum - Copper Alloy," International Journal of Fracture Mechanics, March 1968, 4, (1), p. 19-29.

34. "Aluminum Shows Its Mettle in Cryogenic Tests," Modern Metals, vol. 18, no. 9, Oct. 1962, p. 64.

35. Tellom, R., "Aluminum in Cryogenics," Light Metal Age, vol. 18, Feb. 1960, p. 6-17.

36. Kaufman, J.G. and Bogardus, K.O., "Tensile Properties and Notch Toughness of Aluminum Alloys at -452°F. in Liquid Helium," Advances in Cryogenic Engineering, Plenum Press, N.Y., 1968, vol. 13, p. 294-308.

37. Rice, L.P., Campbell, J.E., and Simmons, W.F., "The Tensile Property Evaluation of One 5000 Series Aluminum Alloy at the Temperature of Liquid Helium," Advances in Cryogenic Engineering, Plenum Press, N.Y., 1963, vol. 8, p. 671-677.

38. Rice, L.P., Campbell, J.E., and Simmons, W.F., "Tensile Behavior of Parent Metal and Welded 5000 Series Aluminum Alloy Plate at Room and Cryogenic Temperatures," Advances in Cryogenic Engineering, Plenum Press, N.Y., 1962, vol. 7, p. 478-489.

39. "Can Aluminum Plate Compete for Cryogenic Jobs?" Iron Age, vol. 188, Dec. 14, 1961, p. 116-118.

40. Mayer, L.W., Anderson, W.A., Brandt, J.L., and Kaufman, J.G., "What Four New Aluminum Alloys Have to Offer," Metal Progress, 95, (5), May 1969, p. 68-72.

41. Steigerwald, E.A. and Hanna, G.L., "Influence of Work-Hardening Exponent on the Fracture Toughness of High Strength Materials," Transactions of the Metallurgical Society of the American Institute of Mining Engineers,

Feb. 1968, 242, (2), p. 320-326.

42. Van Horn, K.R., Aluminum, American Society for Metals, Metals Park, Ohio, 1967, vol. 1, p. 148.

43. Ryder, D.A. and Vian, R.E., "Some Observations on 45° Fracture in a Commercial Polycrystalline Aluminum - Zinc - Magnesium - Copper Alloy," Journal of the Institute of Metals, 90, (10), 1961-62, p. 383-386.

44. Spuhler, E.H., Knoll, H., and Stickley, G.W., Survey of Candidate Aluminum Materials for Supersonic Aircraft Applications, S.A.E. paper no. 742F, Society of Automotive Engineers, vol. 485.

45. Zinkham, R.E., "Longitudinal and Short Transverse Fatigue and Fracture Properties of Heavy Aluminum Alloy Plates, Produced by Forging and Rolling," Transactions of the Metallurgical Society of the American Institute of Mining Engineers, July 1964, 245, (7), p. 1519-1525.

46. Zinkham, R.E. and Baker, C., "Variation of Mechanical Properties, Toughness, and Texture in Two Differently Fabricated High Strength Aluminum Alloy Plates (7179-T651)," Engineering Fracture Mechanics, 1969, vol.1, p. 495-506.

47. Kaufman, J.G. and Sicha, W.E., "Notch Toughness of Some Aluminum Alloy Castings at Cryogenic Temperatures," Advances in Cryogenic Engineering, Plenum Press, N.Y., 1966, vol. 12, p. 473-483.

48. Turner, A.N. and Bryant, A.J., "The Effect of Ingot Quality on the Short Transverse Mechanical Properties of High Strength Aluminum Alloy Thick Plate," Journal of the Institute of Metals, 1967, vol. 95, p. 353-359.

49. Ransley, C.E. and Talbot, D.E.J., "The Embrittlement of Aluminum Magnesium Alloys by Sodium," Institute of Metals, Journal, vol. 88, Dec. 1959, p. 150-158.

50. Carman, C.M. and Armiento, D.F., "Plane Strain Fracture Toughness of High Strength Aluminum Alloys," Journal of Basic Engineering, Dec. 1965, p. 904-916.

51. Flemings, M.C., "Premium Quality Aluminum Castings," Foundry, vol. 91, no. 7, July 1963, p. 60-63.

52. Bucknall, E.H., Vohra, Y.P., and Greer, J.B., "Cracking of a Strong Aluminum Alloy Used in Aircraft Construction," Indian Institute of Metals, vol. 19, June 1966, p. 85-90.

53. Schwartzberg, F.R., "A Review of Cryogenic Fracture Toughness Behavior," Advances in Cryogenic Engineering, Plenum Press, N.Y., 1967, p. 458-472.

54. Kaufman, J.G. and Johnson, E.W., "New Data on Aluminum Alloys for Cryogenic Applications," Advances in Cryogenic Engineering, Plenum Press, N.Y., 1961, vol. 6, p. 637-649.

55. Develay, R., Feure, A., Lehongre, S., Mungnier, D., and Schroeter, D., "A Versatile Aluminum Alloy for Cryogenic Applications, Advances in Cryogenic Engineering, Plenum

Press, N.Y., 1967, p. 484-495.

56. Van Horn, K.R., Aluminum, American Society for Metals, Metals Park, Ohio, 1967, vol. 1, p. 83-84.

57. Miklyaev, P.G. and Fridman, Ya.B., "Determination of the Anisotropy of Notch Sensitivity of Metals and the Sensitivity to Initial Crack under Single Impact," Industrial Laboratory, Apr. 1967, 33, (4), p. 565-569.

58. Miklyaev, P.G. and Fridman, Ya.B,"On the Evaluation of Mechanical Properties and the Type of Fracture of Anisotropic Metals," Industrial Laboratory, vol. 32, no. 2, 1966, p. 265-270.

59. Hyatt, M.V., Zahn, H.R., and McMillan, J.C., "Parting Plane Effects of Fracture Mode and Toughness in a 7079-T6 Die Forging," Quarterly Transactions of the American Society of Metals, vol. 59, no. 2, June 1966, p. 342-344.

60. Zinkham, R.E., "Anisotropy and Thickness Effects in Fracture of 7075-T6 and -T651 Aluminum Alloy," Engineering Fracture Mechanics, 1968, vol.1, p. 275-289.

61. Mikesell, R.P. and Reed, R.P., "The Tensile and Impact Strength of Annealed and Welded 5086 Aluminum Down to 20°K.," Advances in Cryogenic Engineering, Plenum Press, N.Y., 1960, vol. 4, p. 101-113.

62. Metzger, G.E., "Some Mechanical Properties of Welds in 6061 Aluminum Alloy Sheet," Welding Journal, Oct. 1967,

p. 4575-4695.

63. Baker, W., Bryant, A.J., and Durham, R.J., "Recent Developments in High Strength Aluminum Alloys," Journal of the Royal Aeronautical Society, 1966 (70)(688), p. 757-763.

64. Lewis, R.E., Hays, K.E., and Watts Jr., G.H., "Imperfections in Magnesium and Aluminum Alloy Sheet Weldments," Weld Imperfections, 1968, p. 481-514.

65. Nelson, F.G., Kaufman, J.G., and Wanderer, E.T., "Tensile Properties and Notch Toughness of Groove Welds in Wrought and Cast Aluminum Alloys," Advances in Cryogenic Engineering, Plenum Press, N.Y., 1969, vol. 14, p. 71-82.

66. De Money, F.W. and Wolfe, G.C., "Tensile and Impact Properties of 7075-T6 and 7079-T6 Plate Hard Forgings and Tensile Properties of Plate MIG Weldments between 75 and -320°F.," Advances in Cryogenic Engineering, Plenum Press, N.Y., 1962, vol. 7, p. 466-477.

67. Kaufman, J.G., Nelson, F.G., and Johnson, E.W., "The Properties of Aluminum Alloy 2219 Sheet, Plate, and Welded Joints at Low Temperatures," Advances in Cryogenic Engineering, Plenum Press, N.Y., 1963, vol. 8, p. 661-670.

68. Nelson, F.G., Kaufman, J.G., and Wanderer, E.T., "Tear Tests of 5083 Plate and of 5183 Welds in 5083 Plate and Extrusions, Advances in Cryogenic Engineering, Plenum

Press, N.Y., 1970, vol. 15, p. 91-101.

69. Trzil, J.P. and Hood, D.W., "Electron Beam Welding 2219 Aluminum Alloy for Pressure Vessel Applications," Welding Journal, Sept. 1969, 48 (9), p. 3955-4085.

*Other additional references appear to be pertinent to the discussion but were not available at M.I.T. or else were written in foreign languages (see Appendix D and E respectively).

APPENDIX A

Footnotes

1. Notch strength analysis:

$$S_N = \frac{\sigma_u}{k_u}$$

$$k_u = 1 + C_m k_\omega \sqrt{a}$$

$$k_\omega = \left(\frac{1 - 2a/w}{1 + 2a/w}\right)^{\frac{1}{2}}$$

$$C_m = \frac{2}{\sqrt{\rho'}} \frac{E_u}{E} \qquad (\mathrm{IN})^{\frac{1}{2}}$$

$$S_N < \sigma_y \qquad \text{(necessary condition)}$$

where S_N = net section stress

σ_u = tensile strength

k_u = stress concentration factor

$2a$ = initial crack length

w = specimen width

ρ' = Neuber's constant

E = Young's modulus

E_u = secant modulus

$0 \leq C_m \leq 30$ 0 (perfectly ductile)

30 (brittle material)

This method gives the same results of the ratio of fracture stress (calculated) to fracture stress (experimental) as the fracture mechanics approach even though smaller specimen widths can be used to maintain the requirement $S_N < \sigma_Y$ thus indicating its usefulness(14).

2. The results of Table 2 show a comparison of the plain strain intensity factor, k_{I_c}, as obtained from different sources and specimens. Thus in the context of the previous discussion this may be debatable even though there appears to be a good agreement with values of k_{I_c}.
3. A single crystalline phase which is solid, homogeneous and contains two or more chemical elements.
4. After quenching of the solid solution to room temperature, the alloy can be aged at room temperature (natural aging) or the alloy can be aged at a moderately high temperature for a period of time and then quenched (artificial aging). Artificial aging accelerates the growth of precipitates.
5. A dislocation is a defect in a crystal which forms a boundary between a distorted and undistorted section of a crystal. There can be basically two kinds of dislocations, an edge or screw. The edge dislocation is formed by a partial plane of mismatched atoms which

has a smaller area than any other plane cut parallel to it through the crystal. A screw dislocation relates to the amount of lattice disturbance around the axis of a spiral structure in a crystal. This distortion is created by connecting former parallel planes together in a helical ramp around the dislocation. The helix is characterized by a pitch of one interplanar distance and an assumed axis which is taken as the dislocation. Note both of the above types of dislocations can be seen separately or in combination. And both types have a stress field associated with it which can be either positive or negative and a high strain energy which means that they will migrate until the total energy of the crystal is reduced.

6. Saturation ratio =

   fraction (Zn + Mg + Cu) total/(Zn + Mg + Cu) soluable at 860°F.

7. Note the ordinate scales for the graphs given in Figure 34 should be reversed.

8. The Burgers vector 6 defines the displacement of a dislocation. If it is parallel to the dislocation line, the dislocation is a screw and if it is perpendicular to the dislocation line, it is an edge dislocation.

9. $\frac{1}{\beta}$ is the elastic stress concentration factor, k.

10. V(c) is directly related to the plastic zone size.

$$V(c) = \frac{4\sigma_y c}{\pi E} \ln\left(\frac{R + c}{c}\right)$$

where σ_y is the yield stress of the material

c is half the crack length

E is Young's modulus

and R is the radius of the plastic zone

11. What is meant here is that the critical crack opening displacement, $V^*(c)$, is not a valid measure of toughness per se but can be used to show that different materials can be compared on a relative basis for their toughness.

12. These tempers reflect how much mechanical work has been used to achieve a required dimension and consequent strength. Usually the degree of hardness or working is measured by the second digit of the temper with 9 indicating a maximum hardness or yield strength and largest reduction of area.

13. Extensive work is usually indicated by an 8 in the second digit of the temper designation.

14. $\ln \sigma_y = n \ln \sigma_0 \varepsilon$

where n is the slope of the line

and $\sigma_y = \sigma_0 \varepsilon^n$

15. The reason that the T6 temper lacks toughness is due to the increased number of coherent phases which cause relative ease of void formation and coalescence (43).

16. The T66 temper indicates a substantial degree of cold working after artificial aging, and the T73 temper means that some cold work is applied after the alloy is overaged.

17. Premium strength casting refers to a controlled foundry practice which provides better mechanical properties than those obtained by conventional practices.

18. A dendrite is a crystal which is shaped in a tree-like branching pattern.

19. A eutectic is two or more intimately mixed solids which have been formed by cooling a liquid solution. It is also an isothermal reversible reaction.

20. One exception to this is the toughness temperature relationship for 195-T6 which practically remains constant with decreasing temperature. The reason for this is speculated to be the nature in which the copper alloy is present.

21. The 5xxx series wrought alloys have a considerable toughness with temperatures down to -320°F., but with temperatures decreasing below -320°F. the toughness drops off somewhat from the room temperature toughness.

22. The hydrogen level of the molten alloy is controlled by injecting gaseous chlorine (Cl_2) into the liquid for a period of time. The longer the fluxing the less hydrogen will be evident in the melt.

23. If the amount of oxide inclusions is considerable, a possible brittle state can exist for the alloy just as for the case of extensive porosity.

24. Texture refers to a definite orientation of the grains rather than a random orientation.

25. The Miller indices within the parentheses refer to the intercepts of a plane with the three principal axes. The indices are determined by the reciprocal of the intercepts multiplied by a common denominator. For example, a plane having intercepts 1, ∞, ∞ will have reciprocals of $\frac{1}{1}$, $\frac{1}{\infty}$, $\frac{1}{\infty}$ and Miller indices of (100).

26. An x-ray pole figure is a projection showing the statistical average distribution of poles of a specific crystalline plane in a metal, with reference to an extended system of axes. For an istropic metal the pole density is uniform and preferred orientation is shown by means of increased density of poles which are normals to the plane being investigated.

27. Poles are the normals to the surface of the planes of interest.

28. The alloy V95 is composed of 2.6 pct. Mg, 1.8 pct. Cu, 6.2 pct. Zn, 0.3 pct. Mn with base Al (Russian alloy).

29.

$$G_c = \frac{k_c^2}{E}$$

where k_c is the stress intensity factor

E is the modulus of elasticity.

30.&31.

Tungsten arc welding (TIG) refers to the use of a tungsten electrode in an inert gas shielded weld. For a shielded arc weld, the arc and the weld metal are protected by a gaseous atmosphere of inert argon. The difference between tungsten and metal arc welding (MIG) is in the type of electrodes where the latter process uses metal electrodes.

32. In reference to the statement that the welding process has no effect on toughness, what is meant is that neither automatic or semi-automatic welding affects toughness.

APPENDIX B

List of Figures

Figure	Data Source	Page of Data Source
23	16	228
24	16	229
25	18	711
26	20	218
27	3	41
28	2	304
29	2	305
30	20	221
31	23	305
32	23	305
33	26	58
34	25	66
35	27	492
36	28	612
37	29	331
38	32	270
39	29	332
40	2	301
41	39	116
42	7	88
43	41	324
44	7	148
45	44	4

Figure	Data Source	Page of Data Source
46	2	300
47	23	302
48	3	44
49	3	45
50	3	46
51	46	500
52	46	501
53	46	502
54	47	480
55	47	479
56	47	480
57	49	158
58	48	356
59	50	913
60	46	499
61	45	1520
62	52	87
63	7	83
64	57	565
65	59	343
66	60	281
67	4	324-S
68	4	328-S
69	64	501

Figure	Data Source	Page of Data Source
70	63	94-S
71	65	78
72	4	328-S
73	40	70
	65	76
74	64	502
75	64	503

Fig. 2. Notch-strength ratios for aluminum alloy sheet with different designs of notches.

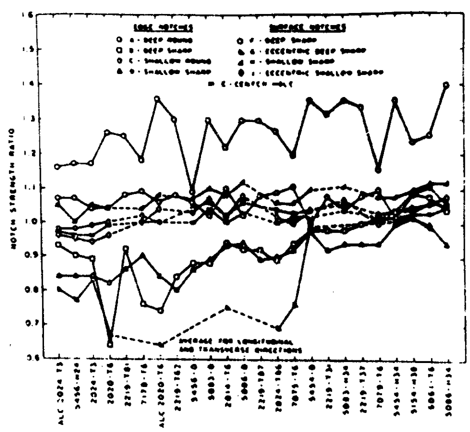


FIGURE 1

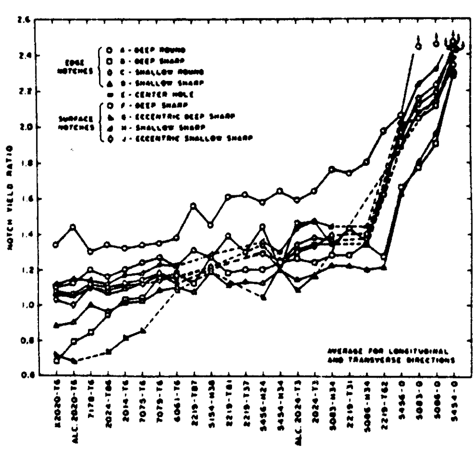


Fig. 3. Notch-yield ratios for aluminum alloy sheet with different designs of notches.

FIGURE 2

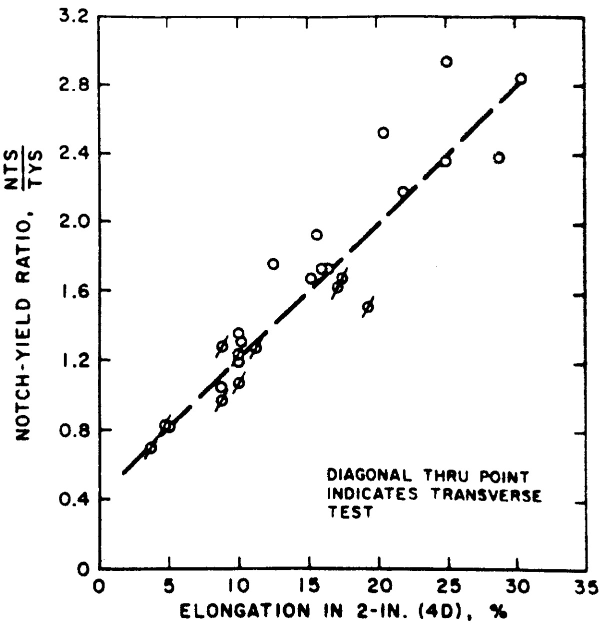
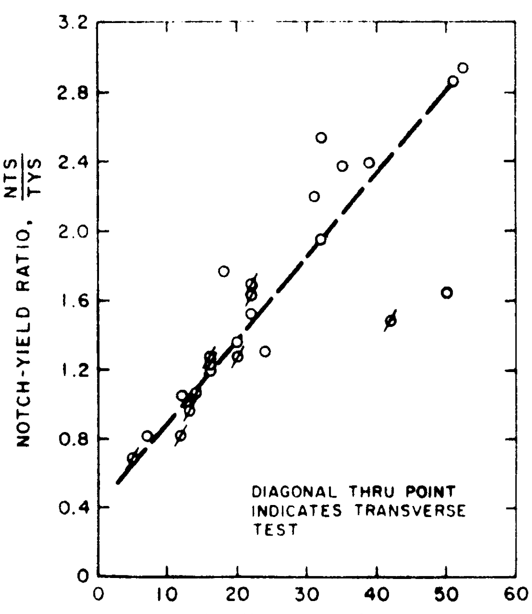


Fig. 1a. Notch - yield ratio vs elongation aluminum alloy plate

FIGURE 3



ig. 1b. Notch - yield ratio vs reduction of area aluminum alloy plate

FIGURE 4

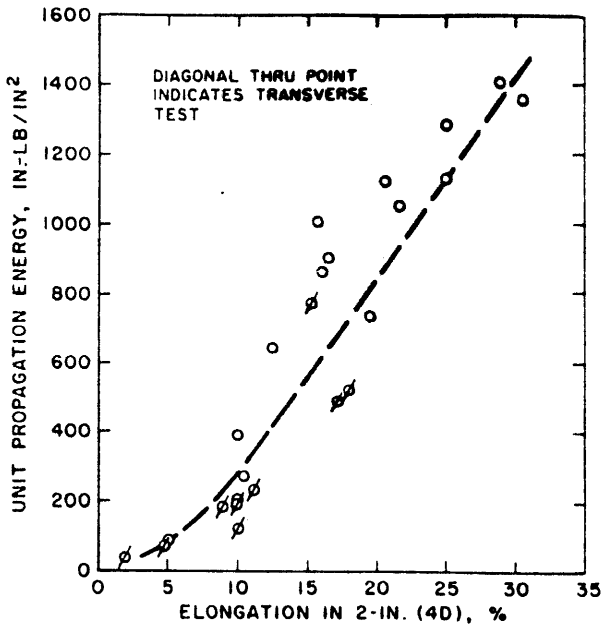


Fig. 4a. Unit propagation energy vs elongation

FIGURE 5

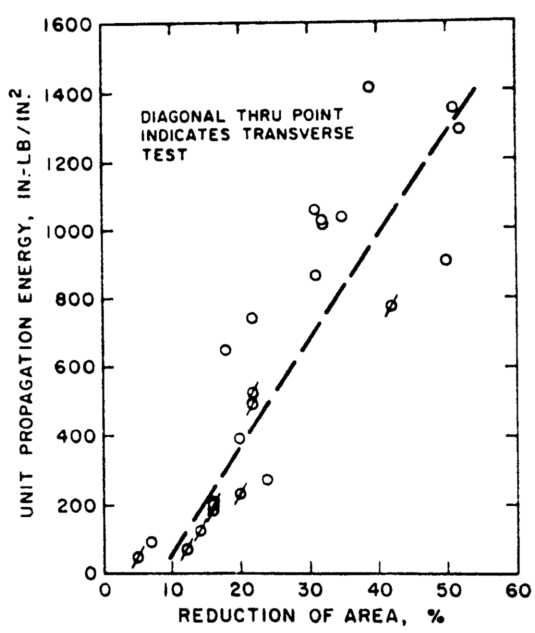


Fig. 4b. Unit propagation energy vs reduction of area

FIGURE 6

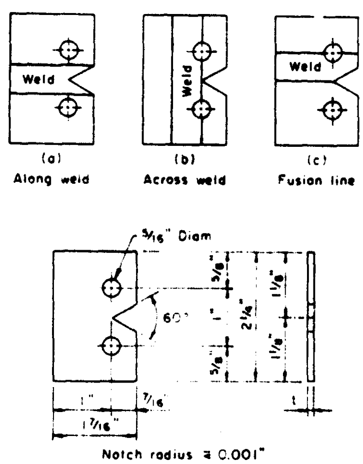


Fig. 1—Tear specimen

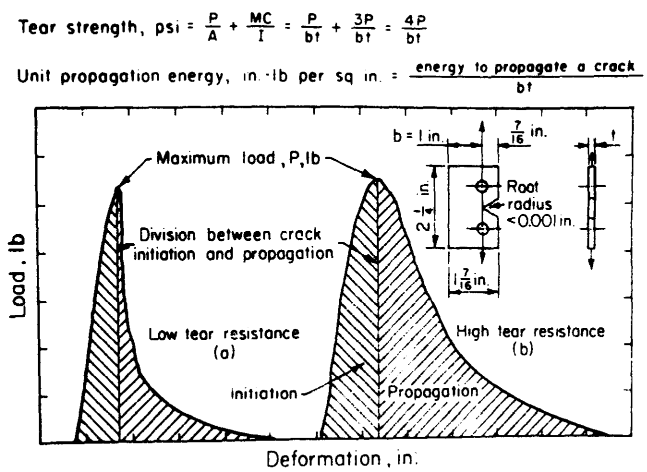


Fig. 2—Representations of tear-test load deformation curves

FIGURE 7

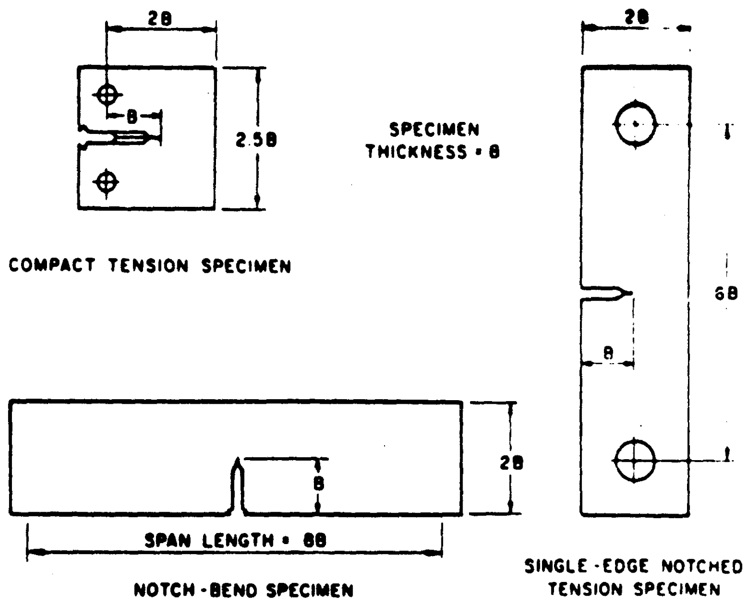


Fig. 1. Fracture toughness specimens for measurement of plane-strain stress-intensity factor, K_{Ic}.

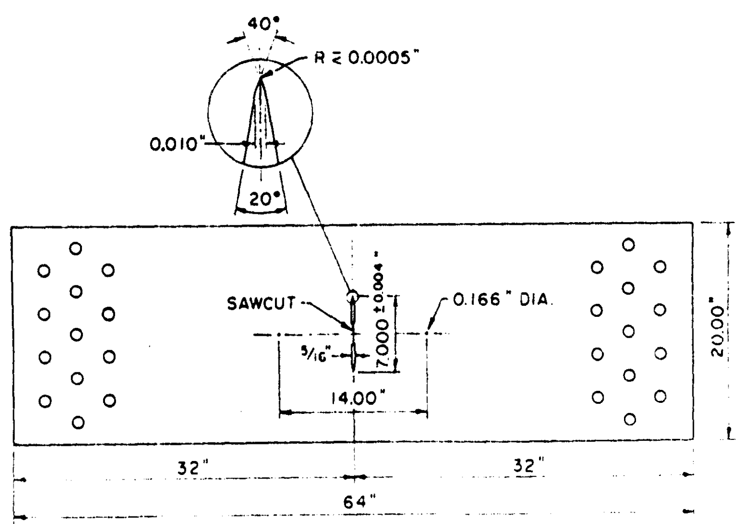


Fig. 1a. Center-notched fracture-toughness specimen (1·0 in. thick).

FIGURE 8

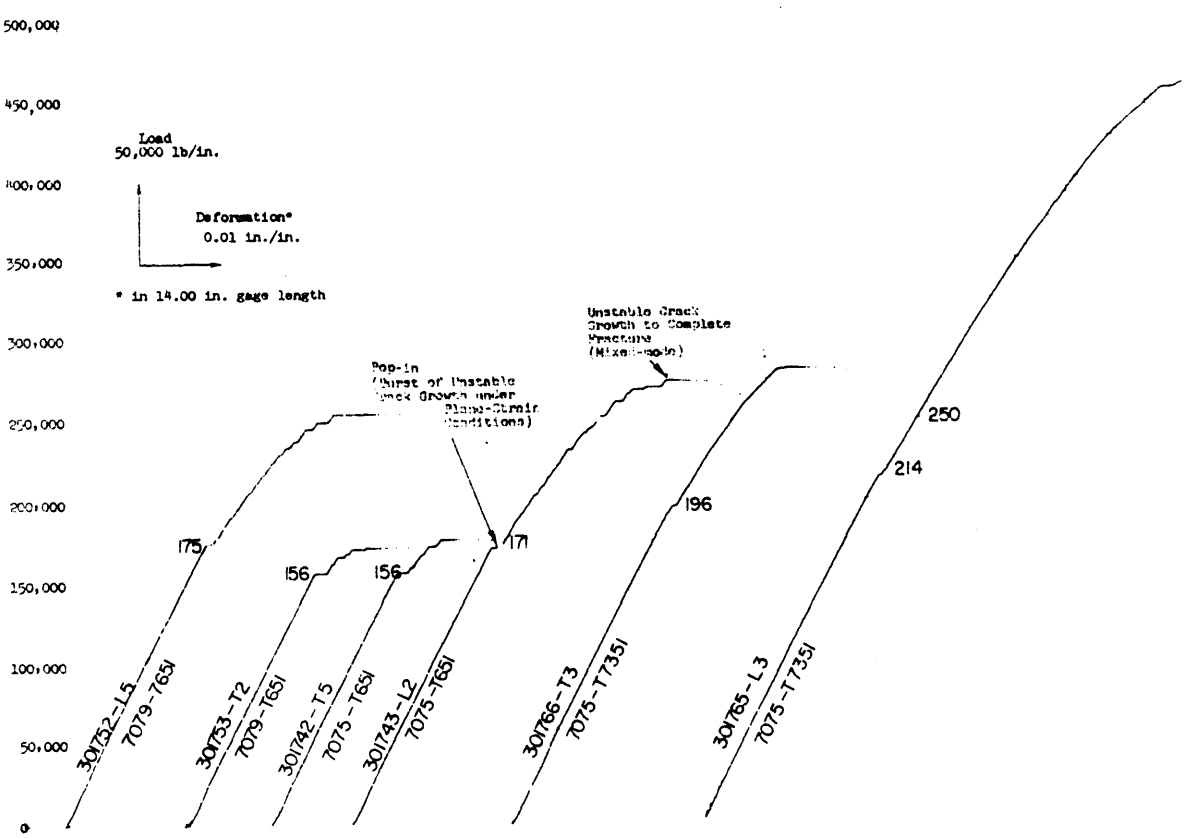
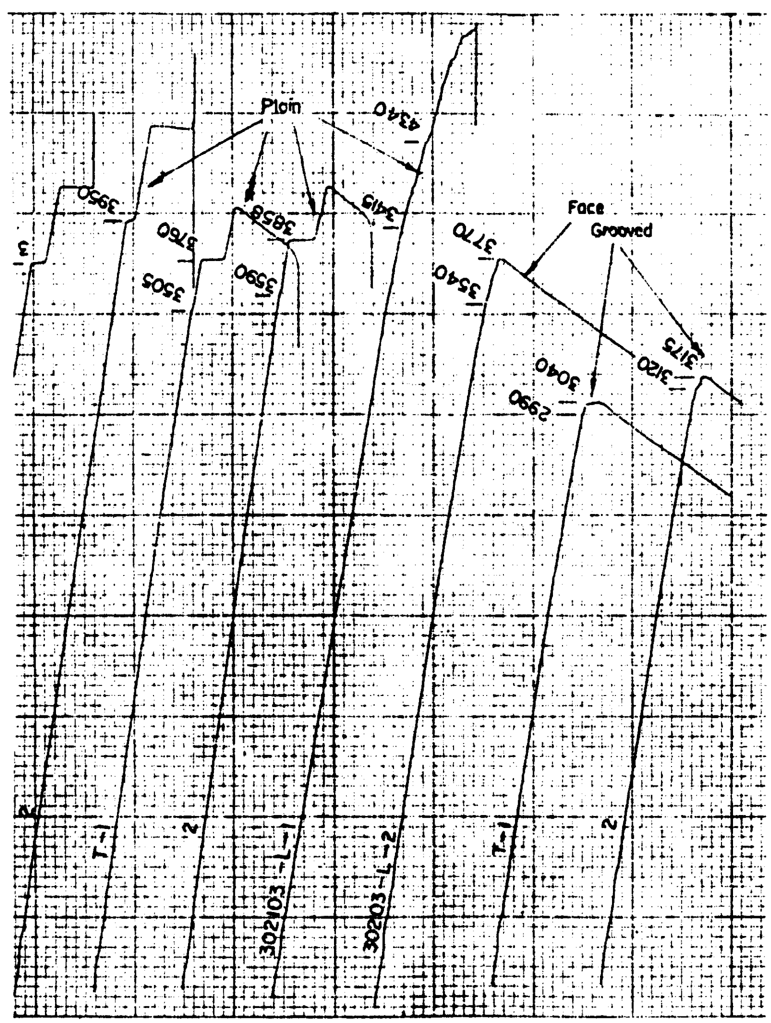


Fig. 7a. Representative load-deformation curves from fracture-toughness tests of center-notched tension specimens.

FIGURE 9



. Representative load-deformation curves from single-edge-notch tension fracture-toughness tests.

FIGURE 10

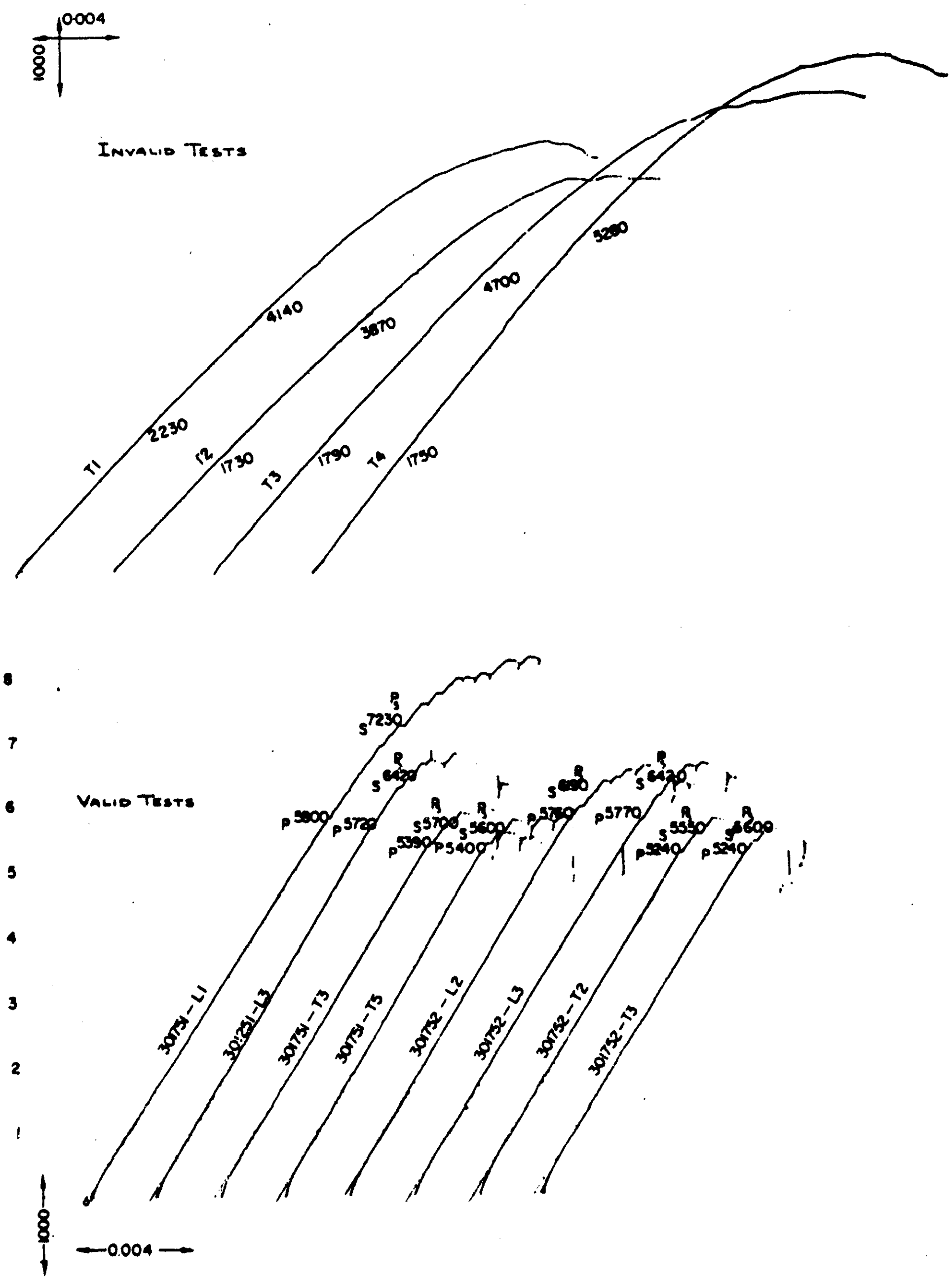
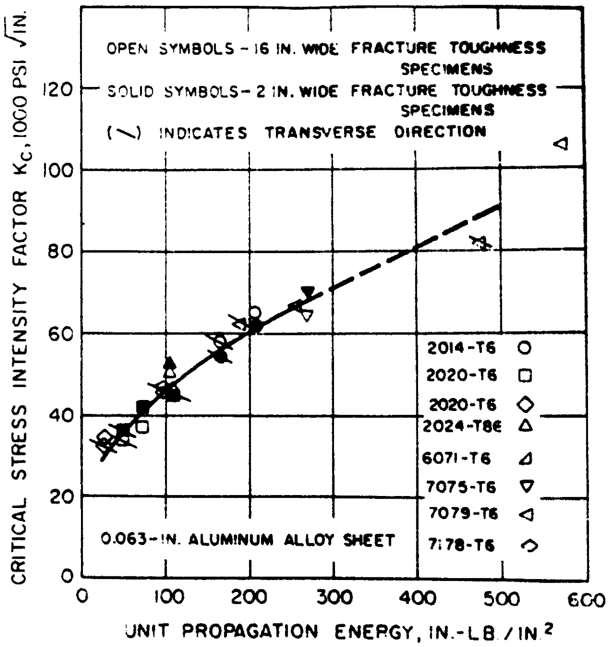


Fig. 7c. Representative load-deformation curves for notch-bend fracture-toughness tests.

FIGURE 11



·Critical Stress-Intensity Factor, K_c, Versus Unit Propagation Energy.

FIGURE 12

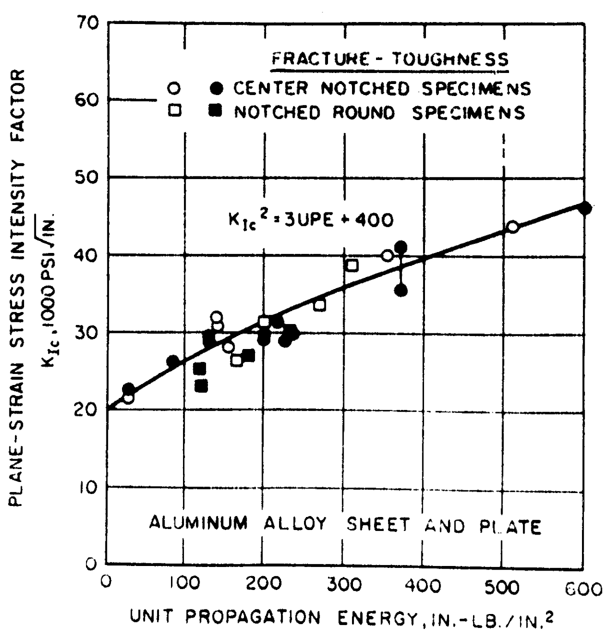


FIG. 12—Relation Between K_{Ic} and Unit Propagation Energy.

FIGURE 13

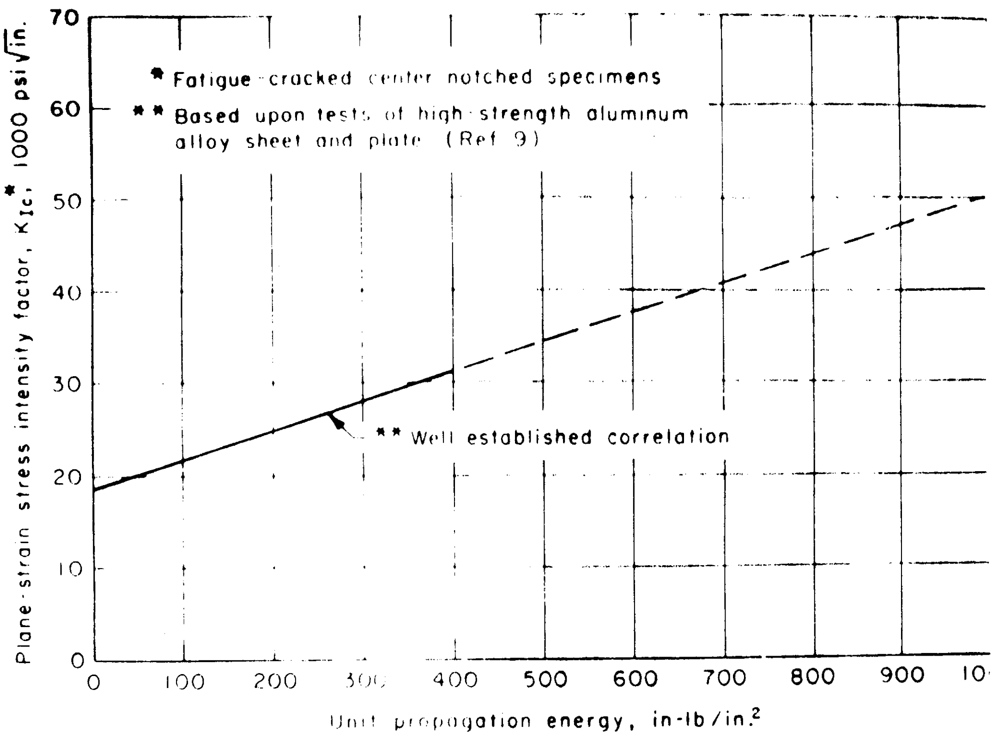


Fig. 3- Relationship between plane strain fracture toughness and unit propagation energy from tear tests

FIGURE 14

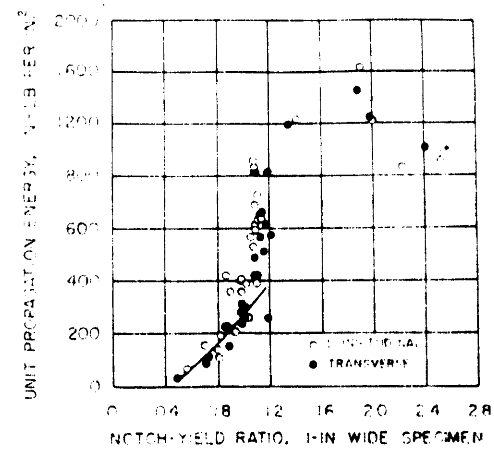


FIGURE 15

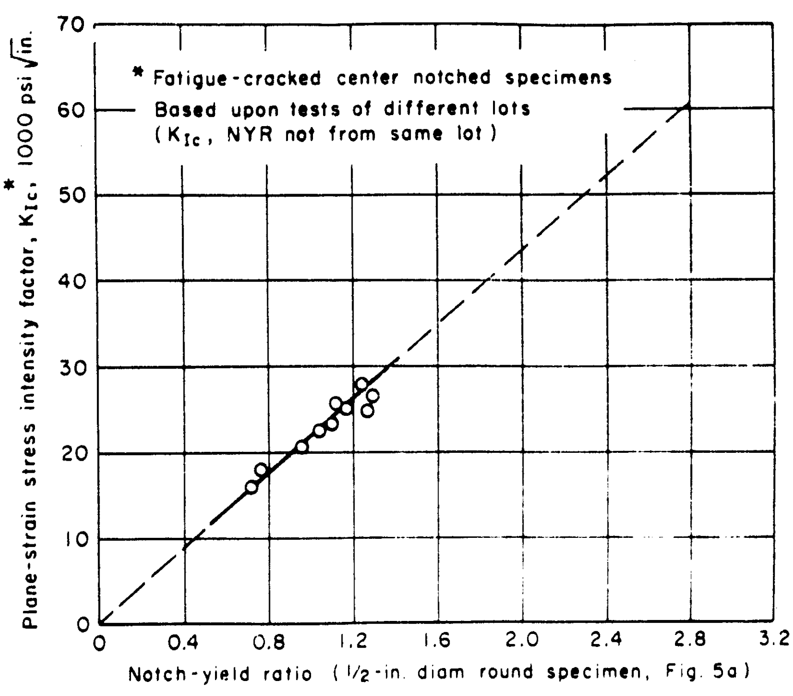


FIGURE 16

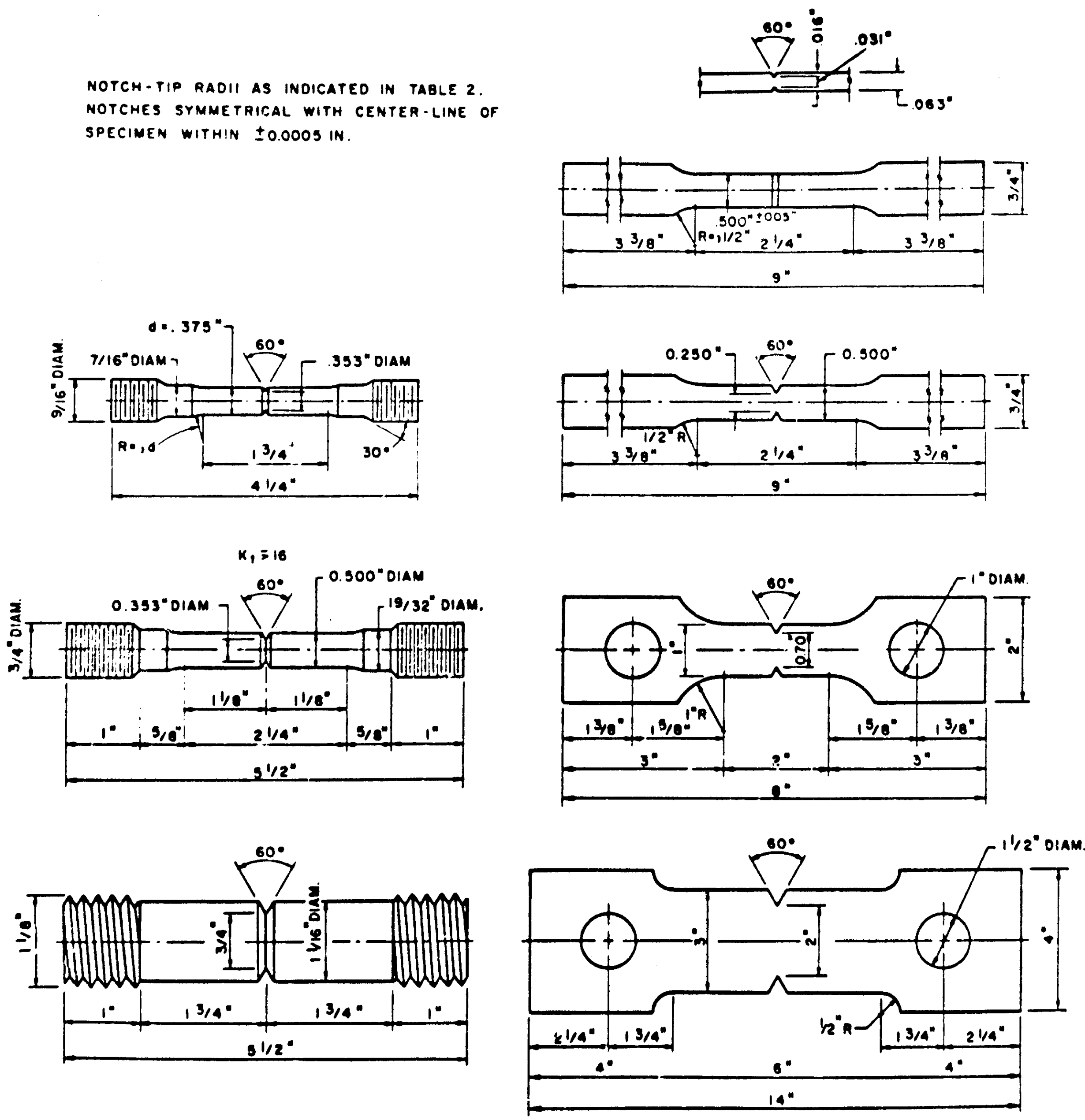


Fig. 1—Designs of notched tensile specimens.

FIGURE 17

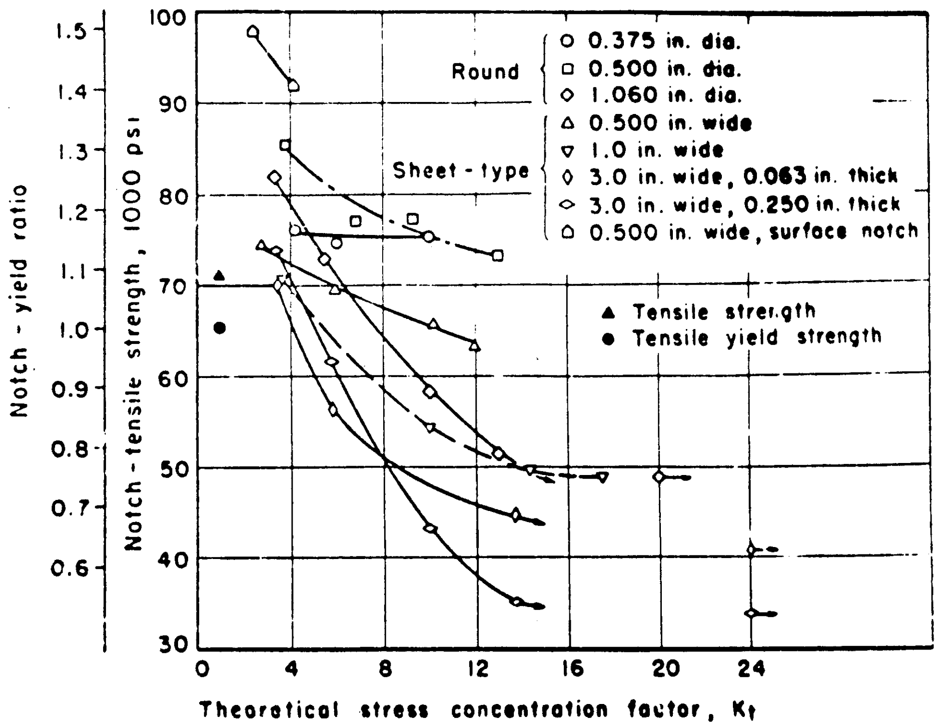


Fig. 2—Average notch-tensile strengths of 2024-T851 plate, transverse.

FIGURE 18

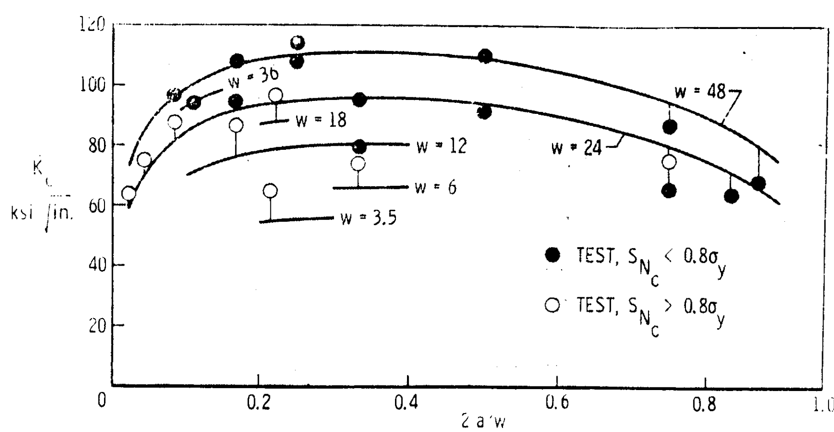


Fig. 2—Experimental and calculated K_c for 2219-T87 sheet (t = 0.10 in.). Curves calculated with $C_m = 0.64$; $\sigma_u = 69.4$ ksi; $\sigma_y = 58.5$ ksi.

FIGURE 19

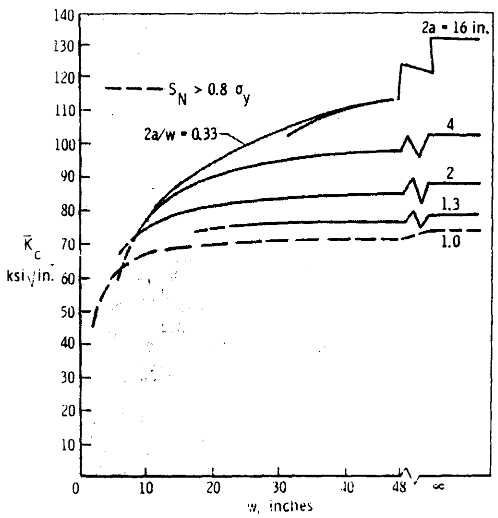


Fig. 3—Calculated K_c for 2219-T87 sheet.

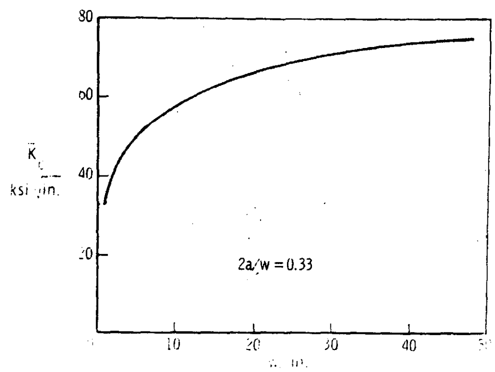


Fig. 4—Calculated K_c for 7075-T6 sheet. ($S_N < 0.8\sigma_y$ for entire curve).

FIGURE 20

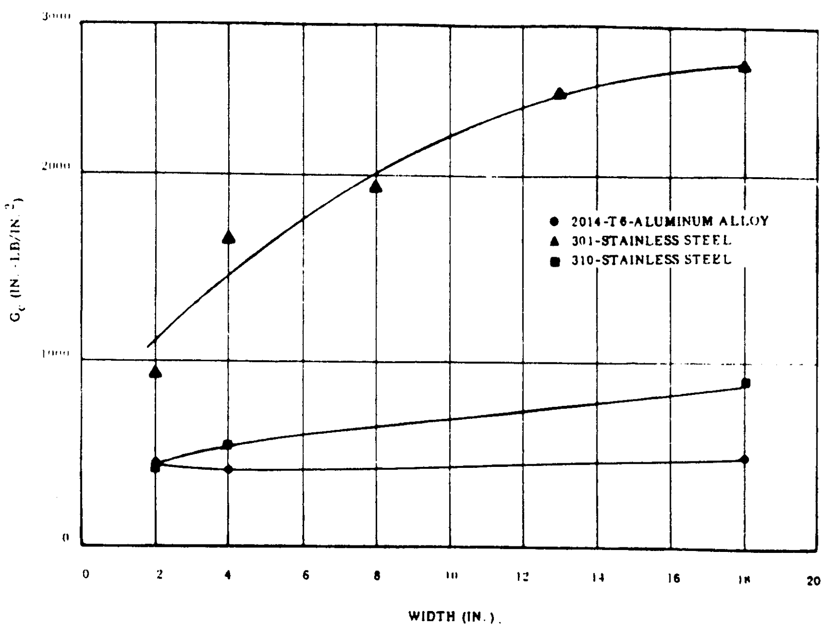
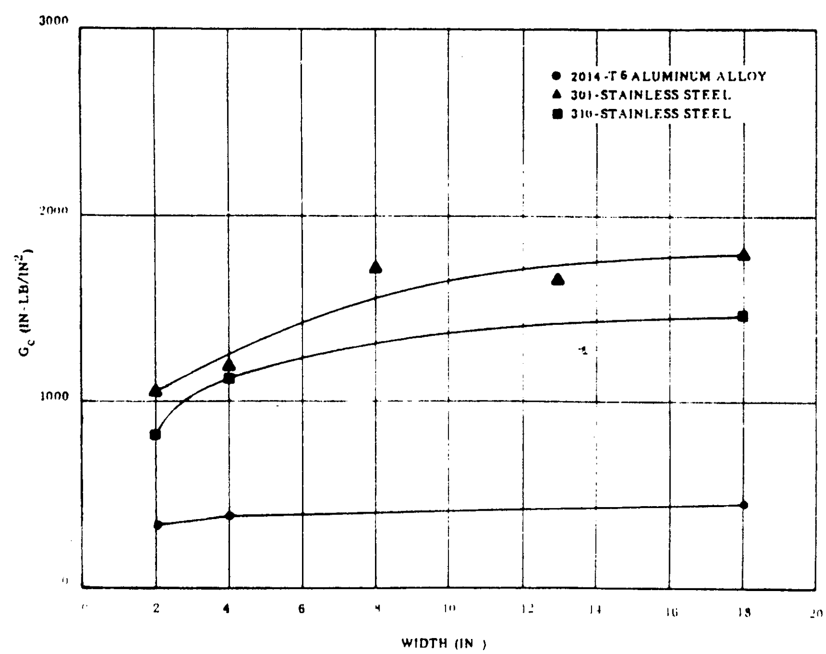


Fig. 6. Variation of crack-extension force (G_C) with width (longitudinal at 78°F).



;. 7. Variation of crack-extension force (G_C) with width (longitudinal at −320°F).

FIGURE 21

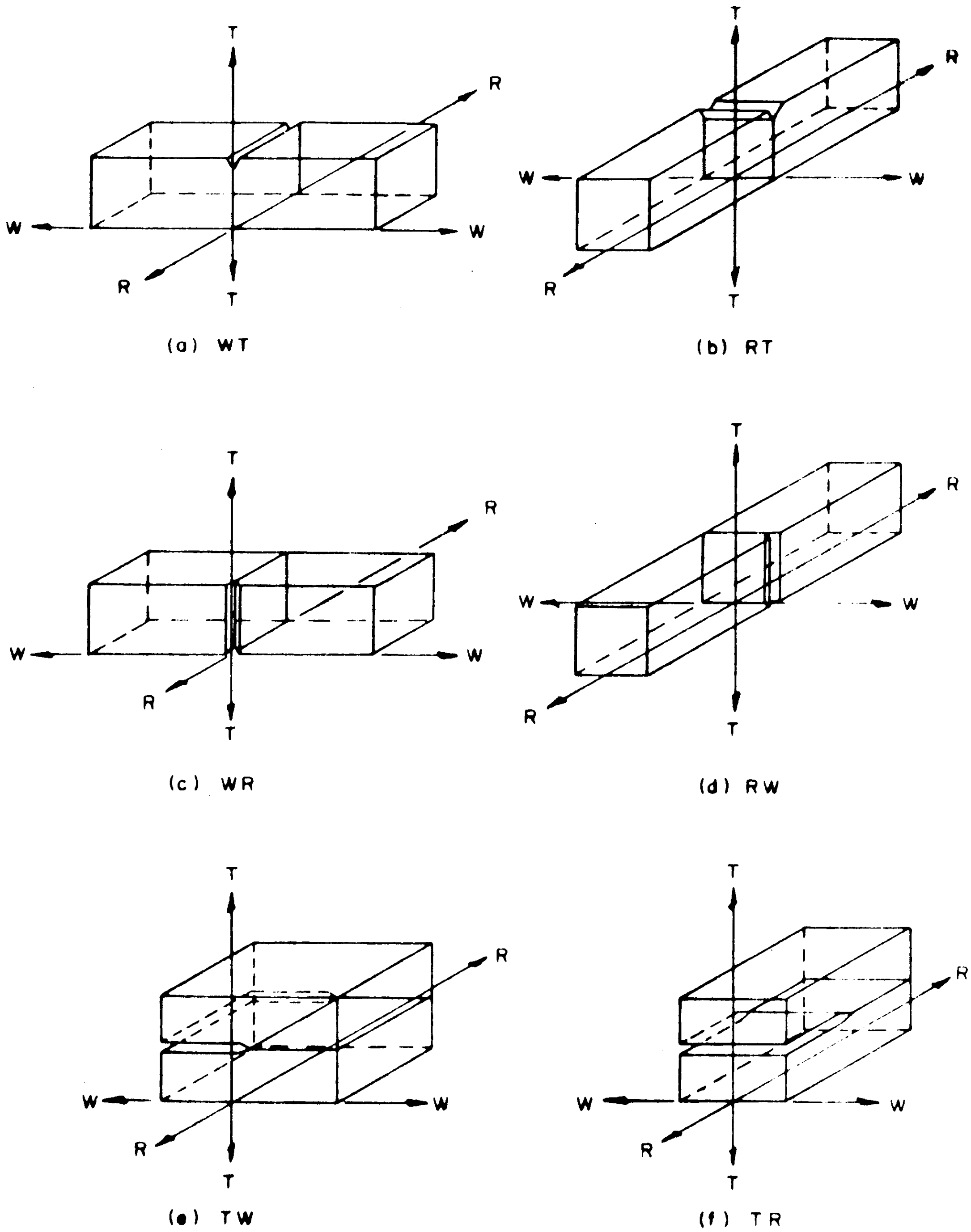


Fig. 17. Crack propagation systems studied.

FIGURE 22

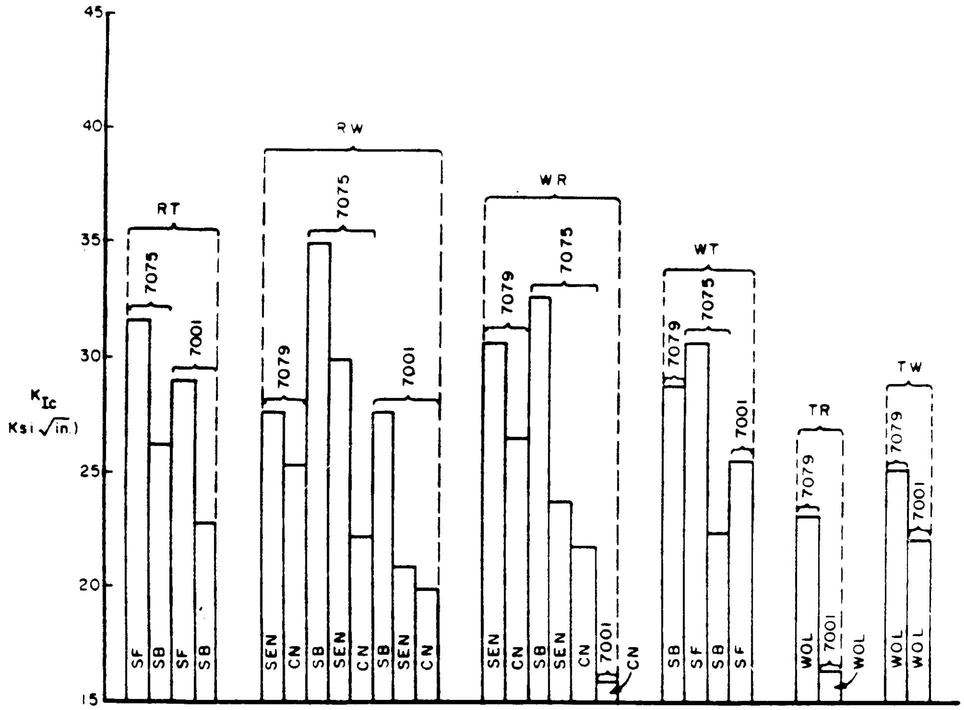


Fig. 18. Summary of effects of specimen type and crack propagation system on K'

FIGURE 23

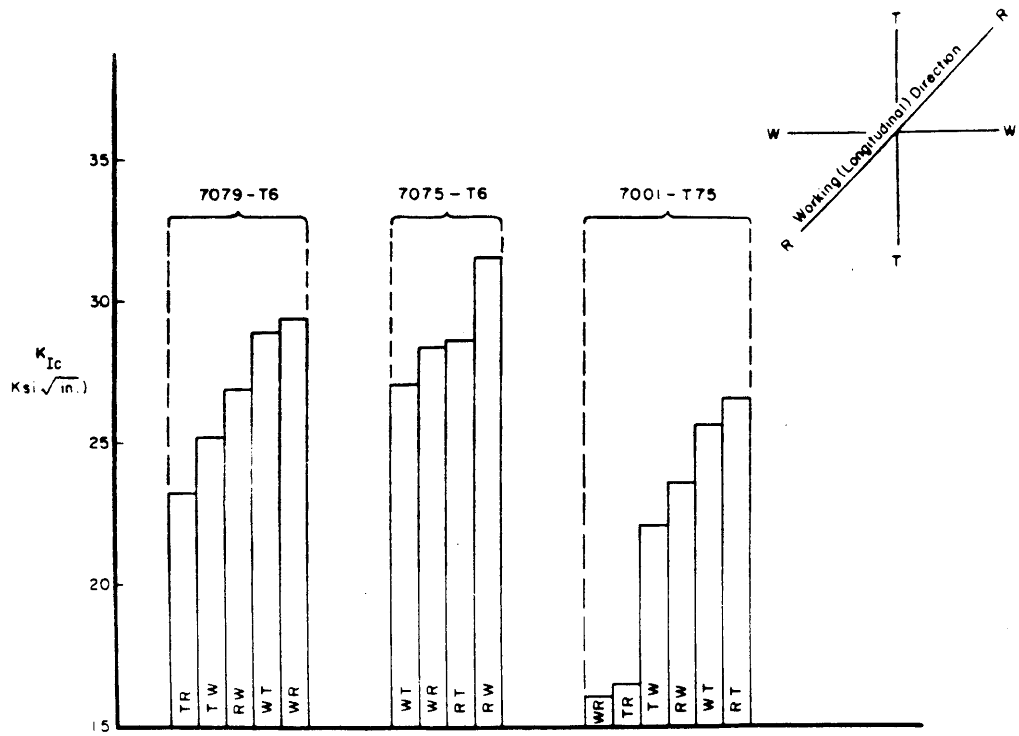


Fig. 19 Variation of K with crack propagation system.

FIGURE 24

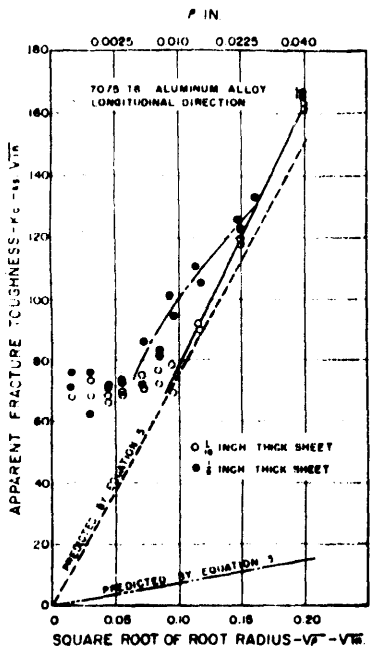


Fig. 2 Apparent fracture toughness as a function of the notch root radius of 1/16 and 1/8 in. specimens, longitudinally oriented, 7075-T6 aluminum alloy

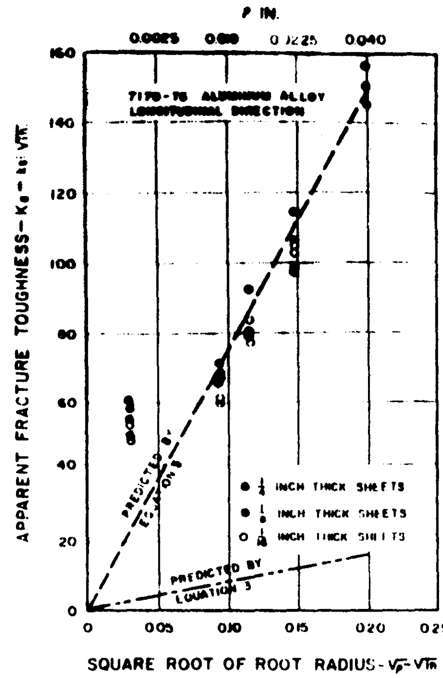
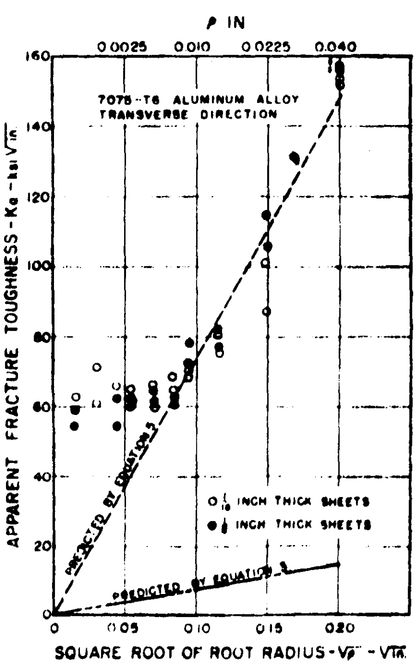


Fig. 4 Apparent fracture toughness as a function of the notch root radius of 1/16, 1/8, and 1/4 in. thick specimens, longitudinally oriented, 7178-T6 aluminum alloy



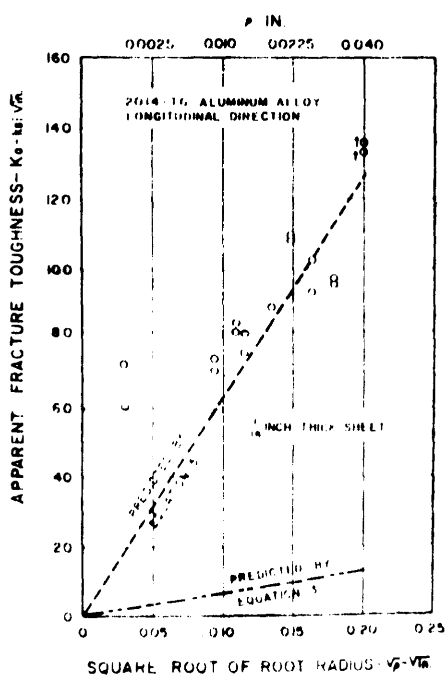


FIGURE 25

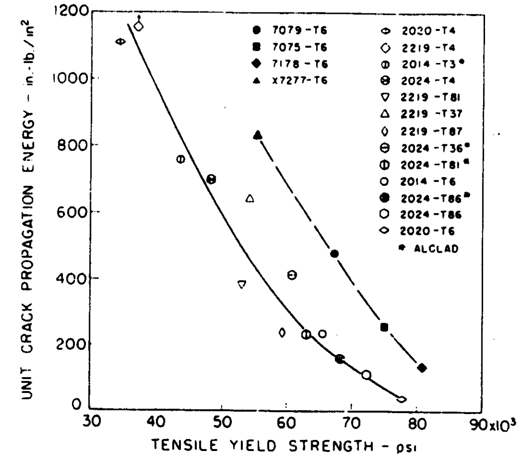


FIGURE 26

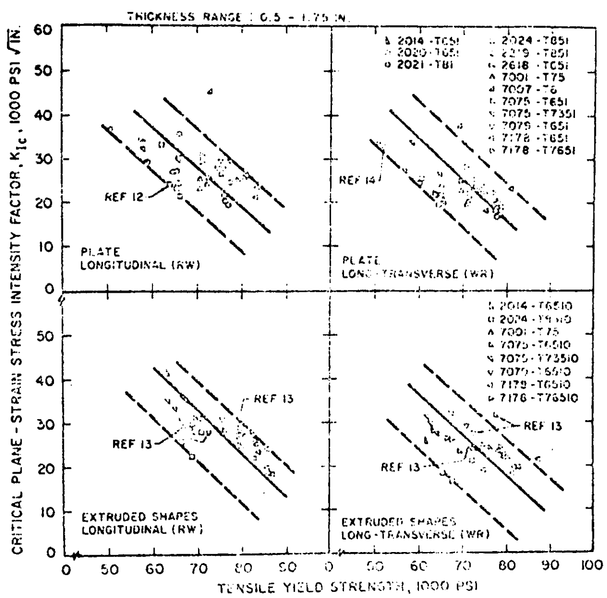


Fig. 2. Plane-strain fracture toughness (K_{Ic}) of commercial aluminum alloy plate and extruded shapes, determined with notched bend, compact tension or single-edge-notched tension specimens essentially within requirements of ASTM proposed test methods.

FIGURE 27

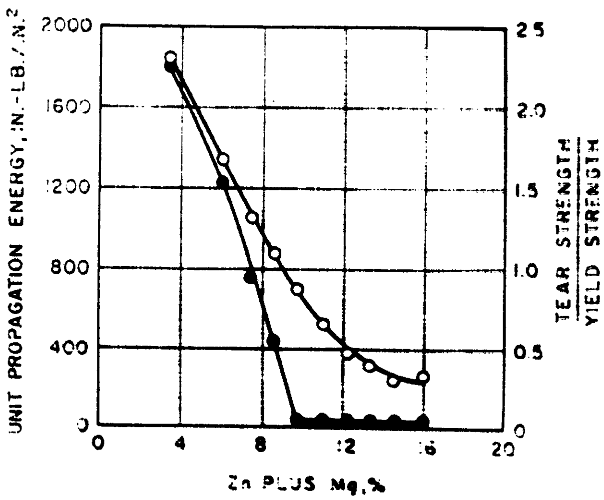


FIGURE 28

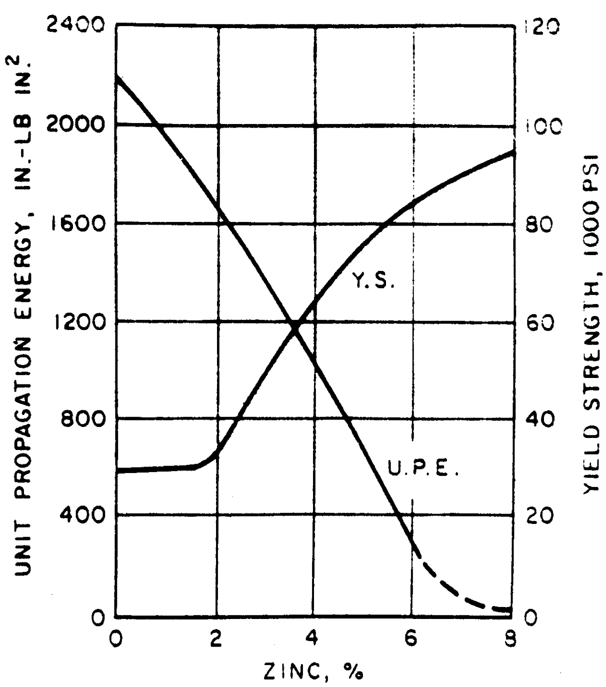


FIGURE 29

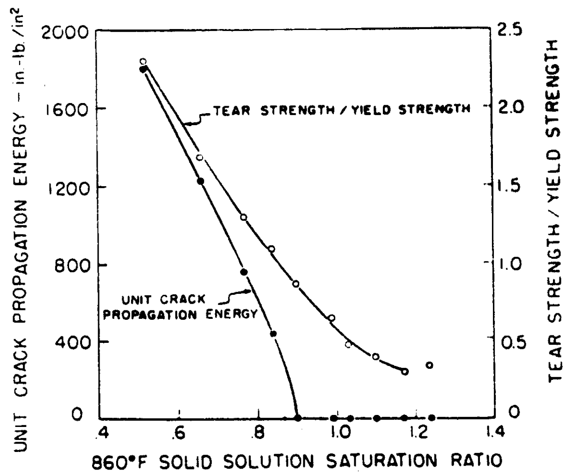
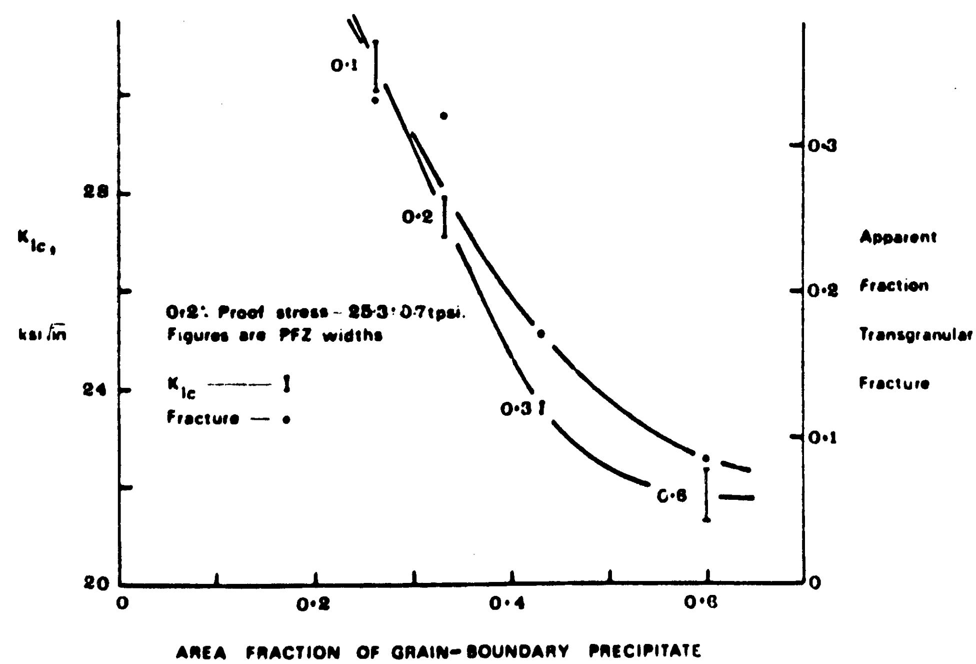


Fig. 9—Above—Unit crack propagation energy values and tear strength-yield ratios for experimental alloy series.

FIGURE 30



The trend of increasing fracture toughness together with increasing proportion of transgranular fracture as the area fraction of grain-boundary precipitate is reduced.

FIGURE 31

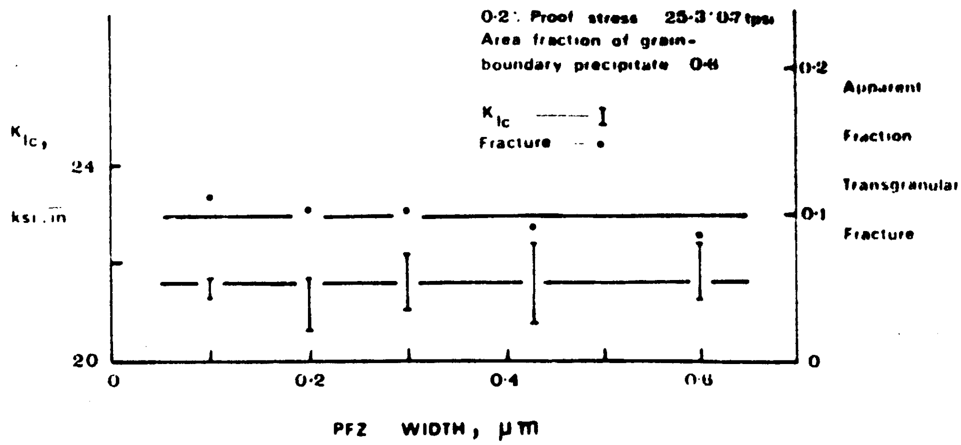


Fig. 7 Fracture toughness and apparent fraction of transgranular fracture plotted as a function of P.F.Z. width.

FIGURE 32

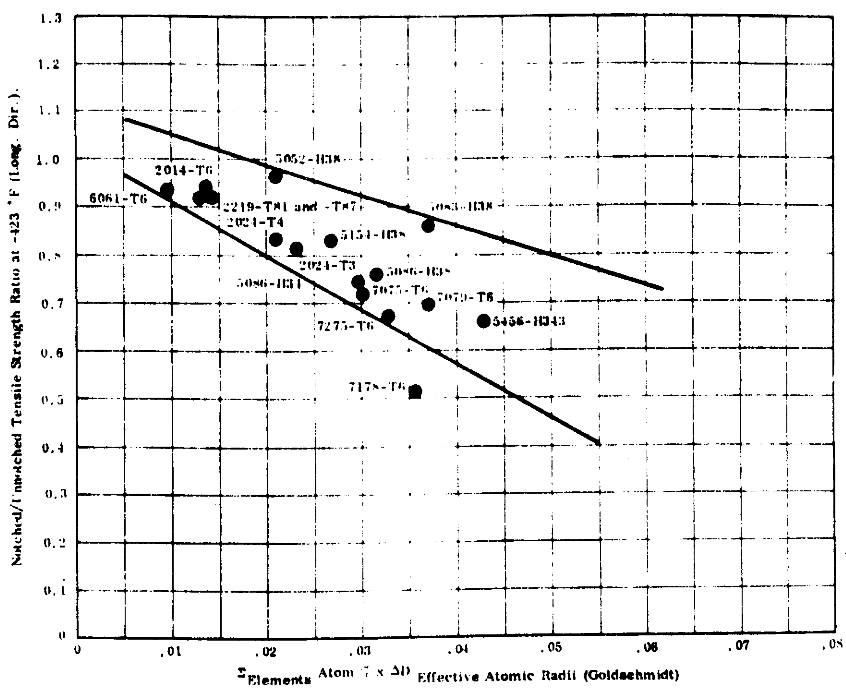


Fig. 6. Effect of atomic radii of alloying elements on toughness of aluminum alloys.

FIGURE 33

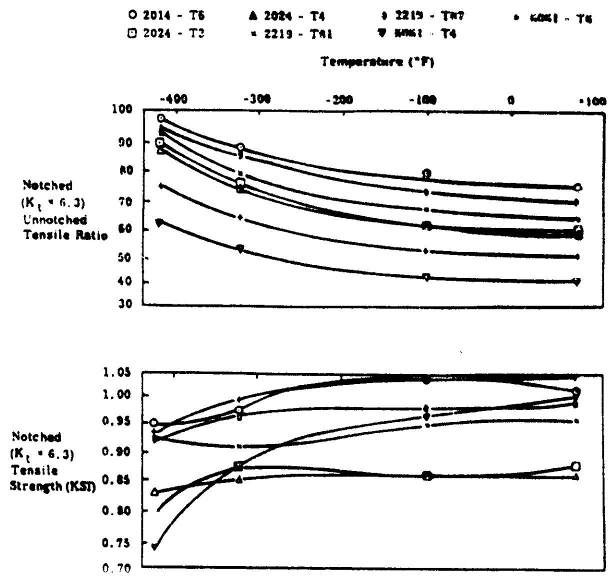


Fig. 2. Notched tensile strengths and notched/unnotched tensile ratio *vs.* temperature.

FIGURE 34

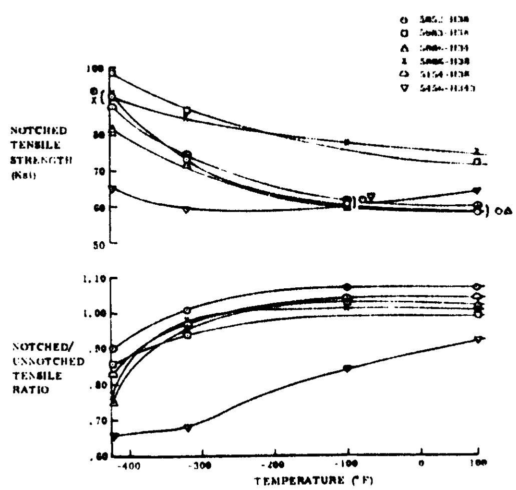


Fig. 2. Notched (K_t=6.3) tensile strengths and notched unnotched tensile ratios vs. temperature.

FIGURE 35

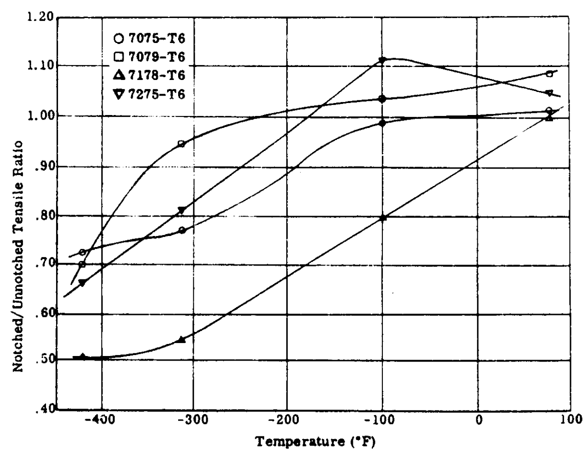


Fig. 3. Notched tensile strengths (ksi) and notched/unnotched tensile ratios for 7000 series aluminum alloy sheet vs. temperature.

FIGURE 36

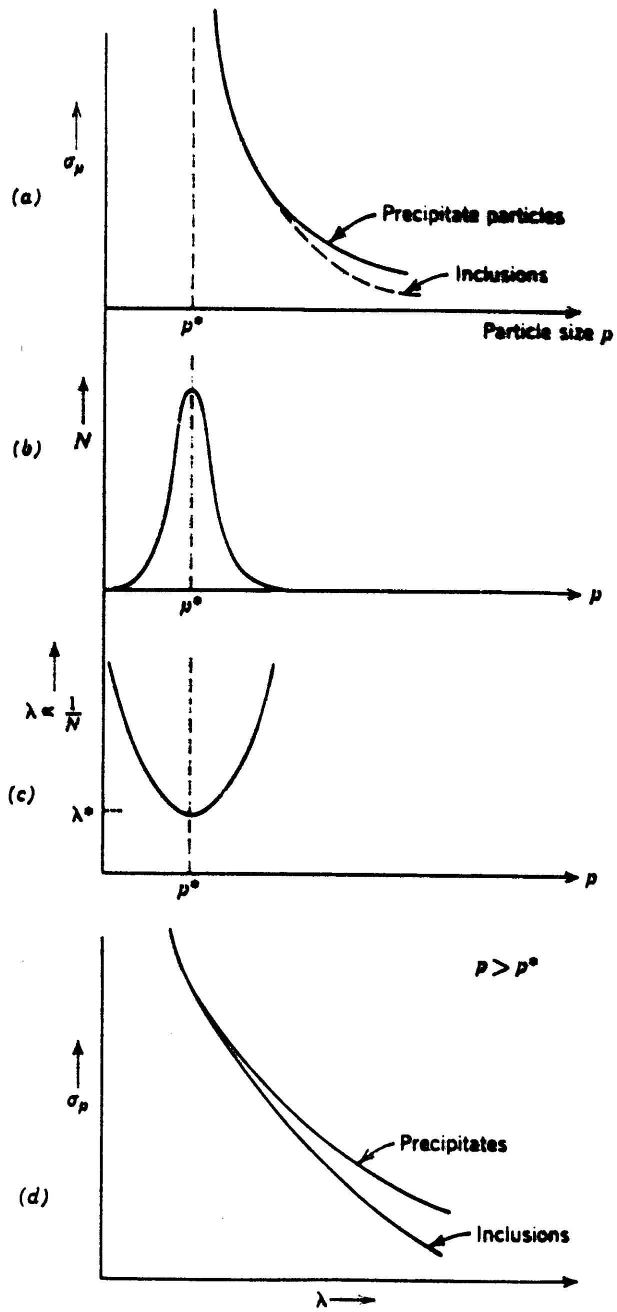


Fig. 7.23. (*a*) Schematic variation of the fracture strength of a hard particle or inclusion σ_p with particle size p. (*b*) Typical distribution of particle sizes about an average value of particle size p^*. (*c*) The variation of λ, the mean distance between two particles of the same size, with particle size p. (*d*) The variation of σ_p with λ for $p > p^*$.

FIGURE 37

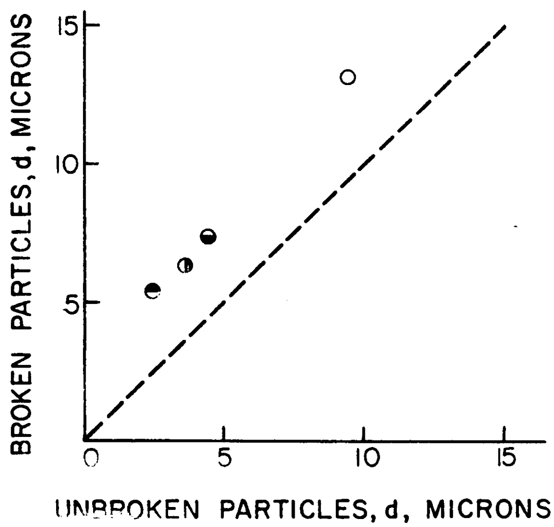


Fig. 5—Comparison of average diameters of broken and unbroken silicon particles in alloy of 6.1 pct Si.

FIGURE 38

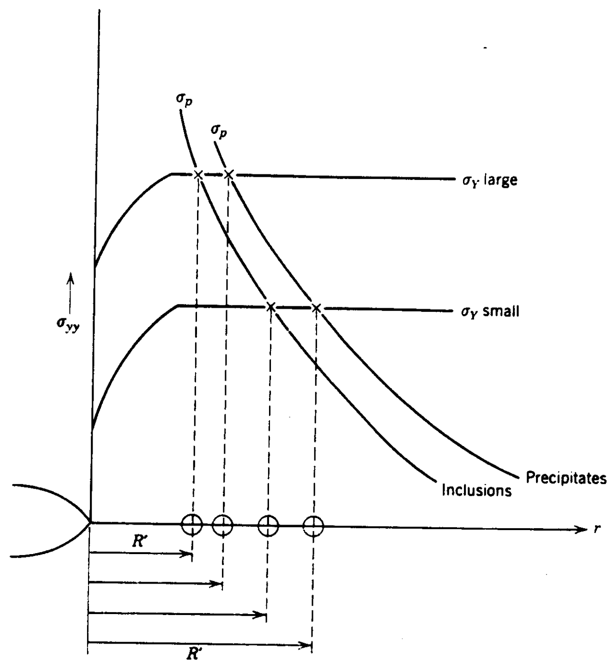
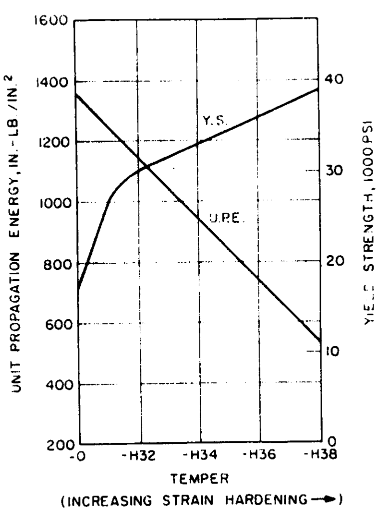


Fig. 7.24. The variation of R' with yield strength level σ_Y and the presence or absence of inclusions, in high strength materials. R' is the distance from the notch tip at which $\sigma_{yy} = \sigma_p$ and a void can be

FIGURE 39



1600
1400
1200
1000
800
600
400
200
UNIT PROPAGATION ENERGY, IN.-LB /IN.^{2}
40
30
20
10
0
YIELD STRENGTH, 1000 PSI
Y.S.
U.P.E.
-0
-H32
-H34
-H36
-H38
TEMPER
(INCREASING STRAIN HARDENING →)

FIGURE 40

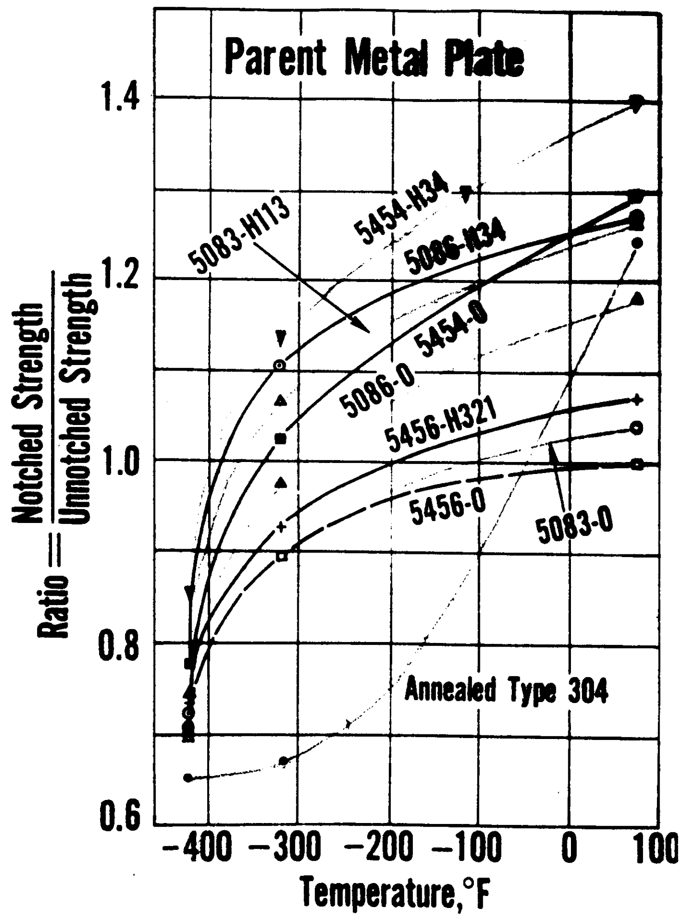


FIGURE 41

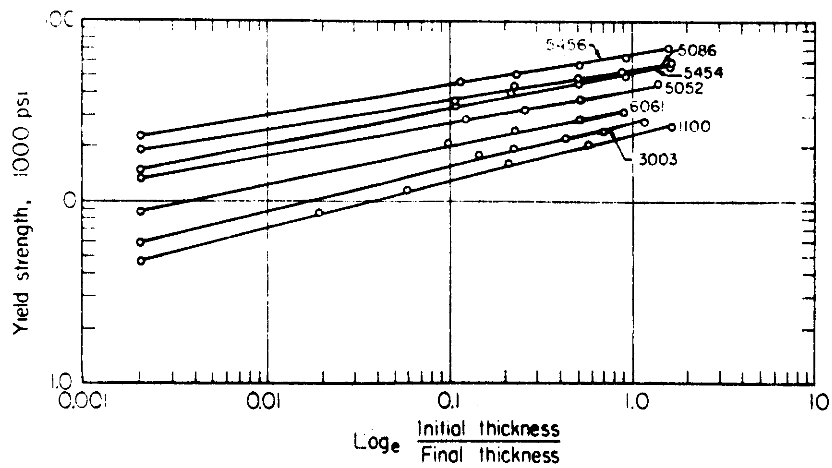


FIGURE 42

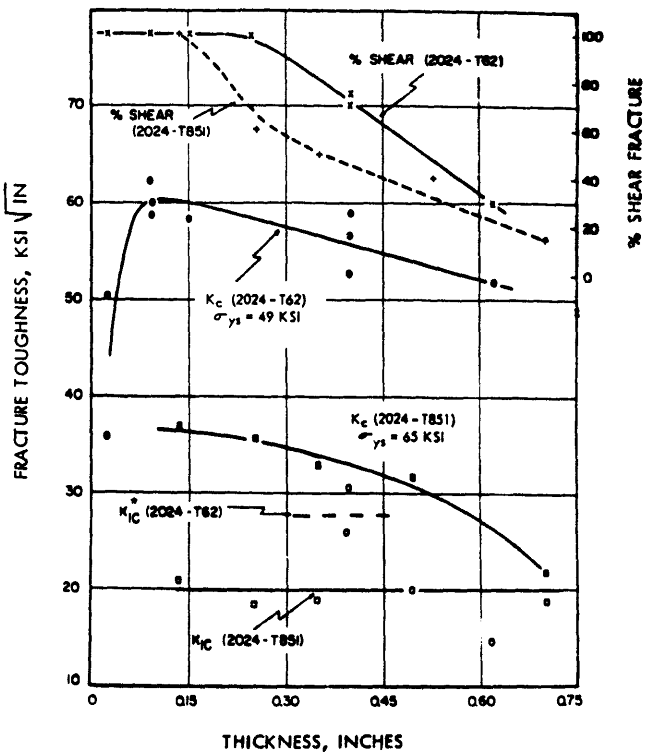


Fig. 7—Influence of center-notch tensile specimen thickness on the fracture toughness of 2024 aluminum (T62 and T851

FIGURE 43

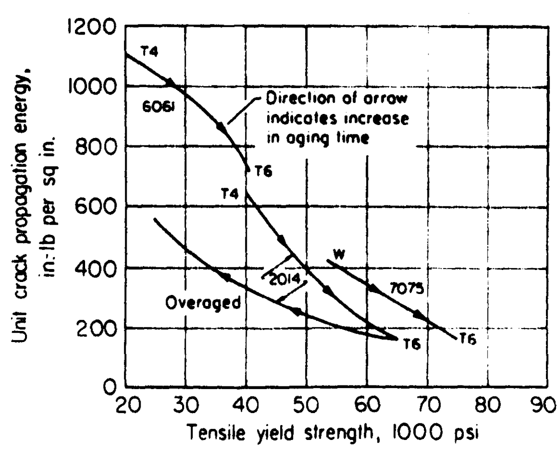


Fig. 24. Effects of precipitation heat treatment on unit crack propagation energy and yield strength of 2014, 6061 and 7075 alloys (J. A. Nock, Jr., and H. Y. Hunsicker, Journal of Metals, 15, 216–224, 1963)

FIGURE 44

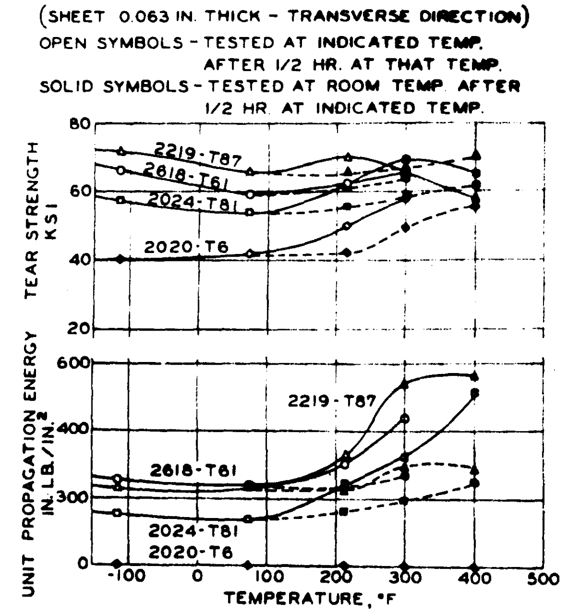


Fig. 8 - Effect of various temperatures on the tear resistance of some aluminum alloys (sheet, 0.063 in. thick - transverse direction)

FIGURE 45

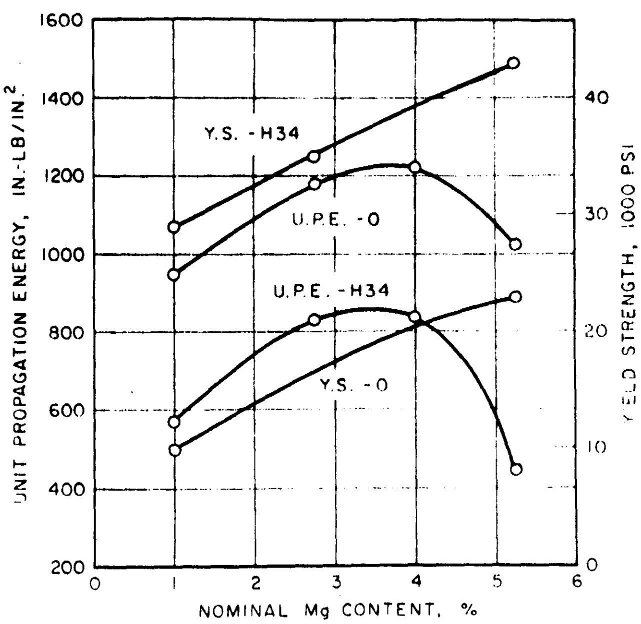


FIGURE 46

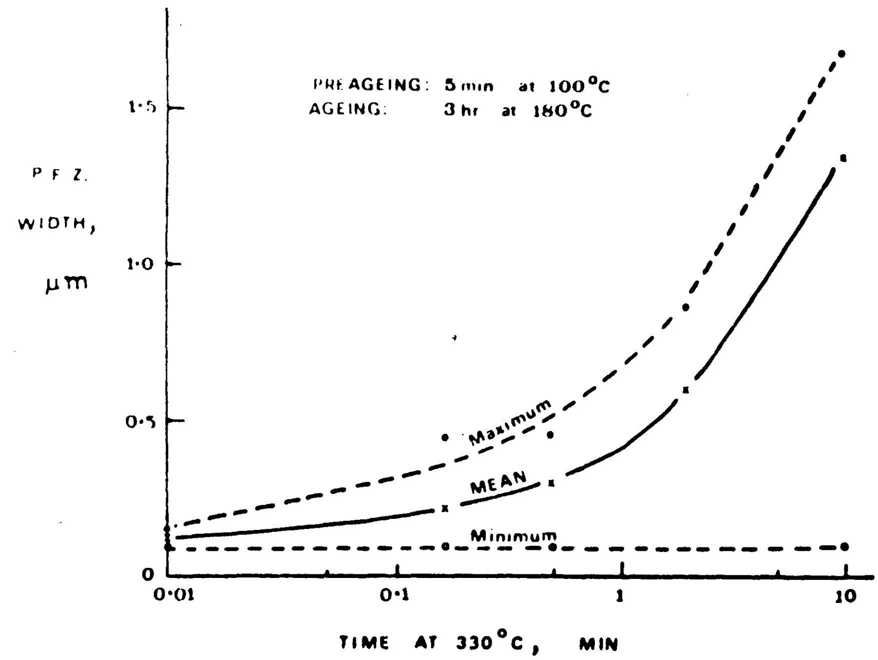
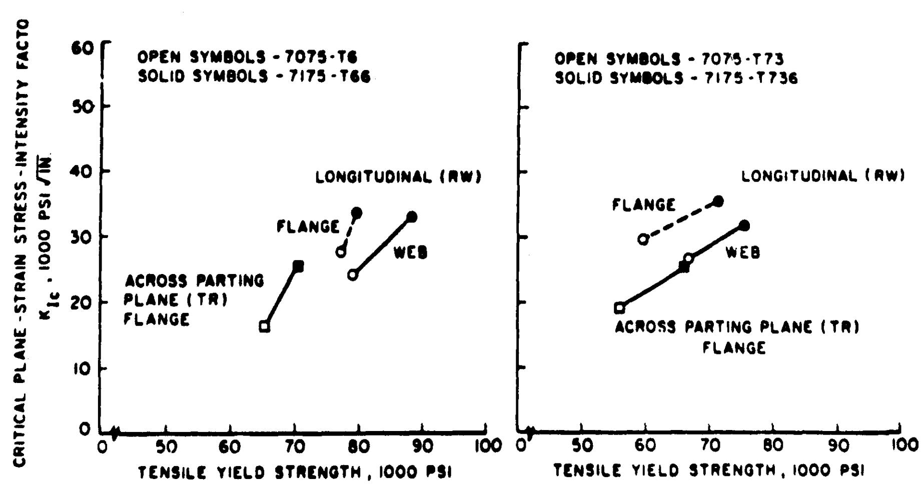


FIGURE 47



Plane-strain fracture toughness of 7175-T66 and T736 die forgings compared with that of commercial 7075-T6 forgings.

FIGURE 48

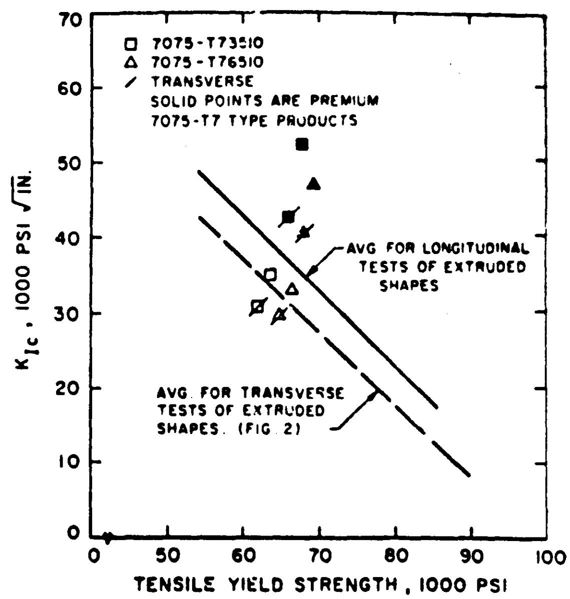


Fig. 7. Plane-strain fracture toughness of premium 7075-T73510 and T76510 extruded shapes compared with that of conventional 7075-T73510 and T76510 extruded shapes.

FIGURE 49

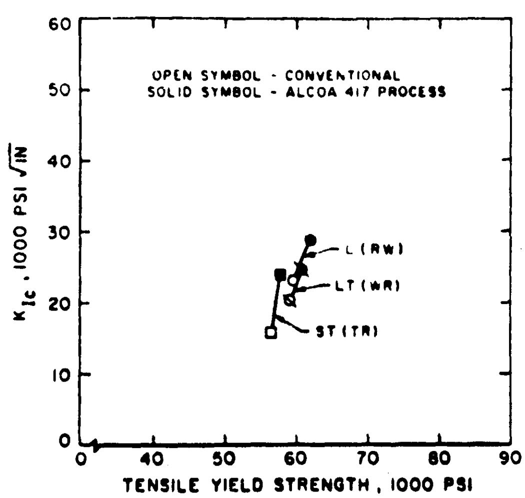


Fig. 9. Plane-strain fracture toughness of 4-in. Alcoa 417 Process 2024-T851 plate compared with that of conventional 4-in. 2024-T851 plate.

FIGURE 50

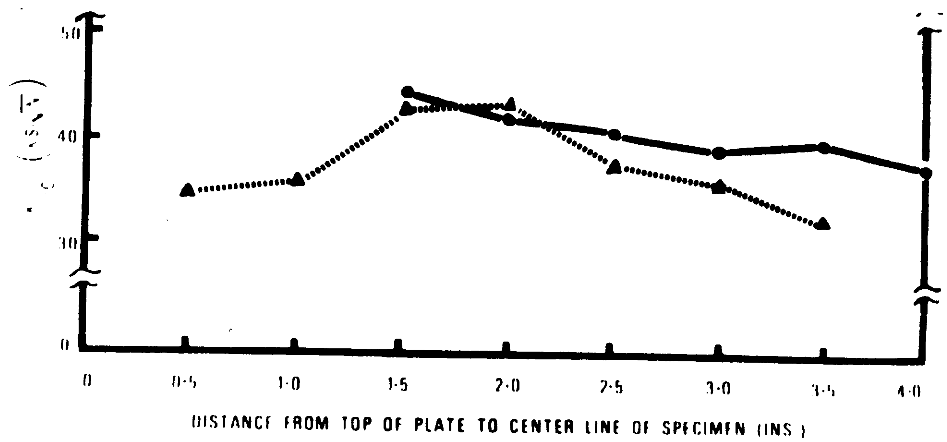


FIGURE 51

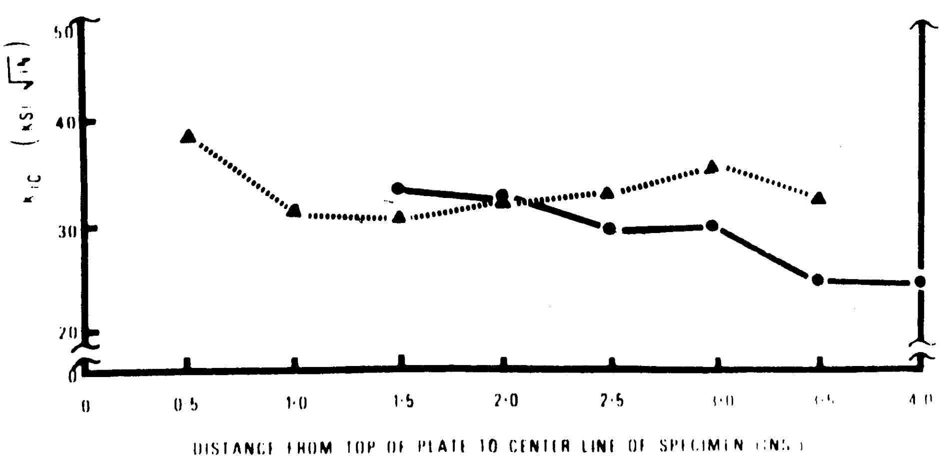


FIGURE 52

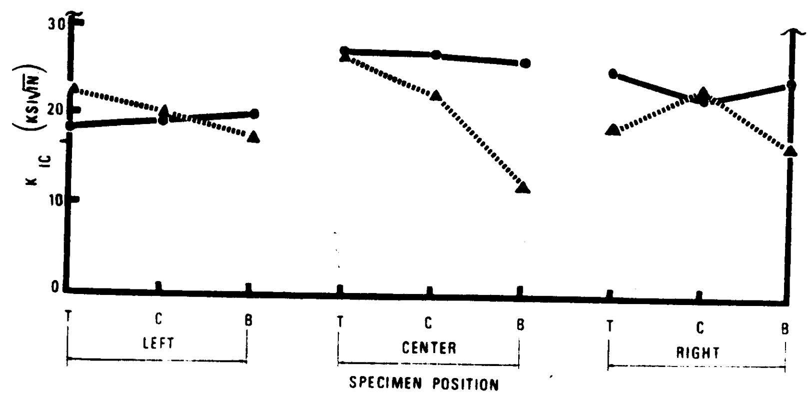


Fig. 7. 7179-T651 plate—short transverse yield strength, elongation and toughness variation of preforged (PF) and standard rolled (SR).

FIGURE 53

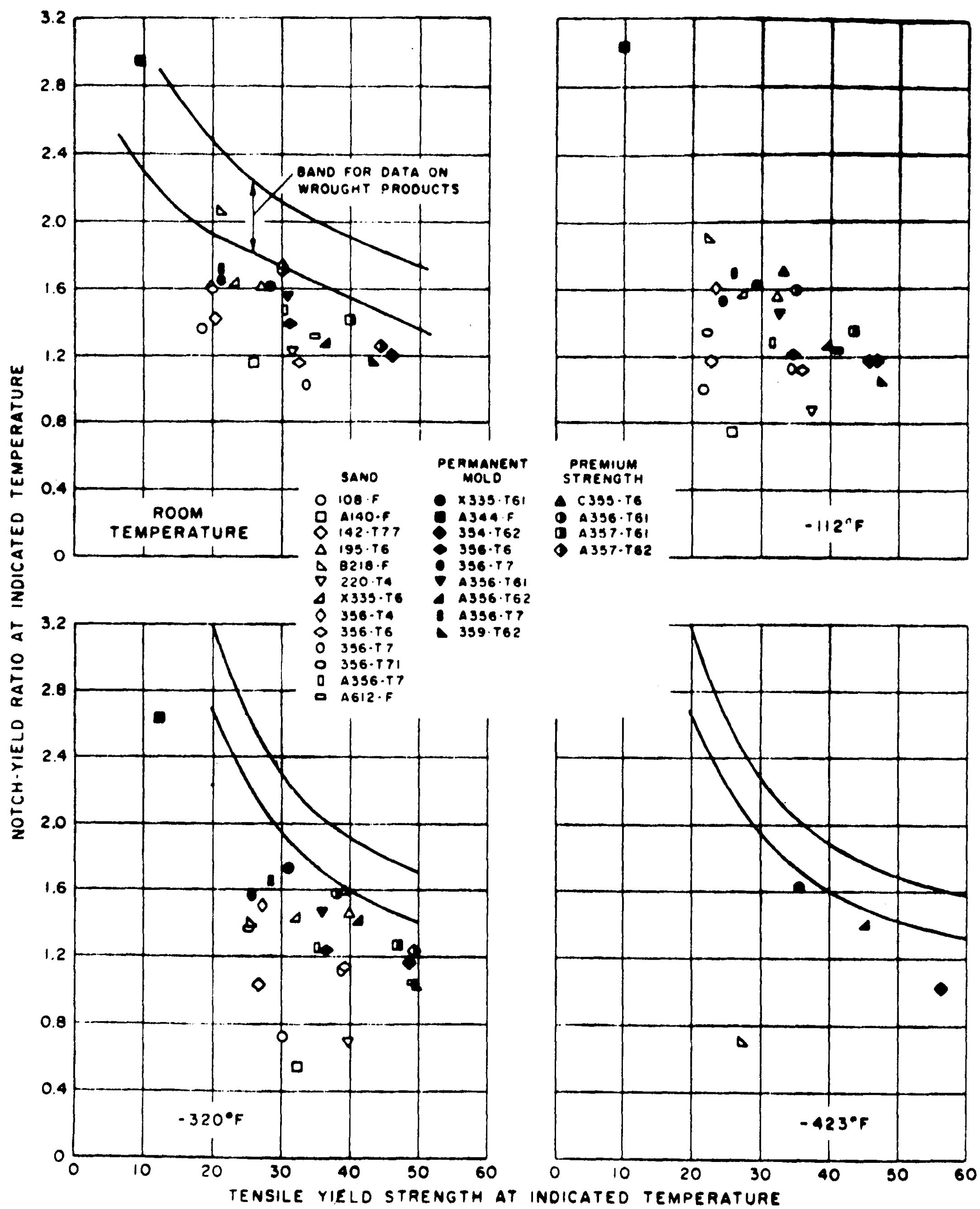


Fig. 5. Notch-yield ratio *vs* tensile yield strength for cast aluminum alloys at various temperatures.

FIGURE 54

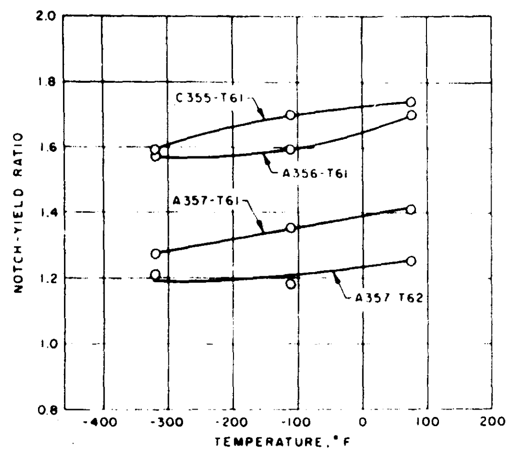


Fig. 4. Notch-yield ratio *vs.* temperature for premium strength cast aluminum alloy slabs.

FIGURE 55

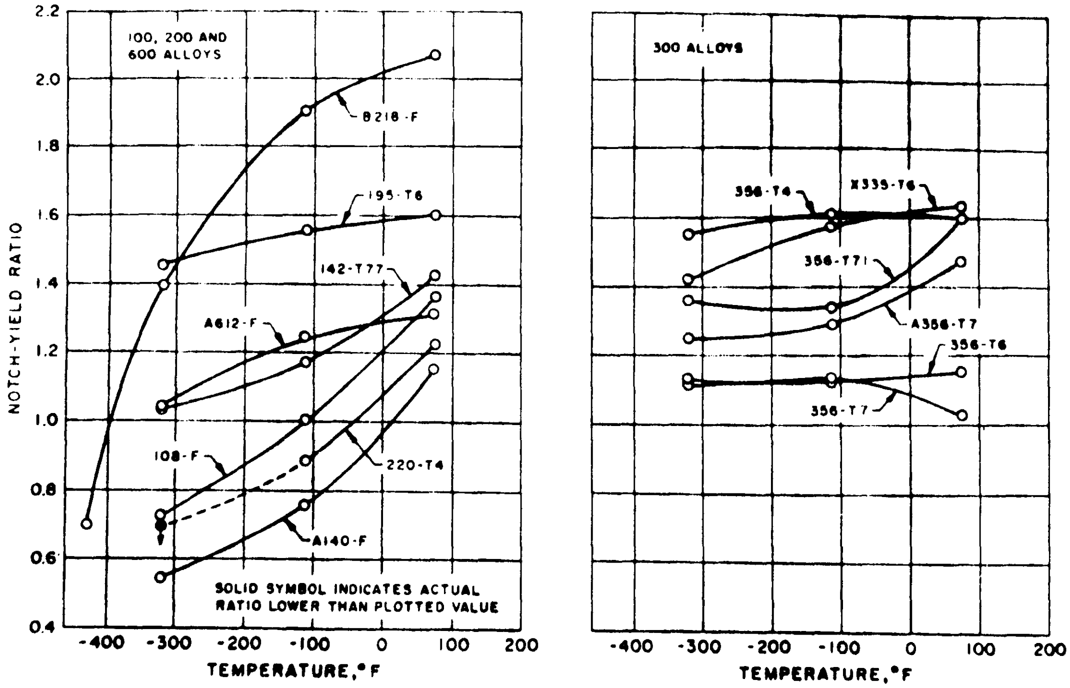


Fig. 2. Notch-yield ratio *vs.* temperature for sand-cast aluminum alloy slabs.

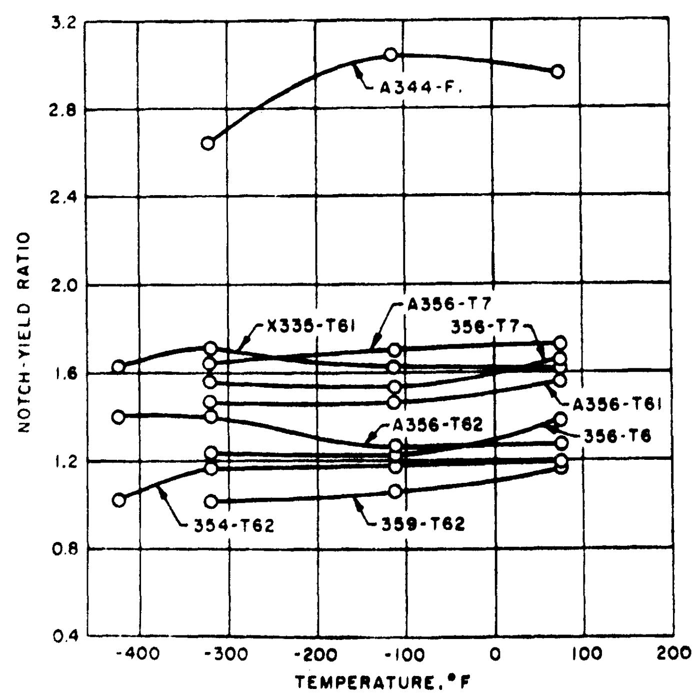


Fig. 3. Notch-yield ratio *vs.* temperature for permanent mold cast aluminum alloy slabs.

FIGURE 56

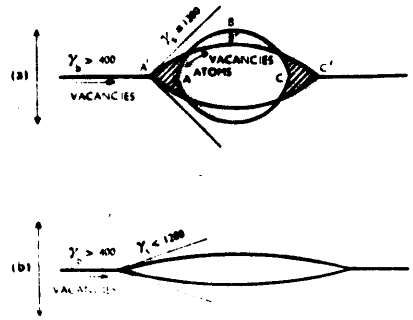


Fig. 19 Pore shape developed under plastic-flow conditions at elevated temperatures. (a) ***Ductile fracture:*** (b) ***brittle*** *fracture associated with reduction in free surface energy by adsorbed impurity.*

FIGURE 57

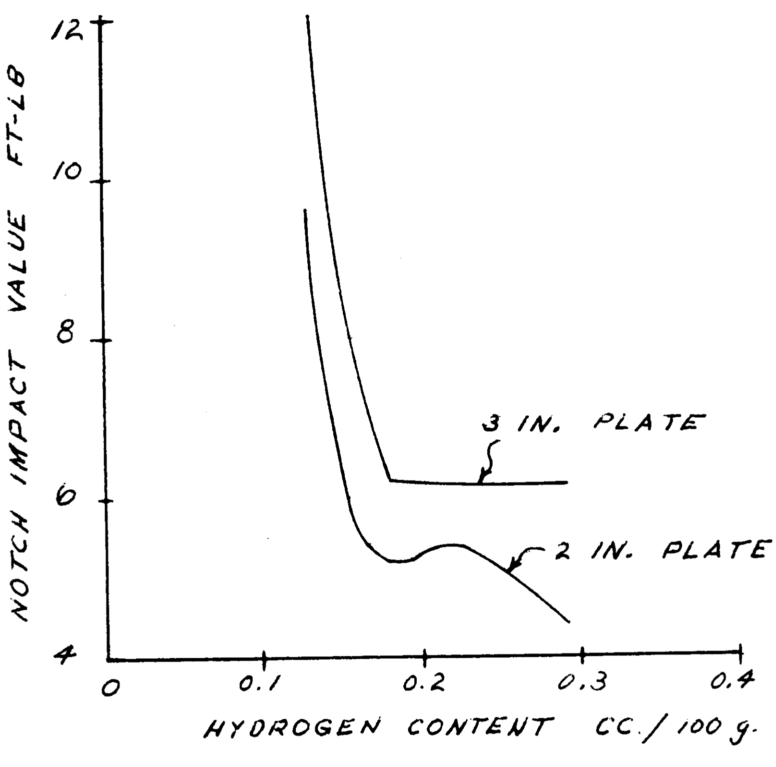


FIGURE 58

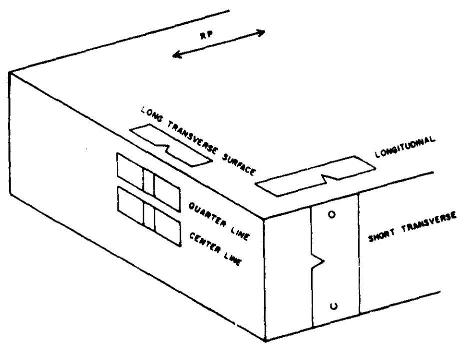


Fig. 18 Sketch showing orientation of single edge-notched specimens in $1^{3}/_{4}$-in-thick plate of 7075-T6 aluminum alloy

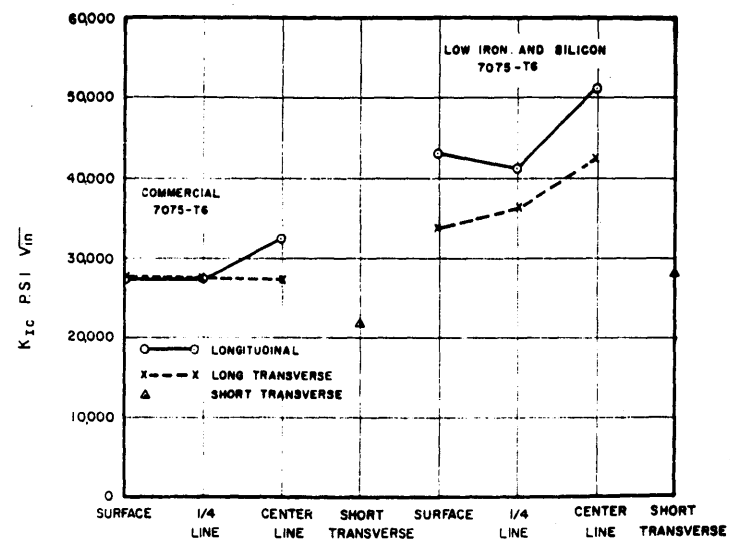


Fig. 19 Plane-strain fracture toughness of 7075-T6 aluminum alloys as a function of specimen location

FIGURE 59

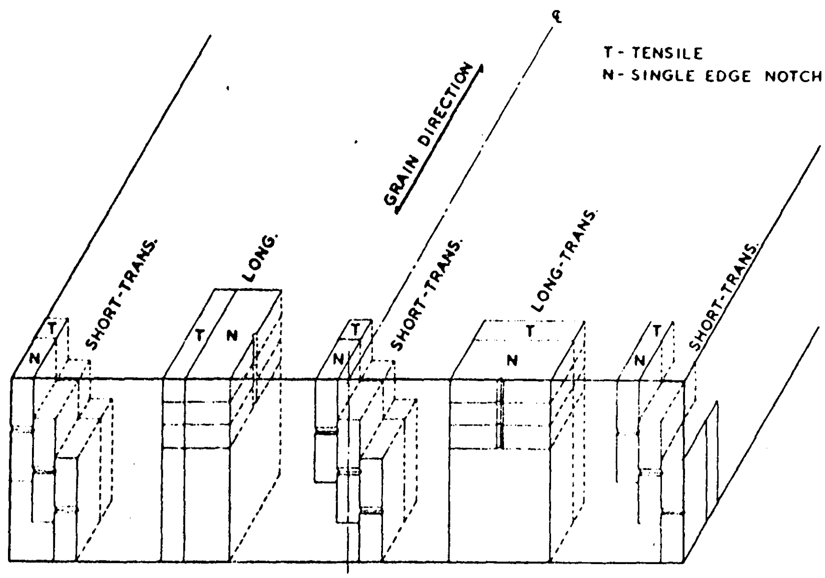


Fig. 4. Specimen location – 7179 alloy plate.

FIGURE 60

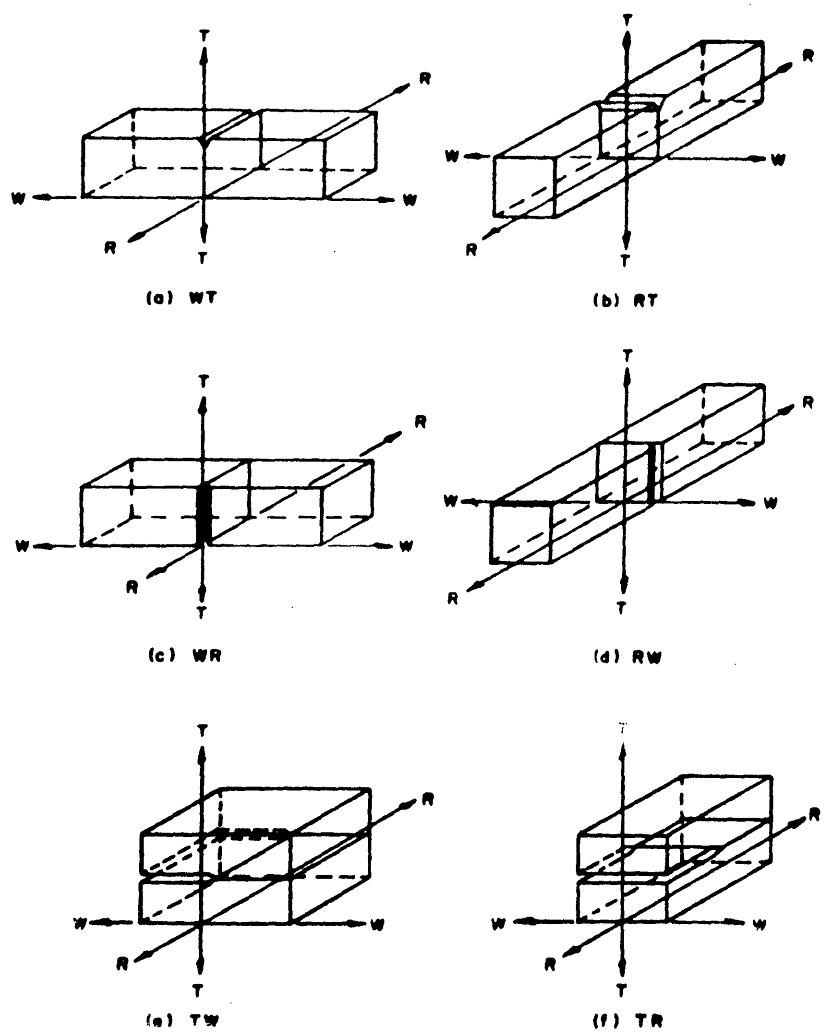
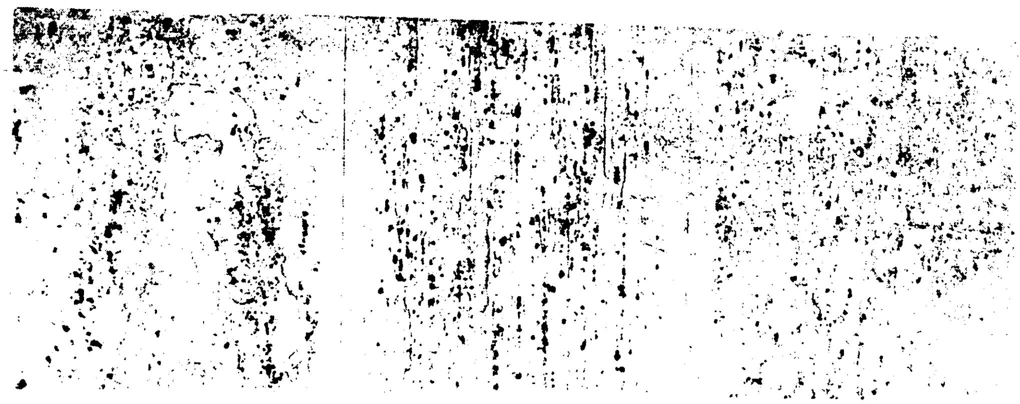


Fig. 3—Conventional crack propagation systems.

FIGURE 61



a. Parallel to rolled surface b. Longitudinal section c. Transverse section
Fig 3—x100 (Enlarged x1. 4 in reproduction) Alloy 7075 plate, etched 0.5 pct HF

FIGURE 62

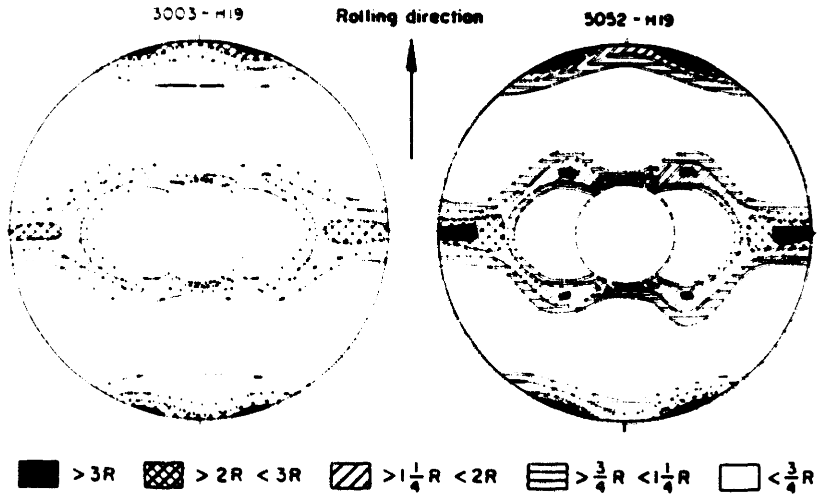


Fig. 3. {111} pole figures of 3003-H19 and 5052-H19 sheet. Densities of poles are in multiples of random concentration, R.

FIGURE 63

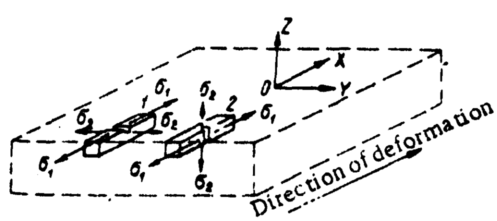


Fig. 1. Orientation of the notch (crack) relative to the plane of hot deformation: 1) notch (crack) parallel to the deformation plane (XOY); 2) notch (crack) normal to the deformation plane (XOY); OX = direction of rolling or pressing.

FIGURE 64

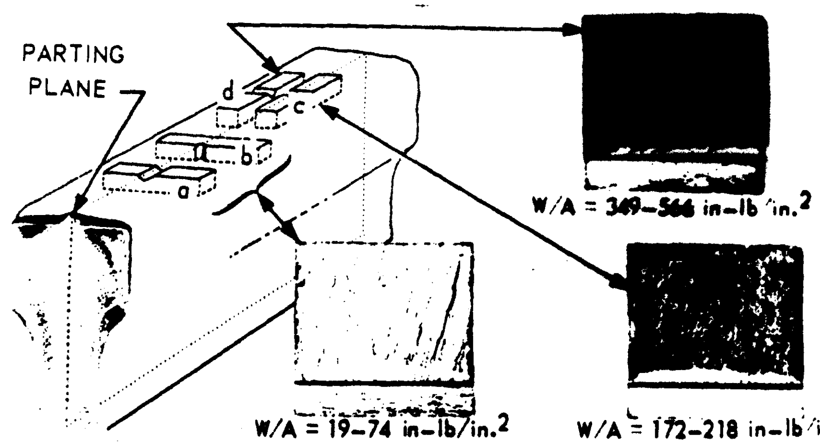


FIG. 1. Location and fracture appearance of precracked Charpy V-notch specimens.

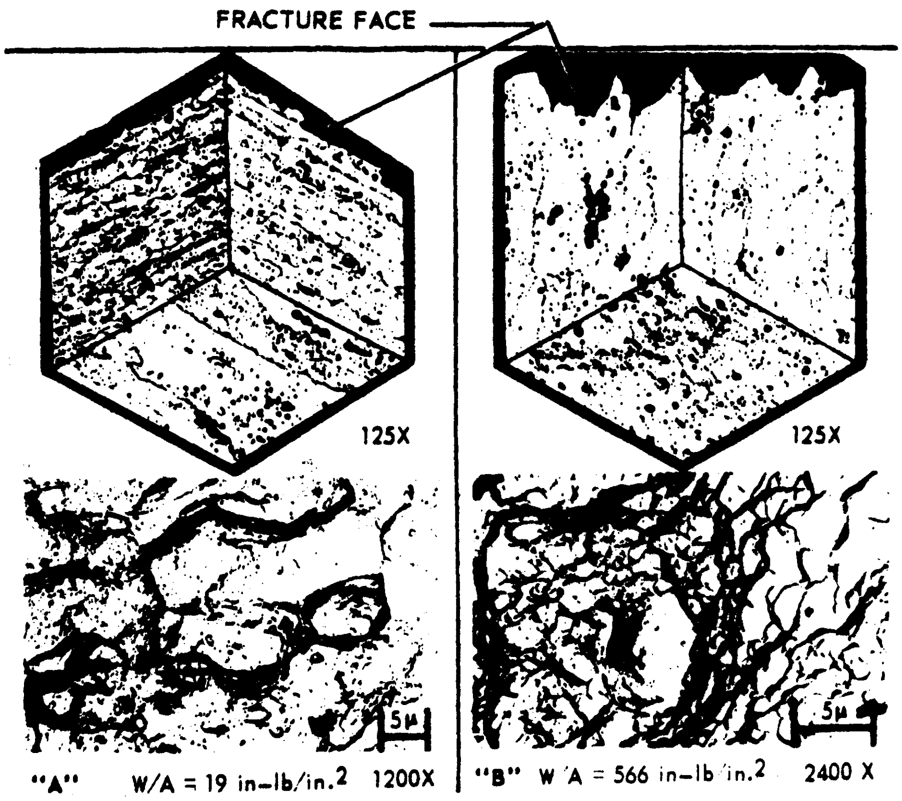


FIG. 2. Fracture profiles and topographies of low toughness (A) and high toughness (B) specimens.

FIGURE 65

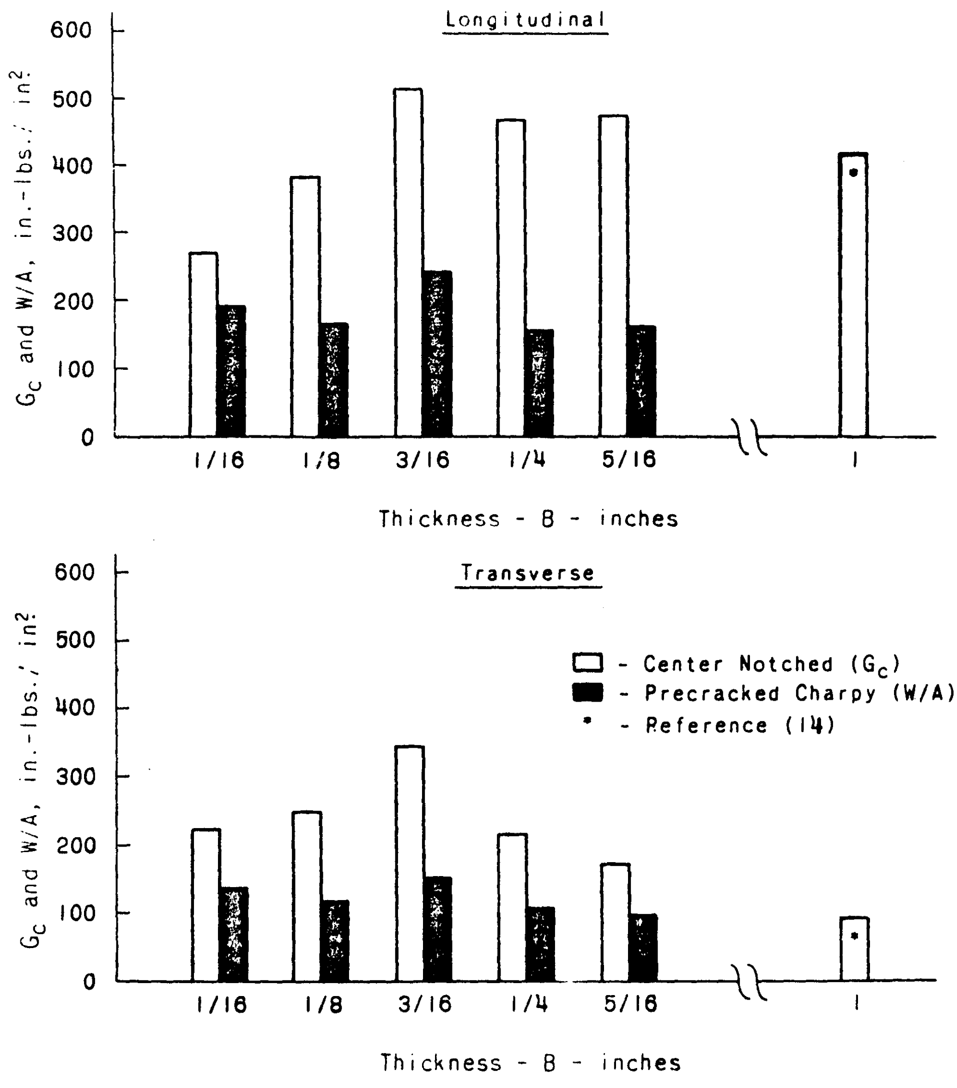


Fig. 4. Toughness anisotropy and thickness comparisons showing trends of both static (G_c) and dynamic (W/A) tests on 7075-T6 and -T651.

FIGURE 66

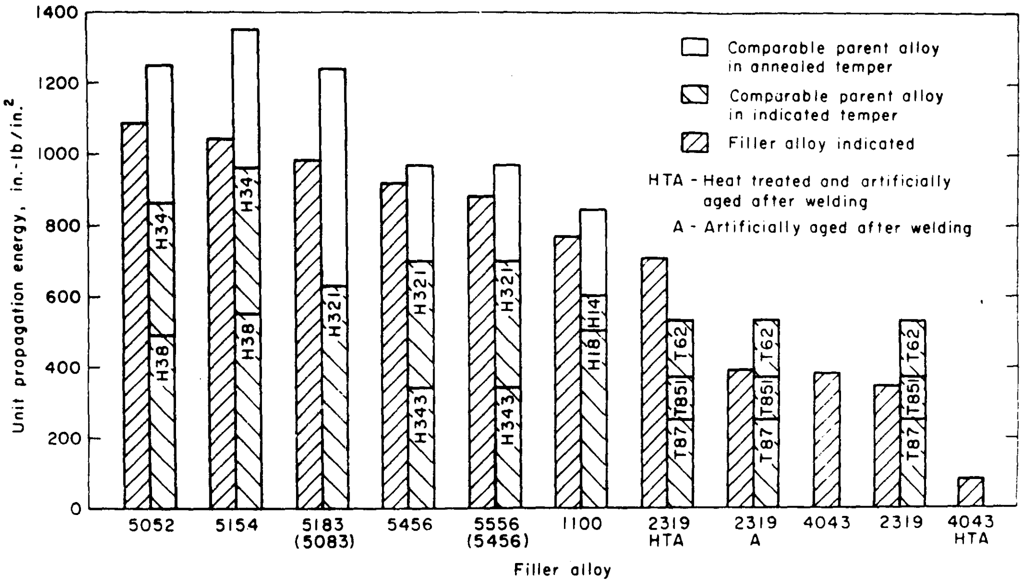


Fig. 4—Ratings of aluminum alloy welds based upon unit propagation energy from tear tests

FIGURE 67

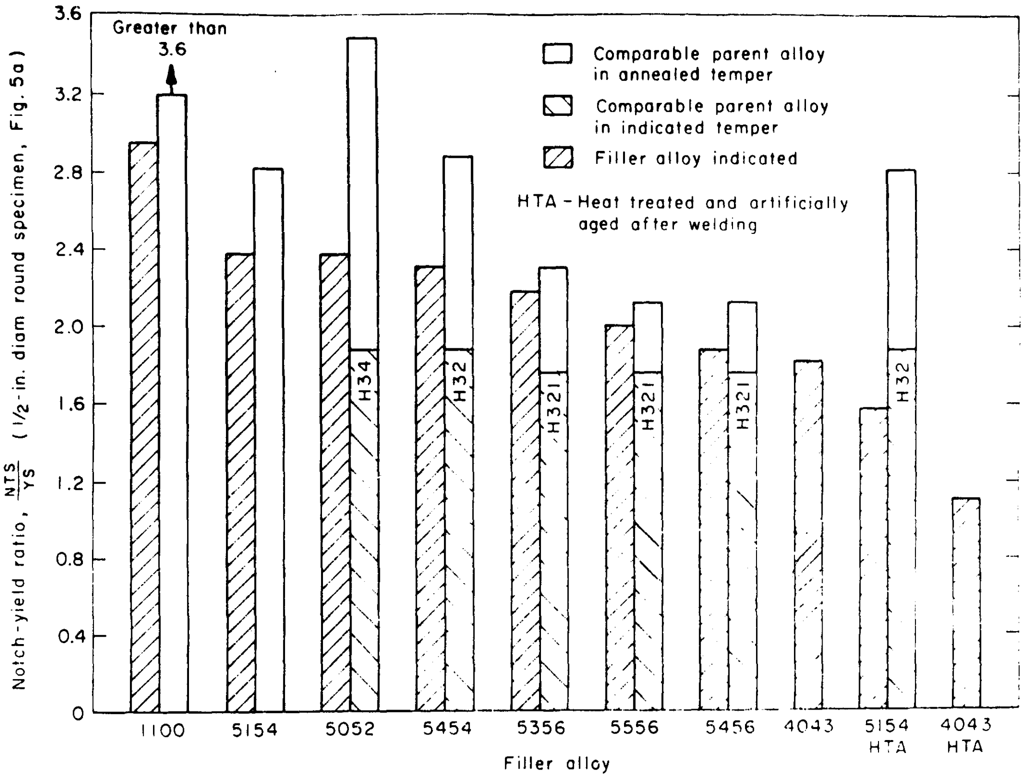


Fig. 7—Ratings of aluminum alloy welds based upon notch-yield ratio from round specimens

FIGURE 68

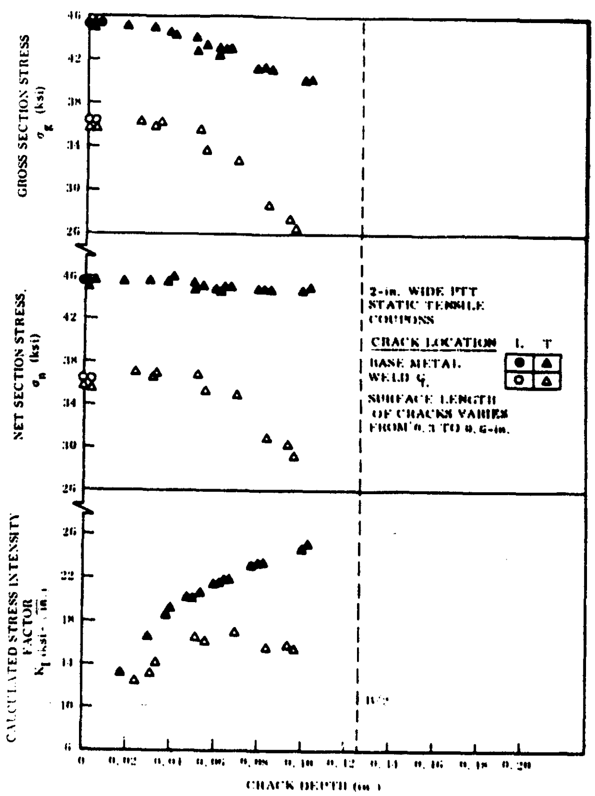


Fig. 9 – Effect of Surface Crack Depth on Toughness in 6061 Aluminum Sheet and Welds

FIGURE 69

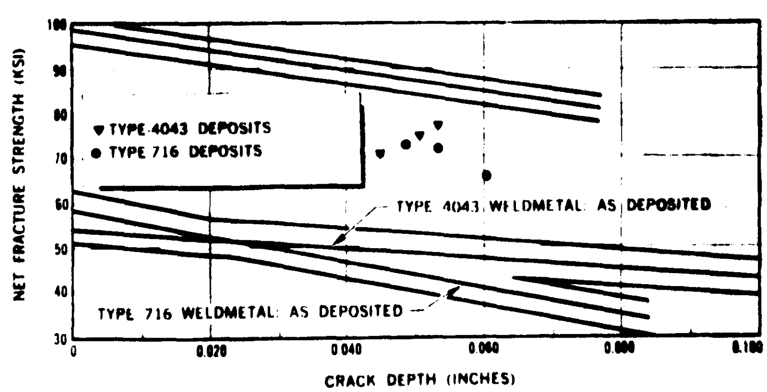


FIGURE 70

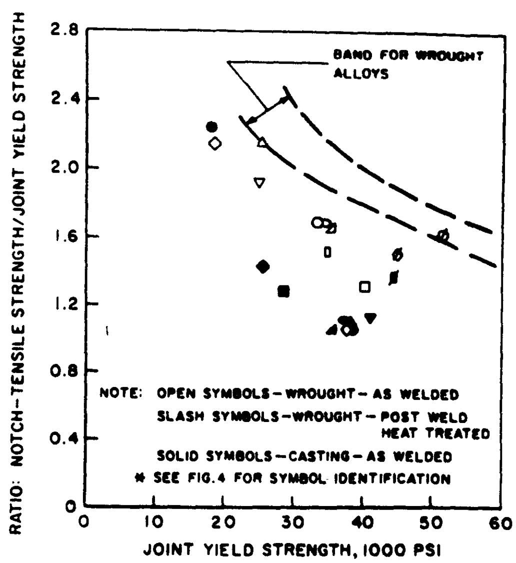


Fig. 5. Joint yield strength *vs.* notch yield ratios for groove welds in wrought and casting alloys at −452°F.

FIGURE 71

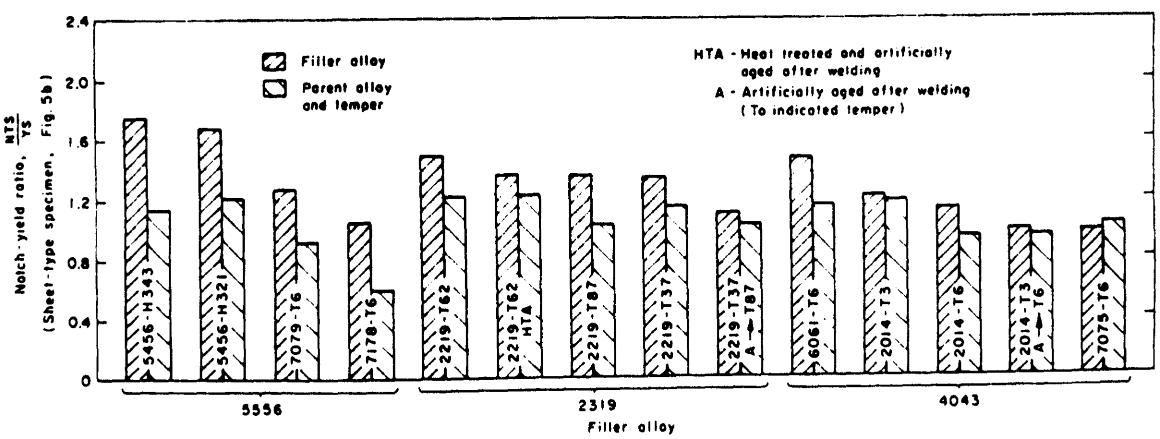
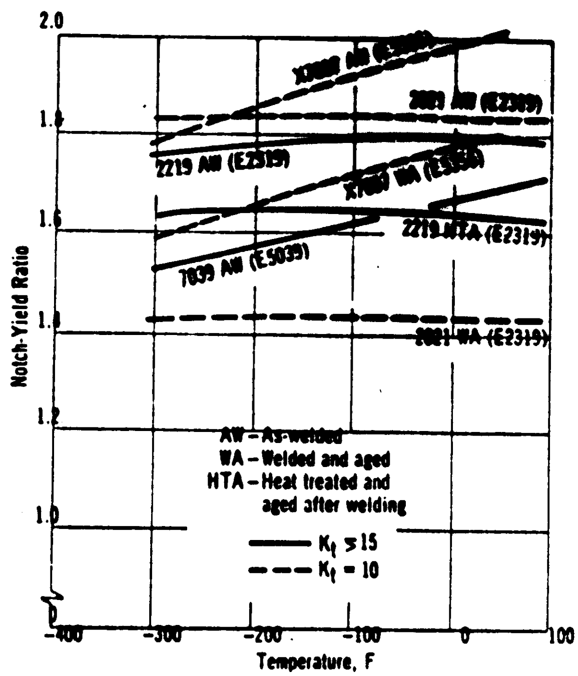
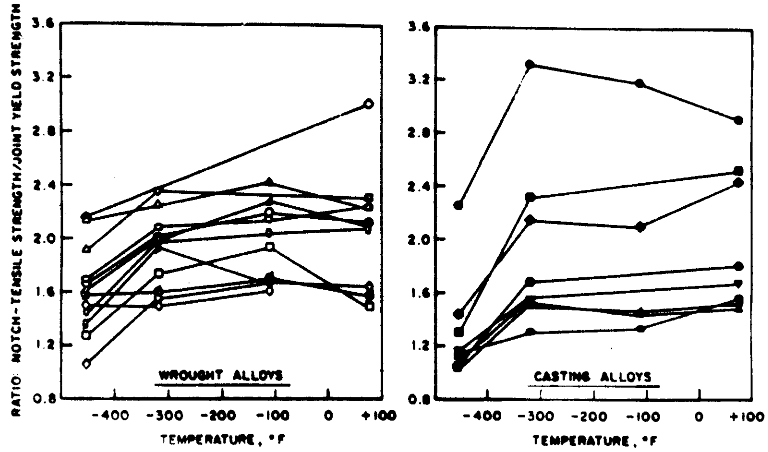


Fig. 8—Ratings of aluminum alloy welds based upon notch-yield ratio from sheet-type specimens

FIGURE 72





SYMBOL	WROUGHT ALLOY	FILLER ALLOY	POST-WELD THERMAL TREATMENT
○	2219-T62	2319	YES *
□	2219-T851	2319	NO
◇	3003-H112	1100	NO
△	5083-0	5183	NO
◿	5083-H321	5183	NO
○	5083-H321	5356	NO
○	5083-H321	5556	NO
▽	5454-H32	5554	NO
◇	6061-T6	4043	NO AND YES *
□	6061-T6	5356	NO AND YES *

SYMBOL	CAST-TO-CAST OR CAST-TO-WROUGHT ALLOY COMBINATION	FILLER ALLOY
●	A344-F/A344-F	4043
■	A344-F/6061-T6	4043
◆	A344-F/5456-H321	5556
▲	354-T62/354-T62	4043
◢	354-T62/6061-T6	4043
●	354-T62/5456-H321	5556
●	C355-T61/6061-T6	4043
▼	C355-T61/5456-H321	5556

* SLASH SYMBOLS

Fig. 4. Notch yield ratio *vs.* temperature for groove welds in wrought and casting alloys at R.T., − 112°, − 320°, and − 452°F.

FIGURE 73

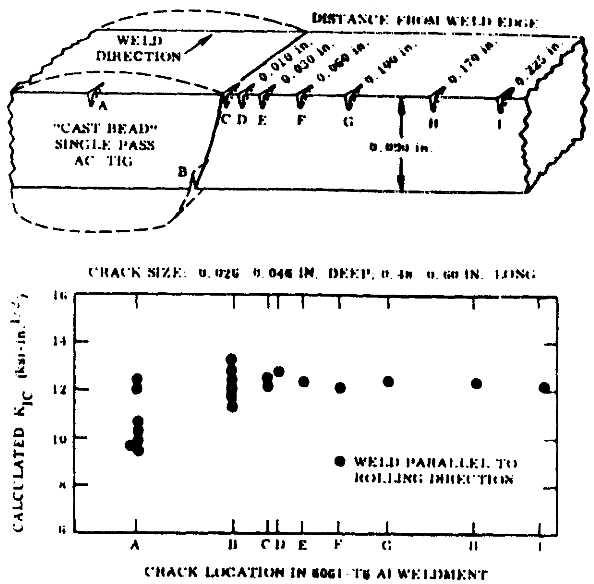


Fig. 11 – Fracture Toughness in an A.C. TIG Weld, "Cast" Into Backup Groove

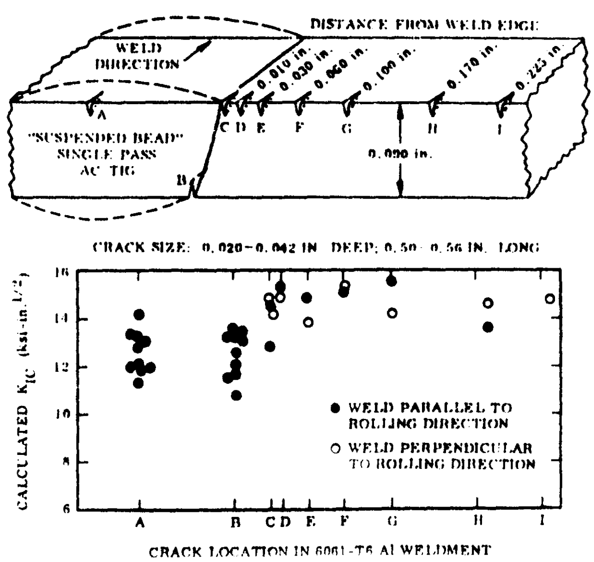
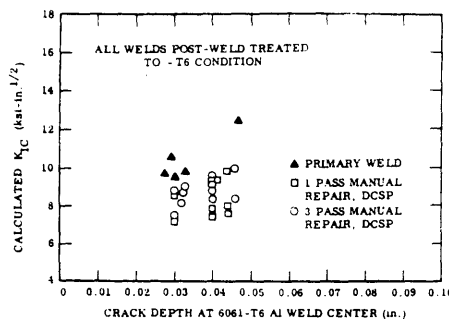


Fig. 12 – Fracture Toughness in an A.C. TIG Weld, "Suspended Bead"

FIGURE 74

Fig. 14 — Effect of Surface Crack Depth on Calculated K_{Ic} in One- and Two-Pass Manual Repair Welds



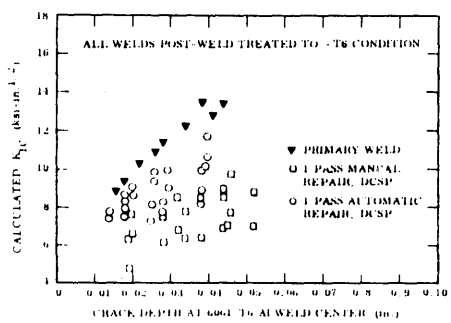
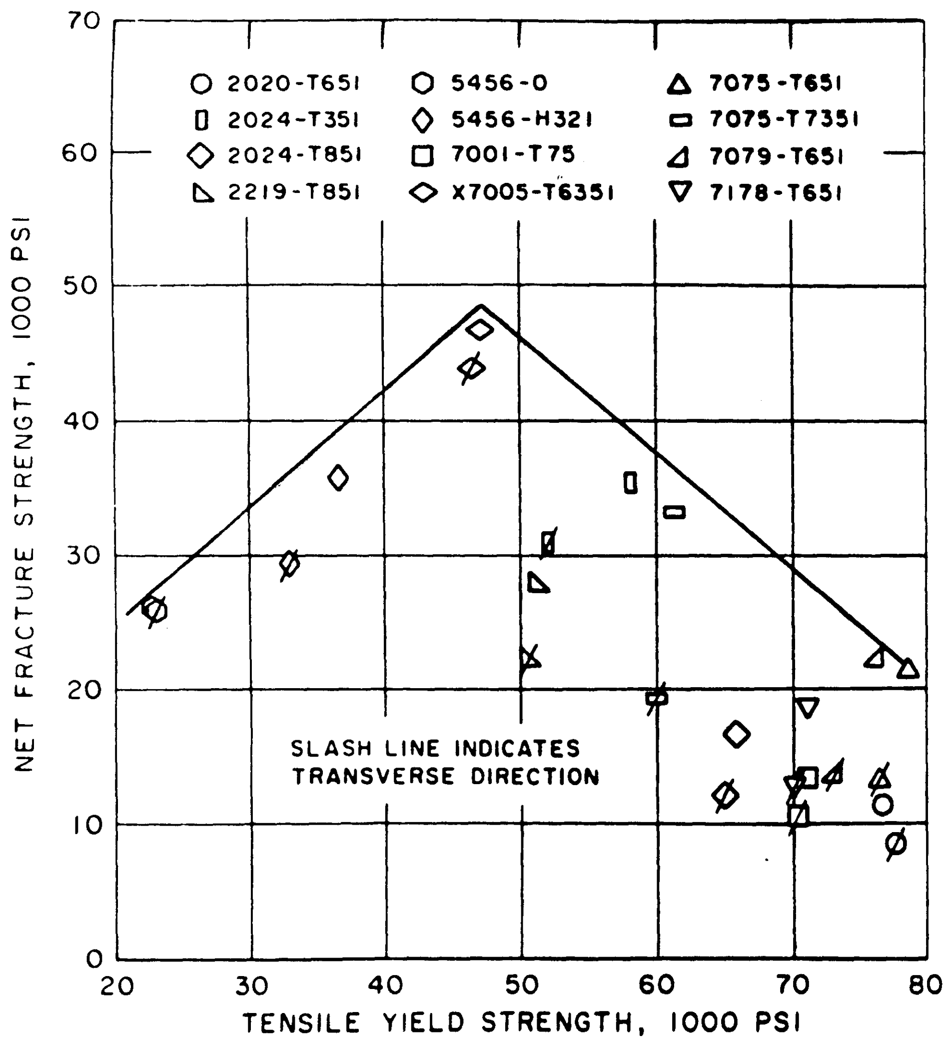


Fig. 15 — Effect of Surface Crack Depth on Calculated K_{Ic} in One-Pass Manual or Automatic TIG Repair Welds

FIGURE 75



70
60
50
40
30
20
10
0
20
30
40
50
60
70
80
2020-T651
2024-T351
2024-T851
2219-T851
5456-0
5456-H321
7001-T75
X7005-T6351
7075-T651
7075-T7351
7079-T651
7178-T651
SLASH LINE INDICATES
TRANSVERSE DIRECTION
NET FRACTURE STRENGTH, 1000 PSI
TENSILE YIELD STRENGTH, 1000 PSI

FIGURE 76

APPENDIX C

List of Tables

Table	Data Source	Page of Data Source
23	28	609
24	28	607
25	28	607
26	3	43
27	53	466
28	55	492
29	57	566
30	60	280
31	8	905
	4	323-S
32	4	323-S
33	4	325-S
34	65	74&75
35	4	321-S
	62	457-S
36	66	471
37	4	326-S
38	67	662-663
39	62	467-S
40	54	646
41	68	96&97
42	69	401-S
43	65	77
44		

Table 2—Description of Notched Tensile Specimens and Results of Tests

Specimen Type	Width or Diameter, in.		Thickness, in.	Notch-Tip Radius, in.	Theoretical Stress-Concentration Factor	Notch-Tensile Strength, psi			
	Gross	Net				1	2	3	Average
Round	0.375	0.353	...	0.0005	10	74 900	75 400	75 400	75 200
				0.0015	6	74 500	74 300	75 200	74 700
				0.004	4.2	76 200	75 200	76 500	76 000
	0.500	0.353	...	0.0007	13	72 500	73 900	73 400	73 300
				0.0015	9.1	76 700	78 500	76 000	77 100
				0.003	6.9	...	77 400	76 200	76 800
				0.010	3.9	84 400	85 600	86 100	85 400
	1.060	0.750	...	0.0005	>20	50 700	50 300	45 400	48 800
				0.0015	13	52 900	49 700	51 600	51 400
				0.003	10	56 400	59 100	59 700	58 400
				0.010	5.5	70 200	74 300	73 200	72 800
				0.0306	3.3	83 000	78 100	84 000	81 700
Sheet; edged notched	0.500	0.250	0.063	0.001	12	63 700	62 500	63 000	63 100
				0.0015	10.2	64 200	67 200	66 200	65 900
				0.005	6.0	68 300	...	70 500	69 400
				0.031	2.8	75 800	73 000	74 700	74 500
	1.000	0.700	0.125	0.001	17.5	48 000	51 500	47 300	48 900
				0.0015	14.4	49 900	50 500	48 400	49 600
				0.003	10.0	56 500	51 200	55 100	54 300
				0.033	3.7	71 800	69 700	70 700	70 700
	3.000	2.000	0.063	0.0005	>30	38 700	43 100	40 600	40 800
				0.005	13.7	44 700	43 600	46 000	44 800
				0.032	5.8	53 300	55 800	59 200	56 100
				0.106	3.5	70 900	68 800	70 300	70 000
	...	...	0.250	0.0005	>30	35 300	34 400	31 900	33 900
				0.005	13.7	35 200	34 700	35 100	35 000
				0.010	10.0	43 900	41 400	44 400	43 200
				0.032	5.8	65 100	57 300	62 300	61 600
				0.106	3.5	72 600	...	65 200	73 900
Sheet; surface notched	0.500	0.031*	0.063	0.0005	4.2	92 800	89 600	92 800	91 700
				0.005	2.5	100 000	96 000	96 100	97 400

* Net thickness.

Table 3—Summary of Results of Tension Tests of Notched Specimens of Various Designs

Specimen		Theoretical Stress-Concentration Factor, K_t			
Type	Dimensions, in.	3–4	5–6	9–10	12–14
Round	0.375 dia	76 000	74 700	75 200	...
	0.500 dia	85 400	76 800	77 100	73 300
	1.060 dia	81 700	72 800	58 400	51 400
Sheet-type; Edge notched	1/2 by 0.063	...	69 400	65 900	63 100
	1 by 0.125	70 700	...	54 300	49 600
	3 by 0.063	70 000	56 100	...	44 800
	3 by 0.250	73 900	61 600	43 200	35 000
Sheet-type; surface notched	1/2 by 0.063	91 700	...	...	...
Range, psi		21 700	20 700	33 900	38 350
Range, % tensile yield strength		+8 to +40	−14 to +18	−34 to +18	−46 to +12

TABLE 1

Notch Beam Test Results					Comparison Data From Other Sources				
Designation	Specimen size in.	Notch/yield strength ratio	Fracture toughness K_{Ic}[1]	Fracture toughness G_{Ic}[2]	Ref. no.	Fracture toughness K_{Ic}[1]	Fracture toughness G_{Ic}[2]	Notch/yield strength ratio	Type and size of specimens
2024-T4	4		47[3]	200[3]	12		300		Center-notched specimens from 1, 2, 4, and 8-in. plates
		0.77	49[4]	217[4]					
	2	1.11	50[4]						
7002-T6	2	1.04	46.1[4]						
	1	1.48	46.6[4]						
7039-T64	3	0.85	48.6[4]						
	1	1.53	50.5[4]						
7075-T6	4	0.47	34.3[4]	107[4]	10	25.8, 27.9		0.40, 0.41	Center-notched sheet, 1/8-in. thick
	2	0.56	30.3[4]						
	1	0.80	30.5[4]		12		115		Center-notched sheet and plate, 1/64 to 1-in. thick
	0.4	1.26	30.4[4]						
					11	44.7		1.03	Circumferentially notched 2-in-dia bar
					11	34.7			Center-notched sheets, 1/8-in. thick
					11	34.2			Edge-notched sheets, 1/8-in. thick
					14	26 to 29			Double-cantilever beams, depth and width varied from 1/2 in. × 1/8 in. to 1 in. × 1/2 in.
					13	28.2, 31.3			Edge-notched sheets, 0.122-in. thick
					13	26.8			Edge-notched sheets, 0.061-in. thick
					13	33.5, 35.1			Center-notched sheets, 0.122-in. thick
7079-T6	3	0.72, 0.71	44.2[3], 43.5[3]		11	41.9		0.97	Circumferentially notched 2-in-dia round bar
	2	0.87, 0.81	43.5[3], 40.1[3]		11	33.0			Center-notched sheet, 1/8-in. thick
	0.8	1.33, 1.31	42.2[3], 41.8[3]		11	36.2			Edge-notched sheet, 1/8-in. thick
7106-T63	3	0.64	34.0[4]						
	1	1.02	32.0[4]						
	0.75	1.35	36.0[4]						
	0.4	1.78	34.8[4]						

[1] K_{Ic}: ksi$\sqrt{\text{in.}}$
[2] G_{Ic}: in-lb/in.^{2}
[3] Calculated by Lubahn's formula [6].
[4] Calculated by Srawley and Brown's formula [7, 17].
Ref. no. = Reference number.

TABLE 2

Table 3. Summary of values of plane-strain stress-intensity factor, K_Q, determined in various types of test

Alloy and temper	Sample number	Longitudinal: Initial deviation CN	Longitudinal: Initial deviation Plain SEN	Longitudinal: Initial deviation Face-grooved SEN	Longitudinal: Initial deviation NB	Longitudinal: 5% Secant offset CN	Longitudinal: 5% Secant offset Plain SEN	Longitudinal: 5% Secant offset Face-grooved SEN	Longitudinal: 5% Secant offset NB	Transverse: Initial deviation CN	Transverse: Initial deviation Plain SEN	Transverse: Initial deviation Face-grooved SEN	Transverse: Initial deviation NB	Transverse: 5% Secant offset CN	Transverse: 5% Secant offset Plain SEN	Transverse: 5% Secant offset Face-grooved SEN	Transverse: 5% Secant offset NB
2020-T651	199002	21,400	23,100	—	21,000	22,400	23,100	—	22,000	19,200	17,900	—	20,800	19,200	18,200	—	20,800
	199012	20,800	21,800	—	22,500	22,000	22,000	—	24,100	18,500	18,100	—	16,500	18,500	18,400	—	17,400
	198992	21,800	21,200	—	20,800	22,500	21,600	—	20,800	19,500	17,100	—	19,300	19,500	18,500	—	19,300
2024-T351	317555	50,200	31,600	32,700	—	56,800	42,800	43,200	—	43,400	28,000	32,300	—	50,300	37,900	35,200	—
2024-T851	189791	25,100	23,500	21,800	19,800	27,600	24,400	24,100	22,500	21,300	19,200	19,600	18,100	23,700	21,600	19,800	18,800
	189801	27,200	24,600	23,600	21,100	29,200	26,200	24,500	23,100	23,000	20,600	—	19,500	23,800	21,300	—	18,800
	189811	27,000	22,800	22,600	25,500	28,900	24,600	23,100	25,500	22,300	21,100	19,300	22,600	24,000	22,200	20,200	22,600
2219-T851	199761	42,700	27,900	32,600	27,000	50,800	37,700	39,400	36,000†	40,000	28,200	32,000	25,300	44,200	32,700	37,800	33,600†
	199971	44,300	28,700	26,300	—	52,400	42,100	38,000	—	33,100	26,600	25,000	—	47,100	38,800	34,200	—
	199981	39,300	31,100	29,200	—	45,800	42,000	41,500	—	38,200	23,800	28,800	—	41,400	39,200	35,200	—
7001-T75		24,900	22,900	21,700	23,800	25,800	25,000	23,900	24,200	21,600	20,100	—	21,800	22,200	20,600	—	21,800
	183781	23,700	24,100	22,300	25,400	27,900	25,800	25,000	25,400	22,900	21,100	—	23,700	23,900	22,500	—	23,700
	183791	24,600	20,800	—	20,700	26,600	24,600	—	22,500	23,200	18,800	—	19,600	23,200	20,700	—	20,400
X7005-T6351	317623	52,200	31,500	—	‡	72,400	48,500	—	34,500†	49,500	29,400	—	‡	66,900	43,400	—	35,000†
7075-T651	183182	32,300	30,000	—	26,400	34,500	30,200	—	28,100	27,600	24,600	—	24,000	27,600	24,900	—	24,600
	183202	31,500	26,200	—	23,000	31,500	26,200	—	24,500	27,100	23,000	—	19,500	27,100	23,000	—	20,600
	195552	31,600	29,000	—	-	32,700	29,000	—	—	28,400	23,200	—	‡	28,400	24,000	—	—
7075-T7351	183184	39,800	31,500	28,800	31,700	45,400	36,600	34,000	32,900	33,300	28,200	26,700	27,600	35,400	31,300	30,300	28,100
	183204	36,700	28,200	28,600	—	43,300	34,600	33,100	—	35,000	26,100	26,000	—	35,700	31,800	24,400	—
	195554	37,000	28,500	29,400	30,600	40,800	34,100	32,800	31,000	31,800	26,400	24,300	26,600	33,200	28,200	25,600	26,900
7079-T651	183331	32,400	27,300	—	30,400	33,200	30,300	—	31,000	27,400	23,300	—	—	27,400	24,200	—	—
	183332	30,500	27,000	—	26,200	30,500	28,800	—	28,700	27,100	23,600	—	23,500	27,100	25,700	—	24,900
	183333	33,400	29,000	—	26,100	43,100	29,800	—	31,000	27,800	24,300	—	24,200	27,800	25,300	—	25,300
7178-T7651	317624	30,100	29,000	27,700	‡	31,200	29,000	27,700	30,900	24,400	23,200	22,600	24,100	26,400	24,000	22,600	28,300

† Not valid; specimen too thin and plastic deformation too great.
‡ Not determined, material too tough; no clear initial cracking.
Note: Underlined values may be considered to be valid K_{Ic} values per[4]
CN – Center-notch;
SEN – Single-edge-notch;
B – Notch-bend.

TABLE 3

Material temper, form	Thickness, in.	Chemistry, wt % Cr	Cu	Fe	Mg	Mn	Si	Ti	Zn
2014-T6 Sheet	0.063	—	4.37	0.36	0.23	0.62	0.69	0.018	—
2024-T3 Sheet	0.025	0.01	4.80	0.32	1.46	0.55	0.19	0.02	0.21
2024-T4 Sheet	0.032	—	4.35	0.31	1.34	0.51	0.15	0.015	—
2024-T4 Plate	2.0	0.01	4.83	0.53	1.84	0.92	0.26	0.02	0.25
2219-T81 Sheet	0.063	—	5.8	0.10	0.01	0.29	0.1	0.066	—
2219-T87 Sheet	0.063	—	6.0	0.15	0.01	0.37	0.1	0.068	—
6061-T4 Sheet	0.025	0.19	0.15	0.50	0.95	0.09	0.56	0.03	0.17
6061-T6 Sheet	0.020	0.18	0.23	0.37	0.94	0.08	0.76	0.06	0.05
6061-T6 Plate	1.0	0.37	0.28	0.66	1.23	0.08	0.83	0.02	0.10

TABLE 4

Table I. History and Chemical Analyses of Aluminum Alloys

Material temper form	Gage, in.	Specification or heat no.	Supplier	Chemistry Cr	Cu	Fe	Mg	Mn	Si	Ti	Zn
5052-H38 (sheet)	0.040	635-521	Alcoa	0.172	0.045	0.249	2.59	-	0.11	0.011	-
5083-H38 (sheet)	0.050	Experimental	Kaiser	0.12	0.04	0.14	4.50	0.80	0.11	0.02	0.04
5086-H34 (sheet)	0.040	106-404	Alcoa	0.11	0.05	0.18	3.66	0.37	0.11	0.01	-
5086-H38 (sheet)	0.050	Experimental	Kaiser	0.15	0.04	0.27	3.93	0.45	0.12	0.02	0.06
5154-H38 (sheet)	0.040	667-471	Alcoa	0.21	0.03	0.22	3.38	-	0.12	0.10	-
5456-H343 (sheet)	0.050	Mil-A-19842	Alcoa	0.08	0.06	0.22	5.27	0.81	0.13	0.02	0.03

TABLE 5

Table I. Chemical Analyses and History of Materials

Material	Temper	Form	Thickness, in.	Hardness 15-N superficial	Composition							
					Cr	Cu	Fe	Mn	Si	Ti	Zn	Mg
7075	T6	Sheet	0.025	63	0.23	1.47	0.32	0.04	0.15	.05	5.85	2.45
7079	T6	Sheet	0.080	62	0.14	0.67	-	0.18<	0.1	.056	3.70	4.26
7178	T6	Sheet	0.036	66	0.26	1.95	0.21	0.04	0.11	.02	6.6	3.0
X7275	T6	Sheet	0.050	65	0.16	1.40<	0.1	0.04<	0.1	.035	5.95	2.94
7075	T6	Plate	2.5	62	0.1[illegible]	1.43	0.25	0.04	0.14	.039	5.46	2.75
7079	T6	Billet	5.0	61	0.13	0.71	-	0.14<	0.1	.060	3.30	4.15

TABLE 6

Table 1. Temper Designations for Strain-Hardened Alloys

Temper	Description
F	**As-fabricated.** No control over the amount of strain hardening; no mechanical property limits.
O	**Annealed, recrystallized.** Temper with the lowest strength and greatest ductility.
H1	**Strain hardened.** H12, H14, H16, H18. The degree of strain hardening is indicated by the second digit and varies from quarter-hard (H12) to full-hard (H18), which is produced with approximately 75% reduction in area. H19. An extra-hard temper for products with substantially higher strengths and greater strain hardening than obtained with the H18 temper.
H2	**Strain hardened and partially annealed.** H22, H24, H26, H28. Tempers ranging from quarter-hard to full-hard obtained by partial annealing of cold worked materials with strengths initially greater than desired.
H3	**Strain hardened and stabilized.** H32, H34, H36, H38. Tempers for age-softening aluminum-magnesium alloys that are strain hardened and then heated at a low temperature to increase ductility and stabilize mechanical properties.
H112	**Strain hardened during fabrication.** No special control over amount of strain hardening but requires mechanical testing and meets minimum mechanical properties.
H321	**Strain hardened during fabrication.** Amount of strain hardening controlled during hot and cold working.
H323, H343	Special strain hardened, corrosion-resistant tempers for aluminum-magnesium alloys.

TABLE 7

The basic temper designations are as follows:

F **As fabricated.** Applies to wrought products that acquire some temper from shaping processes in which no special control is exercised over the amount of strain hardening or thermal treatment. For wrought products in this temper, there are no mechanical-property limits. Applies to castings in the as-cast condition if the alloy is also regularly produced in heat treated tempers.

O **Annealed (wrought products only).** Applies to the softest temper of wrought products.

W **Solution heat treated.** An unstable temper applicable only to alloys that age at room temperature after solution heat treatment. This designation is specific only when the period of natural aging is indicated; for example: W(0.5 hr).

T **Heat treated to produce stable tempers other than F or O.** Applies to wrought and cast products that are heat treated, with or without supplementary cold working to produce stable tempers.

The T is always followed by one or more digits. Numerals 1 through 10 indicate specific sequences of treatments:

T1 **Naturally aged to a substantially stable condition.** Applies to products in which partial solution of alloying elements is provided by elevated-temperature, rapid-cool fabrication.

T2 **Annealed (cast products only).** Designates a temper produced by a type of annealing treatment used to improve ductility and increase dimensional stability of castings.

T3 **Solution heat treated, cold worked, and naturally aged to a substantially stable condition.** Applies to products that are cold worked to improve strength, or in which the effect of cold work associated with flattening or straightening is recognized in applicable specifications. Different amounts of cold work are denoted by a second digit.

T4 **Solution heat treated and naturally aged to a substantially stable condition.** Applies to products that are not cold worked after solution heat treatment, or in which the effect of cold work associated with flattening or straightening may not be recognized in applicable specifications.

T5 **Artificially aged only.** Applies to products that are artificially aged after an elevated-temperature, rapid-cool fabrication process, such as casting or extrusion, to improve strength and /or dimensional stability.

T6 **Solution heat treated and artificially aged.** Applies to products not cold worked after solution heat treatment, or in which the effect of cold work associated with flattening or straightening may not be recognized in applicable specifications.

T7 **Solution heat treated and overaged.** Applies to products that are solution heat treated and artificially aged beyond the condition of maximum strength, to provide controlled special characteristics, such as dimensional stability, lower residual stresses, or improved resistance to corrosion.

TABLE 8

T8	**Solution heat treated, cold worked, and artificially aged.** Applies to products that are cold worked to improve strength, or in which the effect of cold work associated with flattening or straightening is recognized in applicable specifications. Different amounts of cold work are denoted by a second digit.
T9	**Solution heat treated, artificially aged, and cold worked.** Applies to products that are cold worked as a final operation, to improve strength.
T10	**Artificially aged and cold worked.** Applies to products that are artificially aged after an elevated-temperature, rapid-cool fabrication process, such as casting or extrusion, and then cold worked to improve strength.

A period of natural aging may occur between the operations listed for tempers T3 through T10. Control of this period may be necessary to achieve the desired characteristics.

The following designations involving additional digits are assigned to stress-relieved tempers of wrought products:

Tx51(a)	**Stress relieved by stretching.** Applies to products that are stress relieved by stretching the following amounts after solution heat treatment: plate — 0.5 to 3% permanent set; rod, bar and shapes — 1 to 3% permanent set. This designation applies directly to plate and rolled or cold finished rod and bar. These products receive no further straightening after stretching. Additional digits are used in the designations for extruded rod, bar, shapes and tube as follows: T*x*510(a) applies to products that receive no further straightening after stretching; T*x*511(a) applies to products that receive minor straightening after stretching to comply with standard straightness tolerances.
Tx52(a)	**Stress relieved by compressing.** Applies to products that are stress relieved by compressing after solution heat treatment, to produce a nominal permanent set of 2.5%.
Tx53(a)	**Stress relieved by thermal treatment.**

The following temper designations are assigned to some wrought products that are heat treated by the user:

T42	**Solution heat treated** (b).
T62	**Solution heat treated and artificially aged**(b).

(a) The letter *x* represents digits 3, 4, 6, or 8, whichever is applicable. (b) Exceptions not conforming to these definitions are 4032-T62 and 6101-T62.

TABLE 8 CONT.

2219-T87	Plate	1	252314	L	RT	67,400	56,200	11.8	28	82,300	1.22	1.46
					−112	72,000	59,900	12.0	28	82,800	1.15	1.38
					−320	83,500	67,000	14.0	28	91,500	1.09	1.37
					−423	98,600	72,400	15.2	21	102,500	1.04	1.42
					−452	97,500	73,200	15.0	22	100,200	1.03	1.37
						98,100	75,100	15.5	24	—	—	—
2618-T651	Plate	1¾	317406	L	RT	62,400	57,600	10.8	32	81,200	1.30	1.41
					−18	65,900	60,500	10.0	27	—	—	—
					−112	68,200	62,500	10.7	27	87,100	1.28	1.39
					−320	78,000	68,700	13.3	26	92,000	1.18	1.34
					−452	87,600	72,300	15.0	23	98,700	1.13	1.37
3003-H14	Rod	¾	252570	L	RT	22,900	21,100	16.8	68	—	—	—
					−18	24,000	21,900	15.0	58	—	—	—
					−112	25,300	22,300	18.5	59	—	—	—
					−320	36,600	25,900	32.5	56	—	—	—
					−452	58,100	30,100	32.0	49	65,100	1.12	2.16
5083-O	Plate	1	Ref. 6, 8	L	RT▲	46,800	20,400	19.5	26	54,000	1.16	2.65
					−320▲	63,000	23,000	34.0	34	61,000	0.97	2.65
					−423	85,200	25,200	32.0	24	59,300	0.70	2.36
					−452‡‡	80,800	25,800	32.0	33	62,300	0.77	2.42
5083-H321	Plate	1	Ref. 6, 8	L	RT▲	48,600	34,100	15.0	23	61,100	1.26	1.80
					−320▲	66,100	39,700	31.5	33	70,400	1.06	1.77
					−423	90,000	41,800	30.0	24	72,800	0.81	1.74
					−452‡‡	85,800	40,500	29.0	33	73,700	0.86	1.82
5454-O	Plate	1	197127	L	RT	35,800	16,700	24.5	48	48,100	1.34	2.88
					−18	37,200	16,600	26.5	57	—	—	—
					−112	38,200	16,800	29.5	57	—	—	—
					−320	54,300	19,400	39.5	49	60,200	1.11	3.04
					−452	73,900	24,100	34.3	35	65,600	0.89	2.72
5454-H32	Plate	1	197133	L	RT	40,900	28,900	15.7	32	56,200	1.37	1.94
					−18	42,200	28,800	19.5	44	—	—	—
					−112	43,700	29,200	23.0	48	—	—	—
					−320	61,100	34,500	32.0	40	69,200	1.13	2.00
					−452	82,300	39,400	28.6	31	77,700	0.94	1.97

Table I—continued overleaf

TABLE 9

Table V. Average Results of Tear Tests of Some Aluminum-Magnesium Alloys at Room Temperature -320°F

Alloy and temper	Thickness, in.	Sample number	Direction	Room temperature				-320°F				Change from room temperature value			
				Maximum load, lb	Energy required to: Initiate a crack, in.-lb	Energy required to: Propagate a crack, in.-lb	Total energy, in.-lb	Maximum load, lb	Energy required to: Initiate a crack, in.-lb	Energy required to: Propagate a crack, in.-lb	Total energy, in.-lb	Maximum load, lb	Energy required to: Initiate a crack, in.-lb	Energy required to: Propagate a crack, in.-lb	Total energy, in.-lb
5154-0	0.750	171710	L	1230	56	138	194	1520	106	202	308	+24	+89	+47	+59
			T	1240	65	123	188	1500	90	186	276	+21	+38	+51	+41
5086-0	0.375	170115	L	1090	73	139	212	1310	112	183	295	+20	+53	+33	+39
			T	1080	68	120	197	1250	97	157	252	+16	+43	+22	+28
	0.500	170116	L	1115	69	134	203	1345	101	175	276	+21	+46	+31	+36
			T	1115	73	117	190	1350	106	160	266	+21	+45	+37	+40
5454-0	0.375	170155	L	1300	70	107	177	1600	96	190	286	+23	+37	+78	+62
			T	1320	76	107	183	1615	108	158	266	+22	+42	+47	+45
	0.500	170156	L	1345	78	120	198	1660	110	199	309	+24	+41	+66	+56
			T	1375	79	120	199	1660	108	171	279	+21	+37	+43	+40
5454-H34	0.375	170165	L	1425	72	111	183	1765	103	172	275	+24	+43	+55	+50
			T	1435	72	120	192	1700	94	154	249	+18	+31	+28	+30
	0.500	174303	L	1540	43	92	135	1955	81	149	220	+27	+88	+62	+71
			T	1560	48	65	113	1875	73	127	200	+20	+52	+94	+77
5356-0	0.750	167070	L	1245	66	148	214	1370	84	190	274	+10	+27	+29	+28
			T	1220	62	120	182	1390	78	166	244	+14	+27	+39	+34
5356-H321	0.750	167059	L	1700	66	105	171	2055	86	170	256	+21	+31	+62	+50
			T	1605	46	66	110	1880	52	98	150	+17	+14	+48	+36
5083-0	0.375	167246	L	1350	54	112	166	1565	82	148	230	+16	+58	+16	+38
			T	1350	56	96	152	1540	62	132	194	+14	+11	+37	+28
5083-H113	0.375	167245	L	1650	45	112	157	1900	59	133	192	+15	+30	+18	+22
			T	1580	43	84	127	1825	57	108	165	+15	+31	+29	+30
	0.375	170092	L	1585	42	85	127	1850	60	127	197	+17	+43	+49	+55
			T	1540	43	81	124	1795	57	97	154	+17	+33	+20	+24
	0.500	170093	L	1610	38	82	120	1865	57	112	169	+16	+50	+36	+41
			T	1575	38	74	112	1775	36	82	118	+13	-5	+11	+6
5456-0	0.375	170099	L	1465	57	99	159	1655	56	127	183	+13	-2	+24	+15
			T	1450	50	91	141	1650	60	110	170	+33	+20	+21	+21
	0.750	167056	L	1290	44	98	142	1450	52	121	173	+12	+18	+23	+22
			T	1230	42	82	124	1390	49	104	153	+13	+18	+27	+23
5456-H321	0.375	170104	L	1515	42	92	134	1780	52	110	162	+18	+24	+19	+21
			T	1525	46	79	125	1715	49	91	140	+12	+7	+15	+12
	0.500	170105	L	1710	54	104	158	1965	54	128	182	+15	0	+23	+15
			T	1650	48	83	131	1918	52	92	144	+16	+8	+11	+11
	0.750	167059	L	1720	46	75	121	1990	62	114	176	+16	+35	+51	+45
			T	1550	35	62	97	1810	42	66	108	+17	+19	+6	+11

TABLE 9 CONT.

RESULTS OF TEAR TESTS OF SOME ALUMINUM-MAGNESIUM ALLOYS AT ROOM TEMPERATURE AND −320 F

Alloy and Temper	Form	Product Tested		Specimen			Room Temperature				−320°F				Change From Room Temperature Values, %			
								Energy Required to				Energy Required to				Energy Required to		
		Dimensions, in.	Condition	Thickness	Notch	Direction (Longitudinal-Transverse)	Maximum Load, lb	Initiate a crack, in-lb	Propagate a crack, in-lb	Total Energy, in-lb	Maximum Load, lb	Initiate a crack, in-lb	Propagate a crack, in-lb	Total Energy, in-lb	Maximum Load, lb	Initiate a crack, in-lb	Propagate a crack, in-lb	Total Energy, in-lb
5154-H112	Plate	⅜	Plain	0.10	Vee	L	1,335	50	97	147	1,740	93	198	291	+24	+86	+104	+98
						T	1,450	54	96	150	1,755	90	188	278	+21	+67	+98	+85
		⅜	Plain	0.10	Vee	L	1,230	56	138	194	1,520	106	202	308	+24	+89	+46	+59
						T	1,240	65	123	188	1,500	90	186	276	+21	+38	+51	+47
5356-0	Plate	⅜	Plain	0.10	Vee	L	1,245	66	148	214	1,370	84	190	274	+10	+27	+28	+28
						T	1,220	62	120	182	1,390	78	166	244	+14	+26	+38	+34
-H321	Plate	⅜	Plain	0.10	Vee	L	1,700	66	105	171	2,055	86	170	256	+21	+30	+62	+50
						T	1,605	46	66	112	1,880	52	98	150	+17	+13	+48	+34
5083-0	Plate	⅜	Plain	0.10	Vee	L	1,350	54	[illegible]	166	1,565	82	148	230	+16	+52	+32	+39
						T	1,350	56		152	1,540	62	132	194	+14	+11	+38	+28
-H113	Plate	⅜	Plain	0.10	Vee	L	1,650	45		157	1,900	58	133	191	+15	+29	+19	+22
						T	1,580	44	[illegible]	128	1,825	57	108	165	+15	+30	+28	+29
5456-0	Plate	⅜	Plain	0.10	Vee	L	1,290	44	98	142	1,450	52	121	173	+12	+18	+23	+22
						T	1,230	42	82	[illegible]24	1,390	49	104	153	+13	+17	+27	+23
-H321	Plate	⅜	Plain	0.10	Vee	L	1,720	46	75	[illegible]	1,990	63	114	177	+16	+37	+52	+46
						T	1,550	35	62		1,810	42	66	108	+17	+20	+6	+11

TABLE 10

Table 2 - Tear Resistance of 2020-T6 Sheet + at Room Temperature After Elevated Temperature Exposure

Exposure							Energy Required to			
Temp. °F	Time, hr	Tensile Strength, psi	Yield Strength psi	Elong in 2 in.	Tear Strength, psi	Tear Strength / Yield Strength	Initiate a Crack, in.-lb	Propagate a Crack, in.-lb	Total Energy, in.-lb	Unit Propagation Energy, in.-lb/in.^{2}
None		81,000	75,800	7.5	41,600	0.55	3	++	3	++
400	1/2	77,700	71,600	7.5	53,500	0.75	5	++	5	++
400	6	73,000	65,000	6.5	56,100	0.86	4	2	6	35
400	24	67,000	[illegible]	9.0	57,500	1.02	8	11	19	180
500	1/2	60,[illegible]	47,200	9.5	59,300	1.26	7	13	20	200
500	6	50,700	34,900	10.0	56,400	1.62	9	21	30	330
600	1/2	46,300	28,900	11.0	54,100	1.87	12	27	39	430

\+ Transverse direction

++ Too low To measure

TABLE 11

Table III. Mechanical Properties of 2024-T3 Aluminum Alloy*

Test temperature, °F	Direction	Fty, ksi	Ftu, ksi	Elongation, %	Proportional limit, ksi	Elastic modulus, psi × 10^6	Notched T.S. (K_t 6.3), psi	Notched/unnotched tensile ratio
78	Longitudinal	47.4	67.9	18	37.5	10.2	60.2	0.89
78	Transverse	43.9	65.8	18	31.8	10.3	62.9	0.96
−100	Longitudinal	48.9	70.2	21	39.8	10.5	61.2	0.87
−100	Transverse	44.5	67.8	21	35.4	10.6	62.8	0.93
−320	Longitudinal	60.9	87.0	22	48.4	10.9	76.2	0.88
−320	Transverse	56.1	83.4	22	43.2	11.0	74.7	0.90
−423	Longitudinal	73.1	110	17	67.5	11.4	88.8	0.81
−423	Transverse	69.0	107	18	58.0	11.5	86.8	0.81

* 0.025-in. sheet, Aluminum Company of America, QQ-A-355.

TABLE 12

Table IV. Mechanical Properties of 2024-T4 Aluminum Alloy*

Test temperature, °F	Direction	Fty, ksi	Ftu, ksi	Elongation, %	Proportional limit, ksi	Elastic modulus, psi × 10^6	Notched T.S. (K_t = 6.3), psi	Notched/unnotched tensile ratio
78	Longitudinal	42.8	67.7	19	34.1	10.7	59.0	0.87
78	Transverse	41.5	67.1	20	32.7	10.4	57.5	0.86
−100	Longitudinal	43.7	69.8	22	36.3	10.7	60.7	0.87
−100	Transverse	42.7	68.0	24	35.4	10.7	58.9	0.87
−320	Longitudinal	54.1	84.9	27	44.3	11.2	71.9	0.85
−320	Transverse	53.6	81.8	19	43.4	11.0	68.2	0.83
−423	Longitudinal	73.3	107	16	59.9	11.6	88.3	0.83
−423	Transverse	67.5	97.1	10	54.1	11.6	85.4	0.88

* 0.032-in. sheet, Aluminum Company of America, QQ-A-355.

TABLE 13

Table VI. Mechanical Properties of 2219-T81 Aluminum Alloy*

Test temperature, °F	Direction	Fty, ksi	Ftu, ksi	Elongation, %	Proportional limit, ksi	Elastic modulus, psi × 10^6	Notch T.S. (K_t = 6.3), ksi	Notched/unnotched tensile ratio	Weld† T.S., ksi	Weld elongation, %	Joint efficiency, %
78	Longitudinal	52.0	67.5	10	36.7	10.4	64.4	0.95	48.1	3	71
78	Transverse	51.0	67.2	10	34.0	10.4	66.0	0.98	—	—	—
−100	Longitudinal	56.7	73.3	9	39.0	10.6	68.6	0.94	49.7	5	68
−100	Transverse	54.7	72.3	10	38.4	10.4	67.0	0.93	—	—	—
−320	Longitudinal	62.2	85.2	11	47.3	10.9	77.4	0.91	64.3	3	75
−320	Transverse	61.4	84.6	12	47.8	10.8	76.1	0.90	—	—	—
−423	Longitudinal	70.6	102	15	56.4	11.5	93.4	0.92	72.4	2	71
−423	Transverse	67.5	102	15	57.3	11.3	91.3	0.90	—	—	—

* 0.063-in. sheet, Aluminum Company of America.
† Manually welded with 2319 aluminum filler metal, no post heat treatment, tested with bead in place.

TABLE 14

Table VII. Mechanical Properties of 2219-T87 Aluminum Alloy*

Test temperature, °F	Direction	Fty, ksi	Ftu, ksi	Elongation, %	Proportional limit, ksi	Elastic modulus, psi × 10^6	Notch T.S. (K_t = 6.3), ksi	Notched/unnotched tensile ratio	Weld† T.S., ksi	Weld elongation, %	Joint efficiency, %
78	Longitudinal	58.2	70.7	9	42.5	10.5	69.7	0.99	51.1	2	72
78	Transverse	58.6	71.1	9	41.6	10.4	69.7	0.98	—	—	—
−100	Longitudinal	62.4	76.4	9	43.7	10.8	74.6	0.98	49.5	4	65
−100	Transverse	62.8	76.4	9	—	—	73.7	0.96	—	—	—
−320	Longitudinal	69.7	88.4	11	46.2	10.9	85.5	0.97	61.2	2	69
−320	Transverse	70.4	89.4	11	45.2	10.8	83.7	0.94	—	—	—
−423	Longitudinal	76.4	104	14	62.9	11.4	95.6	0.92	73.0	1	70
−423	Transverse	75.9	105	14	62.9	11.4	95.2	0.91	—	—	—

* 0.063-in. sheet, Aluminum Company of America.
† Manually welded with 2319 aluminum filler metal, no post heat treatment, tested with bead in place.

TABLE 15

Orientation and Spec. No.	Thickness, in. B	Thickness, in. B_N	Crack Length / Width	K_Q,[a] ksi $\sqrt{\text{in.}}$	σ_{ys}, ksi	$\frac{K_Q}{\sigma_{ys}^2}$	$\frac{B_N}{(K_Q/\sigma_{ys})^2}$[b]	Plastic Zone Radius, $r = 1/2\pi\,(K_Q/\sigma_{ys})^2$	B_N/r	Growth Rate: Applied K_I, ksi $\sqrt{\text{in.}}$	Growth Rate: da/dN μ in. per cycle
Standard Rolled Plate											
RW-1	0.308	0.241	0.525	37.1	70.5	0.265	0.90*	0.042	5.7	7.35	2.4
RW-2	0.310	0.240	0.525	34.9	70.5	0.235	1.02	0.037	6.4	11.0	8.9
RW-3	0.309	0.243	0.520	38.0	70.5	0.278	0.69	0.044	4.4	14.7	15.7
TW-1	0.311	0.238	0.520	27.4	64.7	0.178	1.33	0.028	8.4	7.35	1.82
TW-2	0.312	0.239	0.525	25.7	64.7	0.156	1.53	0.024	9.6	11.0	8.8
TW-3	0.311	0.239	0.525	24.3	64.7	0.140	1.71	0.022	10.7	14.7	15.6
TR-1	0.311	0.245	0.511	18.2*	64.7	0.078	3.14*	0.012	19.7	7.35	2.3
TR-2	0.311	0.245	0.511	15.8*	64.7	0.059	4.16*	0.009	26.0	11.0	10.1
TR-3	0.310	0.245	0.511	17.5*	64.7	0.072	3.41*	0.011	21.2	14.7	26.4
Preforged and Rolled Plate											
RW-1	Test specimens fatigued out of plane of notch.									7.35	1.1[c]
RW-2										11.0	7.8[c]
RW-3	0.307	0.247	0.511	37.4	74.6	0.244	0.50	0.038	6.3	14.7	15.1
TW-1	0.307	0.245	0.503	18.5	67.8	0.075	3.27	0.012	20.4	7.35	1.86[d]
TW-2	0.308	0.243	0.483	23.1	67.8	0.116	2.10	0.018	13.2	11.0	7.5[d]
TW-3	0.307	0.245	0.483	21.5	67.8	0.122	2.42	0.016	15.2	14.7	14.25[d]
TR-1	0.307	0.240	0.511	20.1*	67.8	0.088	2.73*	0.014	17.0	7.35	2.3
TR-2	0.310	0.245	0.511	20.0*	67.8	0.092	2.67*	0.014	16.6	11.0	8.9
TR-3	0.309	0.245	0.520	19.9*	67.8	0.086	2.85*	0.013	17.9	14.7	20.4

[a] Candidate values of K_{Ic} Based on 2 pct crack extension (ASTM-E24 recommended practice) *considered valid K_{Ic} value if requirements in (b) are met.

[b] These values should be greater than 2.5 for a valid K_{Ic} value (ASTM-E24 recommended practice)

[c] Measured over only first 0.3 in. of growth

[d] Average of 2 tests.

TABLE 16

Table I. Nominal Compositions of Casting Alloys Tested at Subzero Temperatures

Alloy	Nominal composition, element, % Si	Cu	Mg	Ni	Zn	Other	Tempers Tested* Sand cast	Permanent-mold cast	Premium-strength cast†
108	3.0	4.0	—	—	—	—	F	—	—
A140	—	8.0	6.0	0.5	—	0.50 Mn	F	—	—
142	—	4.0	1.5	2.0	—	—	T77	—	—
195	0.8	4.5	—	—	—	—	T6	—	—
B218	—	—	7.0	—	—	—	F	—	—
220	—	—	10.0	—	—	—	T4	—	—
X335	8.0	0.5	0.50	—	—	1.0 Pb, 1.0 Bi	T6	T61	—
A344	7.0	—	—	—	—	—	—	F	—
354	9.0	1.8	0.50	—	—	—	—	T62	—
C355	5.0	1.3	0.50	—	—	—	—	—	T61
356	7.0	—	0.30	—	—	—	T4, T6, T7, T71	T6, T7	—
A356	7.0	—	0.30	—	—	—	T7	T61, T62, T7	T61
A357	7.0	—	0.50	—	—	0.05 Be	—	—	T61, T62
359	9.0	—	0.60	—	—	—	—	T62	—
A612	—	0.50	0.7	—	6.5	—	F	—	—

* For definition of tempers, refer to ASTM Specifications B26-65 and B108-65, ASTM Standards, Part 6 (October 1965).

† The term "premium-strength cast" is used to identify castings produced by controlled foundry practices which provide higher properties than are usually obtained by conventional practices.

Note: Fe restricted to 0.2% maximum in alloys A344, 354, C355, A356, A357 and 359.

TABLE 17

Table II. Results of Tensile Tests of Smooth and Notched Specimens from Aluminum Alloy Sand Castings at Room and Subzero Temperatures

Alloy and temper	Temperature, °F	Tensile strength, psi	Yield strength, psi	Elongation in 4D, %	Reduction of area, %	Notch-tensile strength, psi	NTS/TS	NTS/YS
108-F	RT	25,000	18,500	1.8	2	25,200	1.01	1.36
	−112	26,200	21,700	2.0*	0*	21,600	0.82	1.00
	−320	30,650	30,250	1.3	0	21,700	0.71	0.72
A140-F	RT	33,800	26,000	1.4	2	29,900	0.88	1.15
	−112	32,800†	22,700	†	†	19,400	< 0.58	0.75
	−320	36,800†	32,400	†	†	17,400	< 0.47	0.54
142-T77	RT	29,800	20,400	2.1	4	29,000	0.97	1.42
	−112	32,800†	22,700	†	†	26,600	< 0.81	1.17
	−320	32,800†	26,800	†	†	27,700	< 0.84	1.03
195-T6	RT	42,000	27,100	6.4	10	43,500	1.04	1.60
	−112	45,500	32,000	6.0	5	58,000	1.08	1.45
	−320	53,700	39,900	5.0	5	58,000	1.08	1.45
B218-F	RT	41,200	21,200	12.9	13	43,800	1.06	2.06
	−112	41,600	22,300	10.0	11	42,400	1.06	1.90
	−320	37,300	25,500	3.7	5	35,500	0.95	1.39
	−423	30,800	28,100	0.8	1	20,000	0.65	0.70
220-T4	RT	34,200	31,600	2.1	2	38,400	1.12	1.22
	−112	41,600	37,400	1.3	1	27,900	0.79	0.88
	−320	39,600	39,600‡	0.7	0	27,200	0.69	0.69
X335-T6	RT	37,300	23,400	8.6	12	38,200	1.02	1.63
	−112	42,300	27,400	8.0	10	43,000	1.02	1.57
	−320	51,600	32,100	7.6	10	45,600	0.88	1.42
356-T4	RT	31,100	19,800	4.4	6	31,600	1.02	1.60
	−112	36,600	23,400	4.4	6	37,600	1.04	1.61
	−320	40,800	27,200	2.7	3	42,200	1.04	1.55
356-T6	RT	38,600	32,600	2.2	3	37,400	0.97	1.15
	−112	43,100	35,800	2.7	2	40,000	0.93	1.12
	−320	47,500	39,200	2.7	2	44,000	0.93	1.13
356-T7	RT	37,800	33,700	1.6	2	34,500	0.91	1.02
	−112	41,400	34,400	2.0	2	38,800	0.94	1.13
	−320	45,100	38,800	1.3*	0	43,100	0.96	1.11
356-T71	RT	28,800	20,200	5.0	–	32,000	1.11	1.59
	−112	32,200	22,200	4.4	5	29,600	0.92	1.34
	−320	37,400	25,300	3.0	2	34,400	0.92	1.36
A356-T7	RT	37,100	30,500	4.4	7	44,900	1.21	1.47
	−112	40,000	31,700	4.4	5	41,000	1.02	1.29
	−320	45,600	35,200	3.3	4	44,000	0.96	1.25
A612-F	RT	43,100	34,800	3.2	7	45,500	1.05	1.31
	−112	45,500	41,000	2.7	2	50,800	1.12	1.24
	−320	53,200	49,000	2.4	3	51,000	0.96	1.04

* Broke outside middle third
† Broke in threads
‡ Failed before reaching 0.2% offset

TABLE 18

Table III. Results of Tensile Tests of Smooth and Notched Specimens from Aluminum Alloy Permanent-Mold Castings at Room and Subzero Temperatures

Alloy and temper	Temperature, °F	Tensile strength, psi	Yield strength, psi	Elongation in 4D, %	Reduction of area, %	Notch-tensile strength, psi	NTS/TS	NTS/TYS
X335-T61	RT	40,800	28,400	8.5	13	45,700	1.12	1.61
	−112	40,100	29,200	4.6	7	47,200	1.18	1.62
	−320	45,700	31,000	5.3	5	53,400	1.18	1.72
	−423	54,000	35,800	5.0	5	58,300	1.08	1.63
A344-F	RT	23,200	9,700	22.2	37	28,600	1.23	2.95
	−112	26,200	10,000	19.7	24	30,400	1.16	3.0+
	−320	37,600	12,100	13.3	12	32,000	0.85	2.64
354-T62	RT	50,100	45,500	1.1	3	54,200	1.08	1.19
	−112	54,400	45,600	1.3	2	53,200	0.98	1.17
	−320	61,000	48,700	1.3	3	56,400	0.92	1.16
	−423	60,200	56,100	0.8	1	57,200	0.95	1.02
356-T6	RT	36,800	31,100	1.0	–	43,000	1.17	1.38
	−112	42,000	34,100	3.7	5	41,400	0.98	1.21
	−320	45,700	36,500	3.2	4	45,000	0.98	1.23
356-T7	RT	28,400	21,400	4.3	7	35,300	1.24	1.65
	−112	32,600	24,300	3.7	5	37,200	1.14	1.53
	−320	37,300	25,600	3.0	4	39,900	1.07	1.56
A356-T61	RT	39,400	30,800	4.0	–	47,800	1.21	1.55
	−112	41,900	32,600	3.7	6	47,700	1.14	1.46
	−320	49,400	35,800	4.4	6	52,600	1.07	1.47
A356-T62	RT	40,900	36,700	2.1	6	46,200	1.13	1.26
	−112	45,200	39,600	3.0	5	49,800	1.10	1.26
	−320	48,600	41,400	3.0	5	57,900	1.19	1.40
	−423	55,000	45,300	3.5	3	63,300	1.15	1.40
A356-T7	RT	28,200	21,400	5.4	9	36,900	1.31	1.72
	−112	35,400	25,800	5.7	8	43,900	1.24	1.70
	−320	42,700	28,500	6.4	7	47,100	1.10	1.65
359-T62	RT	46,200	43,200	1.2	3	49,700	1.08	1.15
	−112	52,400	47,300	2.0	4	49,800	0.95	1.05
	−320	57,700	49,500	1.6	4	49,800	0.86	1.01

Table IV. Results of Tensile Tests of Smooth and Notched Specimens from Aluminum Alloy Premium-Strength Castings at Room and Subzero Temperatures*

Alloy and temper	Temperature, °F	Tensile strength, psi	Yield strength, psi	Elongation in 4D, %	Reduction of area, %	Notch-tensile strength, psi	NTS/TS	NTS/YS
C355-T61	RT	43,600	30,300	6.4	9	52,600	1.21	1.74
	−112	48,400	33,200	7.5	8	56,600	1.17	1.70
	−320	54,400	39,400	5.4	6	62,700	1.15	1.59
A356-T61	RT	41,600	30,200	8.8	10	51,400	1.23	1.70
	−112	48,200	34,800	8.9	10	55,200	1.15	1.59
	−320	51,700	38,000	4.0	4	59,800	1.15	1.57
A357-T61	RT	51,200	40,000	6.2	8	56,200	1.10	1.41
	−112	54,400	43,400	4.0	5	58,200	1.08	1.35
	−320	61,500	47,000	4.0	4	59,400	0.96	1.27
A357-T62	RT	51,200	44,400	2.5	4	55,400	1.08	1.25
	−112	53,100	46,700	2.1	3	55,200	1.04	1.18
	−320	62,200	49,300	2.5	4	59,700	0.96	1.21

* The term "premium-strength cast" is used to identify castings produced by controlled foundry practices which provide higher properties than are usually obtained by conventional practice.

TABLE 19

Table 1. Compositions of Commercial Non-Heat-Treatable Wrought Aluminum Alloys

Alloy	Al	Si	Cu	Mn	Mg	Cr	Zn	Other
	Nominal composition, %							
Super-Purity and Pure Aluminum								
1199	99.99	..	..	...	...	...	..	...
1180	99.80	..	..	...	...	...	..	...
1060	99.60	..	..	...	...	...	..	...
EC	99.45	..	..	...	...	...	..	0.02 B
1145	99.45	..	..	...	...	...	..	...
1100	99.00	..	..	...	...	...	..	...
Al-Mg Alloys — General-Purpose and Structural								
5005	Rem	..	..	...	0.8	...	..	...
5050	Rem	..	..	...	1.4	...	..	...
5052	Rem	..	..	...	2.5	0.25	..	...
5056	Rem	..	..	0.10	5.2	0.10	..	...
5082	Rem	..	..	...	4.5	...	..	...
5083	Rem	..	..	0.8	4.45	0.10	..	...
5086	Rem	..	..	0.45	4.0	0.10	..	...
5154	Rem	..	..	...	3.5	0.25	..	...
5454	Rem	..	..	0.8	2.75	0.10	..	...
5456	Rem	..	..	0.8	5.25	0.10	..	...
Al-Mg Alloys — Finishing and Decorative								
3002	Rem	..	..	0.15	0.12	...	..	...
5053	Rem	..	0.05	0.05	3.5	...	..	...
5252	Rem	..	0.1	0.1	2.5	...	..	...
5257	Rem	..	..	...	0.4	...	..	...
5357	Rem	..	..	0.30	1.0	...	..	...
5405	Rem	..	..	...	0.75	...	..	...
5457	Rem	..	0.1	0.3	1.0	...	..	...
5557	Rem	..	0.1	0.2	0.6	...	..	...
5657	Rem	..	..	...	0.8	...	..	...
Al-Mn and Al-Mn-Mg Alloys								
3003	Rem	..	..	1.2	...	...	..	...
3004	Rem	..	..	1.2	1.0	...	..	...
3005	Rem	..	..	1.2	0.35	...	..	...
3105	Rem	..	..	0.9	0.5	...	..	...
5040	Rem	..	..	1.1	1.3	0.2	..	...
Miscellaneous Alloys								
4043	Rem	5.25	..	...	...	...	..	...
4045	Rem	10.0	..	...	...	...	..	...
4145	Rem	10.0	4.0	...	...	...	..	...
4245	Rem	10.0	4.0	...	...	...	10.0	...
4343	Rem	7.5	..	...	...	...	..	...
4643	Rem	4.1	..	...	0.20	...	..	...
7072	Rem	..	..	...	...	...	1.0	...
7472	Rem	..	..	...	1.2	...	1.6	...
8001	Rem	..	..	...	...	...	..	1.1 Ni, 0.6 Fe
8081	Rem	..	1.0	...	...	...	..	20.0 Sn
8280	Rem	1.5	1.0	...	...	...	..	0.5 Ni, 6.25 Sn

TABLE 20

Table 2. Compositions of Commercial Heat Treatable Wrought Aluminum Alloys

Alloy	Al	Si	Cu	Mn	Mg	Cr	Zn	Other
	Nominal composition, %							
				Al-Cu Alloys				
2011	Rem	..	5.5	...	..	...	..	0.5 Pb, 0.5 Bi
2025	Rem	0.8	4.5	0.8	..	...	..	...
2219	Rem	..	6.3	0.3	..	...	..	0.15 Zr, 0.1 V
				Al-Cu-Mg Alloys				
2014	Rem	0.9	4.4	0.8	0.5	...	..	...
2017	Rem	..	4.0	0.50	0.50	...	..	...
2024	Rem	..	4.5	0.6	1.5	...	..	...
2117	Rem	..	2.5	...	0.30	...	..	...
				Al-Mg-Si Alloys—Structural				
2EC	Rem	0.4	...	...	0.6	...	..	...
6053	Rem	0.7	...	...	1.3	0.25	..	...
6061	Rem	0.6	0.25	...	1.0	0.20	..	...
6066	Rem	1.3	0.9	0.9	1.1	...	..	...
6070	Rem	1.3	0.25	0.7	0.8	...	..	...
6101	Rem	0.5	...	...	0.5	...	..	...
6151	Rem	1.0	...	...	0.6	0.25	..	...
6201	Rem	0.7	...	...	0.7	...	..	...
6262	Rem	0.6	0.25	...	1.0	0.09	..	0.50 Pb, 0.50 Bi
6351	Rem	1.0	...	0.6	0.6	...	..	...
6951	Rem	0.3	0.25	...	0.6	...	..	...
				Al-Mg-Si Alloys—Architectural and Decorative				
6063	Rem	0.4	...	...	0.7	...	..	...
6463	Rem	0.4	...	...	0.7	...	..	...
				Al-Zn-Mg and Al-Zn-Mg-Cu Alloys—High-Strength				
7001	Rem	..	2.1	...	3.0	0.30	7.4	...
7075	Rem	..	1.6	...	2.5	0.30	5.6	...
7076	Rem	..	0.6	0.5	1.6	...	7.5	...
7079	Rem	..	0.6	0.20	3.3	0.20	4.3	...
X7080	Rem	..	1.0	0.35	2.25	...	6.0	...
7178	Rem	..	2.0	...	2.7	0.30	6.8	...
				Al-Zn-Mg and Al-Zn-Mg-Cu Alloys—Special-Purpose				
5039	Rem	..	...	...	3.9	...	3.0	...
X5180	Rem	..	...	0.45	4.0	...	2.0	0.15 Zr, 0.15 Ti
X7005	Rem	..	...	...	1.4	0.13	4.6	0.14 Zr
7039	Rem	..	...	0.25	2.8	0.20	4.0	...
				Miscellaneous Heat Treatable Alloys				
2020	Rem	..	4.5	0.5	..	...		0.2 Cd, 1.1 Li
2218	Rem	..	4.0	...	1.5	...		2.0 Ni
2618	Rem	..	2.3	...	1.5	...		1.0 Fe, 1.0 Ni
4032	Rem	12.2	0.9	...	1.1	..		0.9 Ni

TABLE 21

Ingot No.	Casting Conditions	Degassing Treatment	As-cast Hydrogen Content, ml/100g	As-cast Ingot Porosity, % voids	Number of Oxide-Film Defects in Plate†
1	Turbulent pouring	None	0·27	0·55	18·15‡
2	Tranquil pouring, filtered*	None	0·21	0·29	5·00
3	Tranquil pouring	None	0·16	0·11	91·29
4	Tranquil pouring, filtered*	Fluxed with gaseous Cl_2 for 6 min	0·18	0·29	6·50
5	Tranquil pouring	Fluxed with gaseous Cl_2 for 6 min	0·14	0·18	80·45
6	Tranquil pouring, filtered*	Fluxed with gaseous Cl_2 for 12 min	0·18	0·10	1·50
7	Tranquil pouring	Fluxed with gaseous Cl_2 for 12 min	0·13	0·03	37·99

* Liquid metal filtered in mould through woven glass-cloth screen.
† Detected by ultrasonic inspection and micro-examination per 1-ft length of plate.
‡ Ultrasonic inspection difficult owing to porosity.

TABLE 22

Table II. Mechanical Properties of 7075-T6 Aluminum Alloy (0.025 in. Sheet, Kaiser Aluminum Company, of America, QQ-A-283)*

Test temp., °F	Direction	Hardness, 15-N	F_{ty}, ksi	F_{tu}, ksi	Elongation, %	Proportional limit, ksi	Elastic modulus, psi $\times 10^6$	Notch t. s. ($K_t = 6.3$), ksi	Notched/unnotched tensile ratio
+ 78	Long.	65	70.4	79.5	9	53.6	10.5	81.4	1.02
	Trans.	67	69.4	77.6	10	52.7	10.6	77.6	1.00
-100	Long.	66	76.2	85.1	10	66.5	10.8	84.6	0.99
	Trans.	66	74.1	83.2	9	61.6	10.2	80.6	0.97
-320	Long.	65	86.5	97.2	10	74.4	11.0	75.7	0.78
	Trans.	65	81.8	94.9	12	65.2	10.6	73.9	0.78
-423	Long.	65	100	116	8	90.7	12.5	84.3	0.73
	Trans.	66	96.6	112	12	76.8	–	78.7	0.70

*Data represent averages of a minimum of three tests in the longitudinal and two tests in the transverse directions.

TABLE 23

Table V. Mechanical Properties of X-7275-T6 Aluminum Alloy (0.050 in. Sheet, Kaiser Aluminum Company, Experimentally)*

Test temp., °F	Direction	Hardness, 15-N	F_{ty}, ksi	F_{tu}, ksi	Elongation, %	Proportional limit, ksi	Elastic modulus, psi $\times 10^6$	Notch t.s. ($K_t = 6.3$), ksi	Notched/unnotched tensile ratio
+ 78	Long.	69	75.8	86.5	14	62.5	10.0	91.0	1.05
	Trans.	67	72.6	83.6	14	63.5	9.7	89.6	1.07
- 100	Long.	66	75.6	85.2	15	65.8	10.2	95.2	1.12
	Trans.	66	73.3	83.7	14	63.6	11.1	90.6	1.08
- 320	Long.	65	84.9	96.3	6	78.4	10.2	78.6	0.82
	Trans.	67	80.1	96.2	9	66.6	11.5	81.9	0.85
- 423	Long.	65	98.4	114	5	87.8	11.9	75.3	0.67
	Trans.	65	95.2	108	4	85.6	11.4	81.6	0.76

* Data represent averages of a minimum of three tests in the longitudinal and two tests in the transverse directions.

TABLE 24

Material	Temper	Form	Thickness, in.	Hardness 15-N superficial	Composition								Specification	Producer
					Cr	Cu	Fe	Mn	Si	Ti	Zn	Mg		
7075	T6	Sheet	0.025	63	0.23	1.47	0.32	0.04	0.15	.05	5.85	2.45	QQ-A-283	Alcoa
7079	T6	Sheet	0.080	62	0.14	0.67	-	0.18<	0.1	.056	3.70	4.26	Experimental	Kaiser
7178	T6	Sheet	0.036	66	0.26	1.95	0.21	0.04	0.11	.02	6.6	3.0	Mil-A-9180A	Kaiser
X7275	T6	Sheet	0.050	65	0.16	1.40<	0.1	0.04<	0.1	.035	5.95	2.94	Experimental	Kaiser
7075	T6	Plate	2.5	62	0.1[illegible]	1.43	0.25	0.04	0.14	.039	5.46	2.75	QQ-A-283	Kaiser
7079	T6	Billet	5.0	61	0.13	0.71	-	0.14<	0.1	.060	3.30	4.15	0-01041	Alcoa

TABLE 25

Table 2. Data Illustrating Directionality of Fracture Toughness

Alloy and temper	Product	Plane-strain fracture toughness, K_{Ic}, psi $\sqrt{in.}$ L(RW)	LT(WR)	ST(TR)
2014-T651	5-in. plate	20,800	20,600	18,500
7075-T651	1¾-in. plate	27,000	22,300	14,800
7079-T651	1¾-in. plate	27,000	23,900	16,200
7075-T6511	3½ × 7½-in. extruded bar	30,900	20,800	19,000
7178-T6511	3½ × 7½-in. extruded bar	22,700	15,600	14,000

TABLE 26

Table IV. Plane Strain Fracture Toughness of 2219 Aluminum Alloy

Temper	Material form	Specimen type	Specimen geometry, in.	Grain direction	Temperature, F	Fracture/yield strength ratio	Fracture toughness, K_{Ic}, 10^3 psi $\sqrt{in.}$
−T81	5.0-in. plate	NRB	$D = 2.0$, $d = 1.4$	Longitudinal	70	0.84	25.7
					−320	0.81	31.0
				Transverse	70	0.85	25.5
					−320	0.82	28.9
				Short transverse	70	0.68	19.9
					−320	0.65	21.3
			$D = 5.0$, $d = 3.5$	Longitudinal	70	0.64	29.4
					−320	0.60	36.0
		SC	$W = 6$, $B = 2$	Longitudinal	70	0.63	37.4
			$W = 6$, $B = 1$		−320	0.63	43.0
			$W = 6$, $B = 2$	Transverse	70	0.51	29.3
			$W = 6$, $B = 1$		−320	0.52	34.9
	0.125-in. sheet	SC	$W = 1$, $B = 0.125$	Longitudinal	70	1.25	
					−110	1.21	
					−320	1.21	
			$W = 1.5$, $B = 0.125$		−423	0.87	25.9
−T87	1.0-in. plate	SC	$W = 6$, $B = 1$	Longitudinal	70	0.94	49.1
					−320	0.89	54.9
				Transverse	70	0.79	40.0
					−320	0.72	44.5
	1.0-in. plate	SNB	$W = 1$, $B = 0.5$, $a = 0.25$	Longitudinal (LS)†	70		33.8
					−110		32.3
					−320		39.8
					−423		42.1
	1.25-in. plate	SC	$W = 6$, $B = 0.6$	Transverse	70	0.73	33.8
					−320	0.65	37.6
		SC	$W = 6$, $B = 1.25$	Transverse	70	0.63	35.0
					−320	0.59	37.9
				Transverse (TS)	70		28.7
					−110		28.6
					−320		32.9
					−423		37.3

TABLE 27

Table IV. Notch-Ultimate Strength and Notch-Yield ratios for AZ5G Alloy T6 Sheet (transverse and longitudinal to rolling direction), $K_t = 6.3$

Temperature, °K		Ultimate notch strength / Ultimate tensile strength ratio	Ultimate notch strength / Tensile yield strength ratio
300	T	1.01	1.19
	L	0.14	1.24
77	T	0.95	1.23
	L	0.98	1.29
20	T	0.85	1.23
	L	0.86	1.27

TABLE 28

TABLE 1. Impact Strength of Parts Made from V95 and VM65-1 Alloys as a Function of the Direction in Which the Specimens were Cut Out and the Orientation of the Notch Relative to the Plane of Rolling or Pressing

Material	Angle between the longitudinal axis of specimen and the direction of rolling, deg	a_n, kgf·m/cm²	
		Notch parallel to deformation plane specimen 1, Fig. 1	Notch normal to deformation plane specimen 2, Fig. 1
Pressed strip V95 alloy	0	0.85	0.74
	15	0.83	0.68
	30	0.69	0.66
	45	0.64	0.60
	60	0.50	0.56
	75	0.37	0.45
	90*	0 40	0.42
	90 †	0 24	0.29
Pressed strip VM65-1 alloy	0	0.81	0.64
	15	0.77	0.61
	30	0.86	0.63
	45	0.84	0.63
	60	0.80	0.63
	75	0.74	0.61
	90*	0.73	0.60
	90**	0.70	0.68
Rolled plate V95 alloy	0	0.88	0.89
	15	0.84	0.83
	30	0.78	0.75
	45	0.60	0.65
	60	0.56	0.57
	75	0.56	0.57
	90*	0.47	0.53

* Along the width.

† Along the thickness.

TABLE 29

Table 2. Energy and shear lip values of fracture tests on 7075–T6 and –T651 (Avg. 3 tests)

Center-slotted specimens					Precracked charpy	
Thickness (in.)	Specimen width (in.)	G_{Ic} (in. lb/in^2)	G_c (in. lb/in^2)	Shear lip (%)	W/A (in. lb/in^2)	Shear lip (%)
			Longitudinal			
$\frac{1}{16}$	3		305‖			
	9		270	100	192	100
$\frac{1}{8}$	3	66	276			
	9	88 §	380	73	166	52
$\frac{3}{16}$	3	85	431‖			
	9	112 §	515	100	241	40
$\frac{1}{4}$	9	107 §	468	24	155	28
$\frac{5}{16}$	9	113 §	473	75	161	21
1†	20	83 §	416	0‡	–	–
			Transverse			
$\frac{1}{16}$	3		292‖			
	9		222	100	137	50
$\frac{1}{8}$	3	62	173			
	9	75 §	250	60	118	22
$\frac{3}{16}$	3	62	271‖			
	9	88 §	344	35	153	19
$\frac{1}{4}$	9	95 §	316	28	109	12
$\frac{5}{16}$	9	83 §	172	27	98	10
1†	20	59 §	93	0‡	–	–

†See reference [14].
‡Estimated from photograph in [14].
§Apparent value – no fatigue crack.

TABLE 30

Table 3—Average Results of Tear Tests of Groove-Welded Panels, of Aluminum Alloy Plate[a] at Room Temperature[c]

		Along center of weld (Fig. 1a)						Line of fusion (Fig. 1c)						Across weld (Fig. 1b)					
				Energy required to						Energy required to						Energy required to			
Filler alloy	Base metal alloy and temper	Tear strength, psi	Tear strength / Joint YS	Initiate a crack, in.-lb	Propagate a crack, in.-lb	Total energy, in.-lb	Unit propagation energy, in.-lb/in.²	Tear strength, psi	Tear strength / Joint YS	Initiate a crack, in.-lb	Propagate a crack, in.-lb	Total energy, in.-lb	Unit propagation energy, in.-lb/in.²	Tear strength, psi	Tear strength / Joint YS	Initiate a crack, in.-lb	Propagate a crack, in.-lb	Total energy in.-lb	Unit propagation energy, in.-lb/in.²
1100	1100-H112	19,500	3.20	48	76	124	755	18,400	3.02	44	88	132	880	20,800	3.41	57	84	141	840
	3003-H112	24,000	3.16	40	78	118	785	24,400	3.22	46	111	157	1110	24,500	3.23	41	91	132	910
4043	6061-T6	35,500	2.34	19	38	57	380	39,200	2.58	37	41	78	410	36,600	2.41	16	12	28	120
	6061-T6[b]	36,100	0.93	5	8	13	80	54,100	1.40	13	18	31	180	60,700	1.57	19	7	26	70
5052	5052-H112	37,000	2.66	45	108	153	1085	...	...	..	...	...	...	37,400	2.69	67	157	224	1570
5154	5154-H112	36,200	2.50	50	104	154	1040	43,100	2.97	86	124	210	1240	37,400	2.58	44	135	179	1350
5456	5456-H321	51,700	1.70	46	92	138	920	52,100	1.72	48	84	132	840	47,400	1.56	41	95	136	950

[a] Specimen 0.10 in. thick. [b] Heat-treated and aged after welding. [c] All ratios based on tensile properties shown in Table 2.

TABLE 31

Alloy and Temper	Longitudinal	Transverse
2020-T6	121	49
7001-T75	128	86
2024-T851	296	124
7178-T7651	310	152
7079-T651	550	175
7075-T651	685	151
2219-T851	770	383
7075-T7351	1000	510
2024-T351	1450	1140
5456-H321 (welded 5556)	1740	1450
5456-H321	2395	1340
5456-O (welded 5556)	2670	2220
5456-O	4580	2400
X7005-T6351	3920	2950

TABLE 31- CONT.

Table 4—Average Results of Tear Tests[a] of Groove Welded Panels of Aluminum Alloy Sheet and Plate[b] at Room Temperature and —320° F[e]

		Room temperature						—320° F						320°/RT	
				Energy required to						Energy required to					
Filler alloy	Base metal alloy and temper	Tear strength, psi	Tear strength, Joint YS	Initiate a crack, in.-lb	Propagate a crack, in.-lb	Total energy in.-lb.	Unit propagation energy, in.-lb/in.^{2}	Tear strength, psi	Tear strength, Joint YS	Initiate a crack, in.-lb	Propagate a crack, in.-lb	Total energy, in.-lb	Unit propagation energy, in.-lb/in.^{2}	Tear strength, psi	Unit propagation energy, in.-lb/in.^{2}
1100	1100-H112	19,500	3.20	48	76	124	755	32,200	4.02	82	122	204	1220	1.65	1.62
	3003-H112	24,000	3.16	40	78	118	785	43,500	4.02	93	126	219	1260	1.81	1.61
2319	2219-T81, -T87	68,300	2.73	26	21	47	346	86,200	2.94	80	70	100	1106	1.26	3.20
	2219-T81, -T87[c]	73,200	2.04	18	24	43	391	89,400	2.07	28	69	97	1106	1.22	2.83
	2219-T62[d]	87,200	2.10	33	44	77	705	104,800	2.06	35	39	74	624	1.20	0.88
5052	5052-H112	37,000	2.66	45	108	153	1085	50,700	3.11	85	178	263	1780	1.37	1.64
5154	5154-H112	36,300	2.50	50	104	154	1040	43,900	2.60	50	107	157	1070	1.21	1.03
5183	5083-0	50,200	...	38	97	135	970	54,600	...	33	74	107	740	1.09	0.76
	-H113	51,600	...	33	99	132	990	57,200	...	34	70	114	700	1.11	0.71
5456	5456-H321	51,700	1.70	46	92	138	920	58,200	...	40	116	156	1160	1.13	1.26
5556	5083-H113	48,200	...	36	85	121	850	55,500	...	33	83	116	830	1.15	0.98
	5456-0	51,700	...	38	91	129	910	56,800	...	36	76	112	760	1.10	0.84

[a] Crack along center of weld, Fig. 1a.
[b] 1 in. thick except 2319, which was 0.063 in. thick.
[c] Aged after welding.
[d] Heat treated and aged after welding.
[e] All ratios based on tensile properties shown in Table 2.

TABLE 32

Table 5—Average Results of Tensile Tests of Notched Specimens[a] From Groove Welded Panels of Aluminum Alloy Plate at Room Temperature and —320° F[c]

		Room Temperature				—320° F				Ratio	
Filler metal	Base metal alloy and temper	Notch-tensile strength, psi	Reduction of area, %	Notch-strength ratio, NTS/TS	Notch-yield ratio, NTS/TYS	Notch-tensile strength, psi	Reduction of area, %	Notch strength ratio, NTS/TS	Notch yield ratio, NTS/TYS	NTS (—320) / NTS(RT)	NYS (—320) / NYR(RT)
1100	1100-H112	17,800	5	1.53	2.92	32,200	9	1.41	4.03	1.81	1.38
	3003-H112	22,700	6	1.41	2.99	...	...	...	...	...	...
4043	6061-T6	27,500	3	1.05	1.81	33,800	2	0.87	1.86	1.23	1.03
	6061-T6[b]	42,300	6	0.98	1.10	47,900	1	0.90	1.02	1.13	0.93
5052	5052-H112	32,800	5	1.13	2.36	45,500	7	0.99	2.79	1.33	1.18
5154	5052-H112	32,100	4	1.10	2.34	45,700	6	1.00	2.76	1.42	1.18
	5154-H112	34,100	5	1.05	2.35	43,000	6	0.88	2.54	1.26	1.08
	6061-T6	33,800	3	1.35	2.41	...	...	...	...	...	...
	6061-T6[b]	42,900	4	1.20	1.57	...	...	...	...	...	...
5356	5086-H32	41,400	5	1.07	2.17	48,400	4	0.92	2.38	1.17	1.10
5456	5456-0	40,700	4	0.93	1.87	48,500	4	0.85	1.95	1.19	1.04
5554	5454-H32	39,300	5	1.16	2.30	52,800	7	0.97	2.35	1.34	1.02
5556	5083-H113	42,000	5	1.02	1.97	55,800	3	0.93	2.26	1.33	1.14
	5456-H321	45,200	5	1.01	2.01	52,400	3	0.89	2.00	1.16	1.00

[a] Notched specimens per Fig. 5. K ~ 16.
[b] Heat treated and aged after welding.
[c] All ratios based on tensile properties shown in Table 2.

TABLE 33

Table II. Results of Tensile and Notch-Tensile Tests of Some Groove Welded Panels of Wrought Aluminum Alloys at Room and Subzero Temperatures

Product	Alloy and temper	Thickness, in.	Filler alloy	Post-weld thermal treatment	Temp., °F	Tensile strength, psi	Joint yield strength,* psi	Elongation in 4D, %	Reduction of area, %	Joint efficiency,** %	Location of failure§	Notch-tensile strength,† psi	Ratios NTS/TS	Ratios NTS/JYS
Plate	2219-T62	1	2319	Yes	RT	57,300	40,200	7.5	7	99	C	63,700	1.11	1.58
					−112	60,400	40,500	6.5	8	96	C	68,600	1.14	1.69
					−320	68,900	46,600	5.5	6	94	C	74,800	1.09	1.61
					−452	72,000	51,500	3.5	5	80	C	82,800	1.15	1.61
	2219-T851	1¼	2319	No	RT	32,700	26,800	2.0	5	50	C	40,700	1.24	1.52
					−112	40,800	25,000	4.0	15	57	C	48,300	1.18	1.93
					−320	51,700	28,000	3.5	10	62	C	48,500	0.94	1.73
					−452	59,600	40,200	2.5	10	62	C	52,700	0.88	1.31
	3003-H112	1	1100	No	RT	16,100	7,600	24.0	67	100	C	22,700	1.41	3.00
					−112	19,300	8,300	26.5	66	100	C	—	—	—
					−320	33,700	10,800	31.0	52	100	C	—	—	—
					−452	51,100	18,500	28.0	25	100	C	39,800	0.78	2.15
	5083-0	1	5183	No	RT	42,500	20,100	21.5	—	100	A	44,700	1.05	2.22
					−112	43,900	20,700	31.0	44	100	C	49,900	1.14	2.41
					−320	58,200	22,400	19.0	20	99	A	50,100	0.86	2.24
					−452	55,300	25,200	27.0	37	69	B	53,900	0.98	2.14
	5083-H321	1	5183	No	RT	44,200	26,000	14.0	39	96	C	54,500	1.23	2.10
					−112	47,000	26,200	19.0	48	98	A	59,500	1.27	2.27
					−320	64,700	31,400	19.0	23	100	A	62,400	0.96	1.98
					−452	66,100	35,700	9.0	14	80	A	58,800	0.89	1.65
		1	5356	No	RT	41,500	24,300	13.5	47	90	A	53,800	1.30	2.22
					−112	43,900	27,000	14.5	52	91	A	57,500	1.31	2.13
					−320	61,900	29,100	15.5	33	97	A	60,600	0.98	2.08
					−452	66,000	34,100	9.0	17	80	C	57,700	0.87	1.69
Plate	5083-H321	1	5556	No	RT	44,400	25,600	14.0	36	97	A	53,700	1.21	2.10
					−112	46,300	26,700	18.5	46	96	A	58,100	1.26	2.19
					−320	65,300	30,600	20.5	26	100	A	60,500	0.93	1.98
					−452	68,800	34,600	13.0	17	83	A	57,900	0.84	1.68
	5454-H32	1	5554	No	RT	33,900	17,100	18.0	42	85	A	39,300	1.16	2.30
					−112	36,000	17,400	22.0	47	86	A	—	—	—
					−320	54,700	22,500	29.0	29	93	A	52,800	0.97	2.35
					−452	61,400	26,100	14.5	—			47,900	0.78	1.91
	6061-T6	1	4043	No	RT	31,000	20,900	6.0	19	69	C	34,000	1.10	1.63
					−112	34,600	23,600	6.0	19	71	A	38,600	1.12	1.64
					−320	44,000	25,800	5.5	12	75	A	39,600	0.90	1.54
					−452	49,100	37,600	4.5	9	63	A	39,900	0.81	1.06
				Yes	RT	43,300	35,900	11.0	44	96	B	57,500	1.31	1.57
					−112	47,800	38,300	21.5	38	98	B	61,500	1.27	1.60
					−320	57,300	42,300	16.5	12	97	A	64,800	1.13	1.53
					−452	65,600	44,800	15.0	16	84	A	67,200	1.02	1.50
		1	5356	No	RT	32,700	22,600	8.0	31	73	A	46,900	1.44	2.07
					−112	37,100	24,700	9.0	36	76	B	50,100	1.35	2.03
					−320	47,000	27,300	13.5	39	80	B	54,100	1.15	1.98
					−452	57,700	35,300	13.5	24	74	A	53,300	0.92	1.50
				Yes	RT	40,500	29,300	9.5	33	90	B	—	—	—
					−112	46,400	35,100	12.0	44	95	A	57,800	1.25	1.65
					−320	57,100	33,900	20.0	29	97	B	66,400	1.16	1.96
					−452	69,100	44,500	19.0	24	89	A	60,800	0.88	1.37

*Offset equals 0.2%; gauge length, 2.0 in.
**Based on typical values.
†All failures occurred through weld.
§A—through weld.
B—½ to 2½ in. from weld.

TABLE 34

Table 1—Chemical Composition of Aluminum Alloys

Alloy	Abbreviated composition	Form	Nominal composition, %[a]						Composition by analysis, %[a]	
			Mg	Si	Mn	Cu	Cr	Ti	Mg	Si
Base metal										
6061		Sheet No. 1	1.0	0.6	...	0.25	0.25	...	0.98	0.56
		Sheet No. 2	1.0	0.6	...	0.25	0.25	...	1.04	0.51
		Sheet No. 3	1.0	0.6	...	0.25	0.25	...	1.03	0.59
		Sheet No. 4	1.0	0.6	...	0.25	0.25	...	0.87	0.68
Filler metal										
4043	Al5Si	Wire	...	5.0	...	...	...	...	0.26	5.06
		Rod	...	5.0	...	...	...	...	0.24	5.31
718	Al12Si	Wire	...	12	...	...	...	...	0.22	11.69
		Rod	...	12	...	...	...	...	0.30	11.10
5554	Al3Mg	Wire	2.8	...	0.8	...	0.1	0.1	3.25	0.04
5556	Al5Mg	Wire	5.2	...	0.8	...	0.1	0.1	5.06	0.05
		Rod	5.2	...	0.8	...	0.1	0.1	5.01	0.05

[a] Balance consists of Al and normal impurities.

Table 1—Nominal Compositions of Filler Metal Alloys

Alloy	Element, %							
	Si	Cu	Mn	Mg	Cr	Ti	V	Zr
1100				99.00% Al, min				
2319	...	6.3	0.3	...	...	0.06	0.10	0.18
4043	5.0	...	...	...	...	...	...	...
5052	...	...	...	2.5	0.25	...	...	...
5154	...	...	...	3.5	0.25	...	...	...
5183	...	...	0.8	4.45	0.10	0.1	...	...
5456[a]	...	...	0.8	5.25	0.10	...	...	...
5556	...	...	0.8	5.25	0.10	0.1	...	...

[a] Filler alloy 5456 has been replaced by 5556 and is no longer commercially available. Its mechanical properties are essentially the same as those of 5556.

TABLE_35

Table II. Notched and Unnotched Tensile Properties of 3/4-in. Plate and Butt MIG Weldments of 5083-H113, 7075-T6, and 7079-T6

Material	Test temp., °F	Unnotched tensile properties: TS, psi	YS, psi	Elong., % in 2 in.	Notched (K_t = 6.3) TS, psi	Ratio: Notched TS / Unnotched TS
5083-H113	(Min*)	44,000	31,000	12	-	-
	75	47,400	34,100	17	61,800	1.30
	-320	67,900	39,600	31	73,300	1.08
5083-H113 MIG 5183 weld	75	42,800	21,100	15	47,000	1.10
	-320	61,400	24,100	20	57,200	0.92
7075-T6	(Min*)	77,000	66,000	6	-	-
	75	84,300	74,700	9	89,300	1.06
	-320	99,900	88,200	2	77,000	0.77
7075-T6 MIG 718 weld	75	36,700	26,000	2	31,400	0.86
	-320	44,500	34,300	1	38,100	0.86
7079-T6	(Min*)	74,000	65,000	8	-	-
	75	77,900	70,300	11	98,300	1.26
	-320	94,300	84,900	10	73,500	0.78
7079-T6 MIG X5184 weld	75	51,800	35,300	5	50,000	0.97
	-320	57,700	43,500	2	46,500	0.81

TABLE 36

			Longitudinal						Long-transverse					
Filler alloy	Base metal alloy and temper	Temp, °F	Tensile strength, psi	Joint yield strength,[a] psi	Elong in 2 in., %	Notch-tensile strength,[b] psi	NTS/TS	NTS/JYS	Tensile strength, psi	Joint yield strength,[a] psi	Elong in 2 in., %	Notch-tensile strength,[b] psi	NTS/TS	NTS/JYS
2319	2219-T37	RT	41,800	28,700	4.0	36,600	0.88	1.28	42,700	27,400	4.0	38,300	0.90	1.40
		−320	58,600	35,500	6.9	49,400	0.84	1.38	56,100	36,600	4.0	49,900	0.89	1.36
		−423	58,500	43,500	2.0	54,400	0.93	1.25	55,800	42,800	1.6	54,000	0.97	1.26
2319	2219-T37[c]	RT	43,200	39,100	1.9	43,700	1.01	1.12	42,700	38,200	2.3	41,100	0.96	1.08
		−320	54,600	48,300	1.9	55,600	1.02	1.15	53,400	44,700	1.7	52,600	0.99	1.18
		−423	63,000	51,900	1.3	55,300	0,88	1.07	56,800	48,200	1.0	54,800	0.99	1.14
2319	2219-T62[d]	RT	60,500	43,500	7.5	59,500	0.98	1.37	60,200	42,800	9.2	58,000	0.96	1.36
		−320	75,200	51,800	7.5	73,000	0.95	1.41	73,800	51,100	6.2	70,000	0.95	1.37
		−423	81,600	58,500	4.0	78,700	0.96	1.35	79,400	57,100	4.3	74,400	0.92	1.30
2319	2219-T62	RT	44,000	30,500	2.7	47,300	1.08	1.50	45,200	30,800	3.5	44,900	0.99	1.46
		−320	54,300	39,700	2.3	56,500	1.04	1.42	58,100	35,600	4.7	56,700	0.98	1.59
		−423	64,000	50,000	6.0	59,200	0.92	1.18	60,000	41,500	2.0	59,800	1.00	1.44
2319	2219-T87	RT	45,300	32,800	2.3	41,700	0.92	1.27	44,600	31,000	2.2	44,200	0.99	1.43
		−320	59,200	39,000	3.0	53,800	0.91	1.38	58,000	34,600	4.1	55,100	0.95	1.59
		−423	65,900	45,300	2.3	57,800	0.88	1.28	62,200	44,800	2.1	59,000	0.95	1.32
4043	2014-T3	RT	50,000	41,500	2.8	47,400	0.95	1.14	47,600	38,600	1.0	49,500	1.04	1.28
		−320	61,100	50,800	2.2	57,400	0.94	1.13	56,400	47,600	1.3	57,000	1.23	1.20
		−423	65,800	62,300	0.7	61,400	0.93	0.98	61,000	58,600	0.5	64,100	1.05	1.09
4043	2014-T3[c]	RT	54,800	54,800[e]	...	52,900	0.97	0.97	49,300	49,300[e]	...	48,900	0.99	0.99
		−320	62,900	62,900[e]	...	61,200	0.97	0.97	58,800	58,800[e]	...	60,900	1.03	1.03
		−423	66,900	62,300[e]	...	68,400	1.02	1.10	67,600	67,600[e]	...	65,200	0.96	0.96
4043	2014-T6	RT	46,300	37,800	2.8	42,900	0.96	1.13	...	...	...	...	...	...
		−320	60,900	46,800	2.0	45,700	0.75	0.98	...	...	...	...	...	...
		−423	60,000	51,000	1.2	50,900	0.85	1.00	...	...	...	...	...	...
4043	6061-T6	RT	32,200	23,200	5.3	34,100	1.06	1.47	...	...	...	...	...	...
		−320	47,400	28,800	10.5	38,000	0.80	1.32	...	...	...	...	...	...
		−423	65,500	32,000	10.8	41,800	0.64	1.31	...	...	...	...	...	...
4043	7075-T6	RT	46,500	45,000	1.0	43,600	0.94	0.97	...	...	...	...	...	...
		−320	52,300	51,600	1.0	42,200	0.81	0.82	...	...	...	...	...	...
		−423	65,100	65,100[e]	...	39,800	0.61	0.61	...	...	...	...	...	...
5556	5456-H321	RT	51,900	32,600	13.0	54,400	1.05	1.67	51,700	30,300	8.5	51,800	1.00	1.71
		−320	63,700	37,700	15.8	59,700	0.93	1.58	63,200	35,800	9.5	57,100	0.90	1.60
		−423	59,100	49,500	3.8	59,700	1.01	1.20	62,100	55,000	5.0	55,800	0.89	1.01
5556	5456-H343	RT	51,900	30,700	7.0	53,200	1.03	1.73	51,800	30,100	7.3	54,000	1.04	1.79
		−320	59,100	34,100	6.5	59,000	1.00	1.73	63,700	34,300	7.2	57,500	0.90	1.68
		−423	58,800	38,600	3.0	59,900	1.02	1.55	60,900	39,200	3.5	57,500	0.95	1.47

TABLE 37

Filler alloy	Base metal alloy and temper	Temp, °F	Longitudinal: Tensile strength, psi	Longitudinal: Joint yield strength,[a] psi	Longitudinal: Elong in 2 in., %	Longitudinal: Notch-tensile strength,[b] psi	Longitudinal: NTS/TS	Longitudinal: NTS/JYS	Long-transverse: Tensile strength, psi	Long-transverse: Joint yield strength,[a] psi	Long-transverse: Elong in 2 in., %	Long-transverse: Notch-tensile strength,[b] psi	Long-transverse: NTS/TS	Long-transverse: NTS/JYS
5556	7079-T6	RT	50,700	42,300	1.5	52,900	1.04	1.25	49,900	42,100	1.3	55,300	1.11	1.31
		−320	57,200	54,300	1.2	59,200	1.04	1.09	54,800	57,000	1.4	57,000	...	1.04
		−423	57,200	...	0.7	63,200	1.10	1.10	63,600	62,300	1.0	49,400	0.77	0.79
5556	7178-T6	RT	54,100	51,000	1.3	53,000	0.98	1.04	46,400	46,400[c]	...	49,600	1.06	1.06
		−320	53,400	53,400[c]	...	59,500	1.11	1.11	54,300	54,300[c]	...	54,100	1.00	1.00
		−423	55,600	55,600[c]	...	49,800	0.90	0.90	56,400	56,400[c]	...	44,500	0.79	0.79

[a] 2 in. gage length, offset = 0.2%.
[b] Sharply notched specimens (see Fig. 5), $K_t \geq 17$.
[c] Failed before reaching 0.2% offset.
[d] Heat treated and aged after welding.
[e] Aged after welding.

TABLE 37 cont.

Condition	Direction	Room temperature: Maximum load, lb	Tear strength, psi	Ratios S_t/TS	Ratios S_t/TYS	Energy req'd to Initiate a crack, in.-lb	Energy req'd to Propagate a crack, in.-lb	Total energy, in.-lb	Unit propagation energy, in.-lb/in.²	−320°F: Maximum load, lb	Tear strength, psi	Ratios S_t/TS	Ratios S_t/TYS	Energy req'd to Initiate a crack, in.-lb	Energy req'd to Propagate a crack, in.-lb	Total energy, in.-lb	Unit propagation energy, in.-lb/in.²	Ratios: −320°/RT Tear strength, psi	Ratios: −320°/RT Unit propagation energy, in.-lb/in.²
-T62	L	1071	68,000	1.12	1.60	15	36	51	578	1361	86,600	1.11	1.64	19	38	57	605	1.28	1.05
	T	1053	66,800	1.09	1.56	15	33	48	527	1258	80,400	1.02	1.57	17	34	51	539	1.20	1.02
-T81	L	1140	72,400	1.11	1.40	14	26	40	410	1309	83,200	0.99	1.30	18	36	54	576	1.15	1.40
	T	1108	70,400	1.07	1.34	14	27	41	429	1387	87,600	1.03	1.33	16	32	48	508	1.25	1.18
-T87	L	1185	75,200	1.11	1.34	17	—	—	—	1347	85,600	0.98	1.24	17	25	42	392	1.14	—
	T	1058	67,200	0.99	1.19	10	21	31	338	1274	81,000	0.93	1.16	14	25	39	397	1.21	1.18
-T81 (a)	T	1115	70,800	1.52	2.13	31‡	20‡	51‡	324	1303	82,900	1.22	2.12	28	60	88	948	1.17	2.92
-T87 (a)	T	1055	67,000	1.45	2.10	24	22	46	352	1410	89,600	1.33	2.41	33	80	113	1265	1.34	3.59
-T81 (b)	T	1167	74,100	1.53	1.84	20	23	43	363	1431	91,000	1.36	1.82	32	74	106	1180	1.23	3.25
-T87 (b)	T	1140	72,400	1.38	1.79	17	26	43	419	1383	87,900	1.26	1.95	25	64	89	1032	1.21	2.46
-T62 (c)	T	1375	87,200	1.42	2.04	33	44	77	705	1658	104,800	1.32	1.94	35	39	74	624	1.20	0.88

* Average of two tests.
† Inert-gas consumable-electrode method, 2319 I.G. Electrode.
‡ Single test.
S_t Tear Strength.
TS—Tensile Strength.
TYS—Tensile Yield Strength.
(a)—No subsequent thermal treatment after welding.
(b)—Aged after welding.
(c)—Heat-treated and aged after welding.

TABLE 38

Table 10—Fracture Toughness of Weld Metal and Transition Zone

Group	Filler metal	Weld-metal composition, % Si	Weld-metal composition, % Mg	Heat treatment[a]	Fracture toughness, in.-lb/in. — Weld metal Avg[b]	Weld metal s[c]	Transition zone Avg[b]	Transition zone s[c]
1	Al5Si	1.5	0.8	AA	94(3)	5	297(5)	50
				SQ	204(2)	...	491(5)	95
2	Al12Si	2.9	0.9	AA	70(4)	11	184(4)	45
				SQ	176(2)	...	405(2)	...
3	Al5Si	3.2	0.6	AA	150(4)	34	292(8)	39
				SQ	247(2)	...	489(4)	90
4	Al12Si	6.6	0.6	AA	86(4)	12	121(10)	46
				SQ	278(2)	...	337(6)	40
5	Al3Mg	0.5	1.6	AA	475(4)	32	305(8)	121
				SQ	640(2)	...	559(3)	89
6	Al5Mg	0.5	1.9	AA	535(2)	...	456(2)	...
				SQ	1202(2)	...	759(3)	123
7	Al5Mg	0.3	3.2	AA	620(2)	...	454(3)	10
				SQ	1177(2)	...	557(2)	...

[a] Heat treatment after welding: AA—artificial aging; SQ—solution quenching and artificial aging.
[b] The number of specimens included in the average is given in parenthesis.
[c] Standard deviation.

TABLE 39

Parent metal	Filler wire	Condition	Sample number	Room temperature, Smooth specimens: Tensile strength, psi	RT Smooth: Yield strength, psi	RT Smooth: Elongation in 2 in., %	RT Smooth: Reduction of area, %	RT Notched specimens: Tensile strength, psi	RT Notched: Reduction of area, %	RT Notched: Notch strength, ratio	-320°F, Smooth specimens: Tensile strength, psi	-320°F Smooth: Yield strength, psi	-320°F Smooth: Elongation in 2 in., %	-320°F Smooth: Reduction of area., %	-320°F Notched specimens: Tensile strength, psi	-320°F Notched: Reduction of area., %	-320°F Notched: Notch strength, ratio	Ratio * NTS(-320)/NTS(RT)
1100-H112	1100	AW	147079	11 600	6 100	26.5	84	17 800	5	1.53	22 800	8 000	31.0	73	32 200	9	1.41	1.81
3003-H112	1100	AW	147080	16 100	7 600	24.0	67	22 700	6	1.41	33 700	10 800	31.0	52	—	—	—	—
5052-H112	5052	AW	147081	29 100	13 900	18.0	42	32 800	5	1.13	45 800	16 300	25.0	27	45 500	7	0.95	1.33
5052-H112	5052	AW	147082	29 200	13 700	15.0	21	32 100	4	1.10	45 900	16 500	26.0	29	45 700	6	1.00	1.42
5154-H112	5154	AW	147083	32 600	14 500	17.0	29	34 100	5	1.05	48 700	16 900	27.5	31	43 000	6	0.88	1.26
5454-H32	5554	AW	206518	33 900	17 100	18.0	42	39 300	5	1.16	54 700	22 500	29.0	29	52 800	7	0.97	1.34
5086-H32	5356	AW	206517	38 500	19 100	18.0	19	41 400	5	1.07	52 800	20 300	17.0	21	48 400	4	0.92	1.17
5083-H113	5556	AW	206516	41 200	21 200	12.5	22	42 000	5	1.02	60 100	24 700	20.0	21	55 800	3	0.93	1.33
5456-0	5456	AW	174673	43 900	21 700	13.0	25	40 700	4	0.93	56 700	24 700	18.0	26	48 500	4	0.85	1.19
											57 800	25 000	18.0	19				
5456-H321	5556	AW	206519	44 600	22 500	13.0	23	45 200	5	1.01	59 000	26 200	14.5	18	52 400	3	0.89	1.16
6061-T6	4043	AW	147084	26 100	15 200	12.0	41	27 500	3	1.05	38 800	18 200	7.5	12	33 800	2	0.87	1.23
6061-T6	4043	HTA	147085	43 200	38 600	2.0	8	42 300	0	0.98	53 400	46 800	3.0	5	47 900	1	0.90	1.13
6061-T6	5154	AW	147086	25 000	14 000	13.0	38	33 800	3	1.35	41 100	19 000	22.0	39	—	—	—	
6061-T6	5154	HTA	147087	35 600	27 300	5.5	21	42 900	4	1.20	51 400	32 700	5.5	24	—	—	—	

$^{*}\dfrac{\text{Notch Tensile Strength }(-320)}{\text{Notch Tensile Strength (Room Temperature)}}$

TABLE 40

Table II. Results of Tear Tests of 5083-0 Plate and of Groove Welds†† in 5083-0 Plate and Extrusions at Room Temperature and at −320°F

Identification	Spec. No.	Tensile properties: Temperature, °F	Tensile properties: Tensile strength, psi	Tensile properties: Yield strength,* psi	Tensile properties: Elong. in 4D, %	0.1-in.-thick specimen (Fig. 1a): Tear strength, psi	0.1-in.: Tear strength / Yield strength	0.1-in.: Energy required to Initiate a crack, in.-lb	0.1-in.: Energy required to Propagate a crack, in.-lb	0.1-in.: Total energy, in.-lb	0.1-in.: Unit propagation energy, in.-lb/in.²	0.6-in.-thick specimen (Fig. 1b): Tear strength, psi	0.6-in.: Tear strength / Yield strength	0.6-in.: Energy required to Initiate a crack, in.-lb	0.6-in.: Energy required to Propagate a crack, in.-lb	0.6-in.: Total energy, in.-lb	0.6-in.: Unit propagation energy, in.-lb/in.²	Notes
5083-0 Plate, longitudinal	1	RT	43,200	19,000	24.5	49,700		48	128	176	1280	49,000		634	1351‡	1985	1710‡	
	2					50,400		61	114	175	1145	49,300		714	1153	1867	1460	
	Avg.					50,000	2.63	55	121	176	1215	49,200	2.58	674	1153	1867	1460	
	1	−320	†	†	†	57,600		87	129	216	1290	59,600		921	1425	2346	1805	
	2					58,600		91	110	201	1100	59,300		949	1359	2308	1720	
	Avg.					58,100	†	89	120	209	1195	59,400	†	935	1392	2327	1760	
5083-0 Plate, transverse	1	RT	43,200	19,700	24.2	49,500		58	106	164	1060	46,900		506	1015	1521	1280	
	2					49,000		56	112	168	1120	47,300		530	854	1384	1080	
	Avg.					49,200	2.50	57	109	166	1090	47,000	2.38	518	934	1452	1180	
	1	−320	†	†	†	56,100		87	120	207	1195	56,400		789	1185	1974	1500	
	2					55,500		78	120	198	1205	56,100		737	1115	1852	1410	
	Avg.					55,800	†	82	120	202	1200	56,200	†	763	1150	1913	1455	
5183 Welds in 5083-0 plate Semiauto., flat	1	RT	40,600	20,800	10.5	50,000		50	98	148	985	47,900		403	547	950	695	1
	2					50,900		52	80	132	800	49,200		372	701	1073	885	
	Avg.					50,400	2.42	51	89	140	890	48,600	2.34	378	624	1012	790	
	1	−320	53,400	23,300	11.5	59,600		61	84	145	840	57,000		498	732	1230	925	
	2					56,500		50	66	116	665	56,700		487	701	1188	880	
	Avg.					58,000	2.49	56	75	131	750	56,800	2.44	492	716	1209	900	
Semiauto. vertical	1	RT	44,200	20,500	16.5	48,200		37	87	124	875	47,900		396	552	948	695	2
	2					48,200		39	91	130	915	48,800		410	610	1020	770	
	Avg.					48,200	2.35	38	89	127	895	48,400	2.36	403	581	984	735	
	1	−320	59,800	22,700	19.0	49,700		31	80	111	805	57,000		518	723	1241	915	
	2					53,400		37	64	101	645	57,700		522	612	1134	775	
	Avg.					51,600	2.28	34	72	106	725	57,400	2.52	520	668	1188	845	

TABLE 41

Semiauto, horizontal	1	RT	44,600	20,700	16.5	51,000		44	99	143	995	50,500		440	634	1074	800	3
	2					51,100		46	98	144	975	49,400		432	724	1156	915	
	Avg.					51,000	2.46	45	98	143	985	50,000	2.42	436	679	1115	860	
	1	−320	†	†	†	58,000		59	84	143	840	60,000		611	694	1305	880	
	2					60,300		68	82	150	815	59,100		593	641	1234	810	
	Avg.					59,200	†	64	83	147	830	59,600	†	602	668	1270	845	
Auto, flat	1	RT	43,600	19,800	16.5	51,800		57	88	143	880	49,500		469	705	1174	890	2
	2					49,900		49	90	139	895	48,900		454	586	1040	740	
	Avg.					50,800	2.56	53	89	141	890	49,200	2.48	462	646	1107	815	
	1	−320	59,000	21,800	19.0	57,100		56	100	156	1005	56,800		501	618	1119	770	
	2					54,800		50	85	135	855	53,400		517	675	1192	855	
	Avg.					56,000	2.56	53	92	145	930	55,100	2.52	509	646	1156	810	
5183 Welds in 5083-0 extrusions Semiauto., flat	1	RT	43,900	21,900	16.5	48,400		36	95	131	950	49,700		365	604	969	760	2
	2		42,500	20,900	16.0	50,700		40	96	136	980	51,300		418	592	1010	750	
	Avg.					49,600	2.32	38	96	134	975	50,500	2.63	392	598	990	755	
	1	−320	58,500	25,100	18.5	56,100		44	67	111	660	58,100		424	609 685	1032 1109	780 875	4
	2					56,400		43	91	134	905	60,700		460	380 532	840 992	485 655	
	Avg.					56,200	2.24	44	79	123	780	59,400	2.36	442	551	993	695	
HAZ Adjacent to 5183 welds in 5083-0 extrusions Semiauto., flat	1	RT	42,100	20,600	21.9	48,400		39	91	130	910	55,100		744	639	1383	810	
	2		41,400	20,400	17.2	49,400		38	91	129	910	52,800		642	**	**	**	
	Avg.					48,900	2.38	38	91	130	910	54,000	2.64	691	639	1383	810	
	1	−320	†	†		57,600		59	120	179	1200	63,000		808	1364	2172	1725	
	2					59,100		73	114	187	1140	62,100		825	1481	2306	1875	
	Avg.					58,400	†	66	117	183	1170	62,600	†	816	1422	2239	1800	

†† Tear specimens taken with notch in center of weld bead, except for sample designated HAZ, in which notch was located 0.1 in. from square edge of groove weld in heat affected zone.
* 0.2% offset; 2 in. gage length.
† No tensile tests at −320°F; ratio of tear strength to yield strength could not be calculated.
** Specimen tore in direction of loading; energy values were very high, but not meaningful for comparison with other data.
‡ Diagonal crack path.

Notes:
1. Inclusions noted in tensile fractures.
2. Fine porosity noted in tensile fractures.
3. Fine porosity noted in tensile and tear fractures.
4. Load-deformation curves not well defined in one area, so two values of energy are shown; the lowest is a most conservative measure, and the highest is a straight-line estimation. Average values are of all four values in each column.

TABLE 41 cont.

Table 4—Mechanical Properties of 2219-T6E46 Aluminum

Alloy condition[a]	Gage, in.	Temperature, °F	Tensile properties[c] F_{tu}, ksi	F_{ty}, ksi	Elongation, % 0.5 in.	2 in.	Fracture toughness[d], K_{Ic}, ksi $\sqrt{in.}$
Basemetal	0.5	RT	64.3	45.6		18	42.7
Gas tungsten-arc as-welded	0.5	RT	41.0	21.1		8	24.2[e]
							35.2[f]
T6E46 + electron beam	0.5	RT	50.1	37.1	13		37.7[e]
		−320	68.8	44.7	14.5		37.1[f]
T42 + EB + age to-T6E46	0.5	RT	56.6	49.5	10.5		37.8[e]
		−320	73.0	54.5	13.5		
Electron beam + T6E46	0.5	RT	66.1	48.9	16		40.1[e]
		−320	76.6	53.9	15.5		
Groove & fill repair	0.25	RT	38.9	22.6	6		
Re-weld repair	0.25	RT	39.0	22.9	5		
Groove & fill repair[b]	0.50	RT	37.9	21.6	6		
Re-weld repair	0.50	RT	41.8	24.3	6		

[a] T6E46—980° F/65 min, cold water quench; 350° F/12 hr, air cool—electron beam weld (90 ipm, 23.5 kv, 370 ma).
[b] Failed due to lack of side wall fusion.
[c] Averaged results from transverse specimens.
[d] Tested at −320° F.
[e] Weld center flaw.
[f] Fusion line flaw.

TABLE 42

Alloy and temper combination	Thickness, in.	Filler alloy	Temp., °F	Tensile strength, psi	Joint yield strength,* psi	Elong. in 4D, %	Reduction of area, %	Joint efficiency,† %	Location of failure	Notch-tensile strength,‡ psi	Ratios NTS/TS	NTS/JYS
A344-F to A344-F	½	4043	RT	23,800	9,500	12.1	22	100	B	27,500	1.15	2.90
			−112	26,100	10,000	14.3	26	100	B	31,700	1.21	3.17
			−320	33,500	11,500	6.4§	9	89	B	38,100	1.14	3.31
			−452	48,600	18,000	10.0	13	**	A	40,400	0.83	2.24
to 6061-T6	½	4043	RT	24,000	11,400	5.7	23	100	B	29,300	1.22	2.51
			−320	34,700	14,800	7.1	9	92	B	34,100	0.98	2.30
			−452	45,500	28,500	7.1	8	**	B	36,900	0.81	1.29
to 5456-H321	½	5556	RT	24,100	12,200	12.1	27	100	B	29,500	1.22	2.42
			−112	27,000	15,000	5.0	14	100	B	31,100	1.15	2.08
			−320	33,400	16,100	5.7	8	89	C	34,300	1.03	2.13
			−452	37,200	25,400	3.6	7	**	C	36,400	0.98	1.43
354-T62 to 354-T62	½	4043	TR	37,800	21,500	6.4	10	76	A	32,000	0.85	1.48
			−112	40,400	22,900	5.7	11	74	A	33,000	0.82	1.44
			−320	48,900	24,100	5.0	8	84	A	36,900	0.76	1.53
			−452	55,000	38,300	4.3	7	**	A	42,300	0.72	1.10
to 6061-T6	½	4043	RT	30,800	19,000	9.3	39	62	C	28,700	0.93	1.51
			−112	35,800	21,800	7.1	7	66	A	31,500	0.88	1.44
			−320	43,100	23,000	5.0	7	71	A	34,700	0.81	1.51
			−452	45,900	35,700	2.9	4	**	A	37,400	0.82	1.05
to 5456-H321	½	5556	RT	37,700	24,600	3.6	5	75	A	37,700	1.00	1.53
			−112	42,100	27,100	3.6	6	77	A	35,700	0.85	1.32
			−320	47,600	30,400	3.6	5	78	A	39,500	0.83	1.30
			−452	47,700	37,600	2.9	3	**	A	41,300	0.87	1.10
C355-T61 to 6061-T6	½	4043	RT	28,900	19,300	7.1	32	66	C	34,500	1.19	1.79
			−320	44,400	23,300	7.9	19	82	A	38,900	0.88	1.67
			−452	52,300	38,600	6.4	8	**	A	40,400	0.78	1.05
to 5456-H321	½	5556	RT	35,400	24,400	3.6	5	81	A	40,500	1.15	1.66
			−320	45,600	29,300	4.3	7	84	C	45,000	0.99	1.54
			−452	48,300	40,800	2.9	5	**	C	45,500	0.94	1.12

*Offset equals 0.2%; gauge length 2.0-in.
†Based on actual test data.
‡All failures occurred through weld.
§A—through weld.
B—½ to 2½ in. from weld.
C—edge of weld.
§Failed outside of gauge length.
**No parent metal tests for comparison.

TABLE 43

APPENDIX D

Adsit, N.C. and Witzell, W.E., "Fracture Toughness of Aluminum - Boron Composites," paper from Aircraft Structures and Materials Application, Society of Aerospace Material and Process Engineers, Azusa, Calif., 1969, p. 391-398.

Broek, D., "The Effect of the Sheet Thickness on the Fracture Toughness of Cracked Sheet," report, National Aerospace Laboratory, Amsterdam, Jan. 1966, 14 p.

Broek, D., "The Effect of the Sheet Thickness on the Fracture Toughness of Cracked Aluminum Sheet," National Aerospace Laboratory, Amsterdam, 1966, 14 p.

Broek, D., "The Residual Strength of Aluminum Alloy Sheet Specimens Containing Fatigue Crack or Saw Cuts," National Aerospace Laboratory, Amsterdam, March 1966, 8 p.

Brook, R., "Metallurgical and Test Variables," paper from Fracture Toughness, Iron and Steel Institute Publication 121, London, 1969, 67-82.

Campbell, J.E., "Evaluation of Special Metal Properties," Defense Metals Information Center Review of Recent Developments, Metal Properties Council and Defense Metals Information Center, Battelle Memorial Institute, Columbus, Ohio, Jan. 12, 1962, 4 p.

Campbell, J.E., "Low Temperatures of Metals," DMIC Review, MPC and DMIC, Battelle Memorial Institute, Columbus, Ohio, June 12, 1970, 6 p.

Campbell, J.E., "Mechanical Properties of Metals," DMIC Review, Battelle Memorial Institute, Columbus, Ohio, Aug. 5, 1966, 6 p.

Campbell, J.E., "Mechanical Properites of Metals," DMIC Review, Battelle Memorial Institute, Columbus, Ohio, Feb. 16, 1968, 5 p.

Campbell, J.E., "Mechanical Properties of Metals," DMIC Review, Battelle Memorial Institute, Columbus, Ohio, May 10, 1968, 6 p.

Campbell, J.E., "Mechanical Properties of Metals," DMIC Review, Battelle Memorial Institute, Columbus, Ohio, July 19, 1968, 4 p.

Campbell, J.E., "Mechanical Properties of Metals," DMIC Review, Battelle Memorial Institute, Columbus, Ohio, April 25, 1969, 8 p.

Campbell, J.E., "Mechanical Properties of Metals," DMIC Review, Aug. 20, 1969, 6 p.

Campbell, J.E., "Mechanical Properties of Metals," DMIC Review, Battelle Memorial Institute, Columbus, Ohio, Jan. 30, 1970, 6 p.

Campbell, J.E., "Mechanical Properties of Metals," DMIC Review, Battelle Memorial Institute, Columbus, Ohio, May 8, 1970, 6 p.

Campbell, J.E., "Mechanical Properties of Metals," DMIC Review, Battelle Memorial Institute, Columbus, Ohio, July 31, 1970, 5 p.

Campbell, J.E., "Plane-Strain Fracture-Toughness Data for Selected Metals and Alloys," DMIC Report, Battelle Memorial Institute, Columbus, Ohio, June 1969, 26 p.

Carman, Carl M., "Crack-Resistance Properties of High-Strength Aluminum Alloys," U.S. Department of Commerce Clearinghouse Reports, 1965, AD-629 105, 62 p.

Carman, Carl M., Fourney, J.M., and Katlin, J.M., "Plane-Strain Fracture Toughness of 2219-T87 Aluminum Alloy at Room and Cryogenic Temperature," U.S. Department of Commerce Clearinghouse Reports, 1966, AD-639 944, 40 p.

Christian, Jack L., "Problems, Properties and Selection of Structural Metals for Use in Cryogenic Environment," American Society for Metals, Metals Park, Ohio, 1967, 19 p.

Dolowy, J.F. Jr., "Fabrication and Processing Mechanisms Active in Al - B Composites," TMS Paper Selection #569-6, Metallurgical Society AIME, N.Y., 1969, 12 p.

Forsyth, P.J.E., "A Two Stage Process of Fatigue Crack Growth," paper from Crack Propagation Symposium Proceedings, vol. 1, 1961, The College of Aeronautics, Cranfield, England, p. 76-94.

Freed, C.N., "Effect of Side Grooves and Fatigue Crack Length on Plane-Strain Fracture Toughness," Naval Research Laboratory, Washington, D.C., Dec. 7, 1967, 20 p.

Goode, R.J., Huber, R.W., Howe, D.G., Judy, R.W. Jr., and Crooker, T.W., et.al., "Metallurgical Characteristics of High Strength Structural Materials," Naval Research Laboratory, Washington, D.C., Sept. 1967, 162 p.

Grant, N.J., "Intercrystalline Failure at High Temperatures," paper from Fracture, The Technology Press of the Massachusetts Institute of Technology and John Wiley & Sons, Inc., N.Y., 1959, p. 562-578.

Guest, P.J., "Microstructure and Plane Strain Fracture Toughness of High-Strength Aluminum Alloys," Lawrence Radiation Laboratory, California University, Berkley, Oct. 1967, 57 p.

Hanson, M.P., Stickley, G.W., and Richards, H.T., "Sharp-Notch Behavior of Some High-Strength Sheet Aluminum Alloys and Welded Joints at 75, -320, and -423°F.," American Society of Testing Materials, Symposium on Low Temperature Properties of High-Strength Aircraft and

Missile Materials, 1960, 1661, 3-15; discussion, 16-22.

Hickey, C.F. Jr., "Mechanical Properties of Titanium and Aluminum Alloys at Cryogenic Temperatures," ASTM Preprint, #78, 1962, 12 p.

Hunter, M.S. and McMillan, J.C., "Fractography and Microstructure of Aluminum Alloy 7075-T651 and T7351," ASTM paper #47, Race Street, Philadelphia, 20 p.

Hunter, M.S. and McMillan, J.C., "Fractography and Microstructure of Aluminum Alloys 7051-T651 and 7075-T7351," paper from Electron Fractography, ASTM STP, #436, Race Street, Philadelphia, 1968, p. 196-211.

Hyatt, M.V. and Quist, W.E., "The Influence of Heat Treatment and Composition on the Stress Corrosion, Fatigue, and Fracture Properties of 7075 Type Aluminum Alloys," paper from Proceedings of the Air Force Materials Laboratory Conference on Corrosion of Military and Aerospace Equipment, Nov. 1967, p. 827-877.

Irwin, G.C., "Effects of Size and Shape on Fracture of Solids," paper from Properties of Crystalline Solids, ASTM STP #238, American Society for Testing Materials, Philadelphia, 1961, p. 118-128.

Jackson, L.C., McClure, G.M., and Kasuba, J.A., "Some Design Aspects of Fracture in Flat-Sheet Specimens and Cylindrical Pressure Vessels," DMIC,Battelle Memorial

Institute Memo #174, Aug. 9, 1963, 24 p.

Judy, R.W. Jr., Goode, R.J., and Freed, C.N., "Fracture Toughness Characterization Procedures and Interpretations to Fracture-Safe Design for Structural Aluminum Alloys," Naval Research Laboratory, Washington, D.C., March 31, 1969, 29 p.

Kaufman, J.G. and Holt, M., "Fracture Characteristics of Aluminum Alloys," Technical Paper No. 18, 1965, Aluminum Company of America, Pittsburgh, Pa., 48 p.

Kaufman, J.G., "Fracture Characteristics of Aluminum Alloy Plate in Tension Tests of Large Center Slotted Panels," ASTM paper no. 49, ND, Race Street, Philadelphia, 33 p.

Kovacs, W.J. and Low Jr., J.R., Intergranular Fracture in an Al - 15 Wt.% Zn Alloy, Carnegie Mellon University, Pittsburgh, Pa., Feb. 1970, 239 p.

Kuhn, P. and Figge, E.I., "Unified Notch Strength Analysis for Wrought Aluminum Alloys," National Aeronautics and Space Administration Technical Note 1259, May 1962, 76 p.

Newman Jr., J.C., "Fracture of Cracked Plates under Plane Stress," National Aeronautics and Space Administration, Langley Research Center, Langley Station, Va., 1967, 34 p.

Paris, P. and Erdogan, F., "A Critical Analysis of Crack Propagation Laws," American Society of Mechanical Engineers Paper no. 62-WA-234, 1962, 7 p.

Paris, P., "Crack Propagation Caused by Fluctuating Loads," American Society of Mechanical Engineers, Paper no. 62-MET-3, 1962, 8 p.

Porter, D.E., "Fracture Toughness of Precipitation Hardening Aluminum Alloys," Lawrence Radiation Lab., Calif. Univ., Berkeley, Apr. 1969, 135 p.

Ryder, D.A. and Smale, A.C.,"A Metallographic Study of Tensile Fractures in Aluminum Copper and Aluminum - Copper - Zinc - Magnesium Alloys,"Fracture of Solids, 1962, 1963, p. 237-259.

Schwartzberg, F.R., Keys, R.D., and Kiefer, T.F., Cryogenic Alloy Screening, Martin Marietta Corp., Denver, Colo., Feb. 1970, 233 p.

Schwartzberg, F.R., Keys, R.D., and Kiefer, T.F., "Fracture Behavior of Two New High-Strength Aluminum Alloys," WESTEC Technical Paper no. W9-19.1, American Society for Metals, Metals Park, Ohio, 1969, 13 p.

Crack Propagation Symposium,Proceedings, 2 vol., The College of Aeronautics, Cranfield, England, 1961, 566 p.

"Design Properties as Affected by Cryogenic Temperatures," Defense Metals Information Center, Battelle Memorial Institute, DMIC Memo 81, Jan. 24, 1961, 18 p.

"Material Selection for Crack Resistance," paper from Crack Propagation Symposium,Proceedings, vol. 2, The College of Aeronautics, Cranfield, England, 1961, p. 442-466.

"Mechanical Property Data 7178 Aluminum T76 Sheet," Battelle Memorial Institute, Columbus, Ohio, 1969, 8 p.

"Recent Research on the Properties of Aluminum Alloys," Mechanical World and Engineering Record, vol. 141, Dec. 1961, p. 413-415.

"A Review and Comparison of Alloys for Future Solid Propellant Rocket-Motor Cases," Defense Metals Information Center, Battelle Memorial Institute, Memo no. 184, 1963, 40 p.

APPENDIX E

Foreign References

French

Develay, R., "Medium and High Strength Aluminum Alloys and Their Behavior at Low Temperature," _Revue de l'Aluminium_, no. 339, Feb. 1966, p. 193-216.

Gadeau, R., "Cryogenic Al Alloys," _Revue de l'Aluminium_, no. 304, Dec. 1962, p. 1355-1369.

Gobin, P. and Montuelle, J., "Brittleness of Zinc-Rich Al - Zn and Al - Zn - Mg Alloys," _Memoires Scientifiques de la Revue de Metallurgie_, vol. 56, Nov. 1959, p. 617-624.

Mercier, J., "Potentialities of AG4-Type [Aluminium] Alloys at Very Low Temperatures," _Revue de l'Aluminium_, 1962, (295), p. 193-198.

German

Barry, W.G. and Young, G.M., "Extrusion Alloy Alcan 505 HO," _Aluminium_, vol. 38, May 1962, p. 299-303.

Brennan, P., "Change of Mechanical Properties of Cold Rolled Al Mg Sheets Stored at Room Temperature," _Aluminium_, vol. 36, Oct. 1960, p. 589-594.

Dubsky, J., "Effect of Warm Aging on the Mechanical and Electrical Properties of Butt Cold Pressure Welded Copper Aluminum Alloys," _Elektrie_, vol. 18, no. 10, Oct. 1964, p. 73-75.

Honsel, H. and Zimmermann, P., "Influence of Gas Content on Cast Aluminum Alloys," Aluminium, vol. 36, March 1960, p. 130-137.

Kaufman, J.G. and Holt, M., "Fracture Behavior of Aluminum Alloys," Aluminium, Jan. 1970, 46, (1), p. 103-115.

Kochendorfer, A., "Strength and Deformation Properties of Metals at Low Temperatures," Zeitschrift fur Metallkunde, vol. 51, Feb. 1960, p. 73-80.

Scharf, G., "The Influence of the Composition on the Extrudability and the Properties of Al Mg Si O Alloys," Zeitschrift fur Metallkunde, vol. 55, no. 12, Dec. 1964, p. 740-744.

Zinkham, R.E., Dedrick, J.H., and Jackson, J.H., "Fracture Toughness Testing of Al - Zn - Mg - Cu Alloys," Proceedings, 5th International Light Metal Conference, Leoben, 1968, Dusseldorf, Germany, 1969, p. 129-140.

Italian

Mori, L., "The Effect of Temperature on the Impact Strength of Some [Aluminium] Alloys," Alluminio, 1962, 31, (6), p. 295-303.

Japanese

Tukui, T., "Notch Toughness and Fracture Characteristics of Structural Al - Zn - Mg Alloy Welds," Journal of the Japanese Institute of Light Metals, May 1970, 20, (5),

p. 222-233.

Takeuchi, K. and Tanaka, E., "Mechanical Properties of Aluminum Alloys at Low Temperatures," Sumitomo Light Metal Technical Report, 1961, 2, (3), p. 78-81.

Russian

Panin, A.V., "The Possibility of Brittle Fracture of Aluminum Alloy Welded Joints," Avtomat Svarka, no. 1, Jan. 1967, p. 38-41.

Vorobev, G.M., Goldstein, R.M., and Mauritz, I.I., "Effect of Alloys on Basic Mechanical Properties of Silumin," Tsvetney Met, no. 3, 1965, p. 83-86.