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3.6 Development of Al–Zn–Mg–Cu alloys for aircraft

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[Plate 23]

Historical development of the high strength Al–Zn–Mg–Cu alloys during the last 50 years reflects the influence of progressive changes in engineering and functional requirements related to the evolution of aircraft size, design concepts and manufacturing methods with consequent changes in dimensions and configurations of structure components. If the only criterion of merit for such application were high strength uncomplicated by stress-concentrating geometric factors or environmental influences, compositions available at least 40 years ago would have been highly suitable, since strength of the early materials fully equalled that of the alloys in use today.

These early alloys were, however, not adequately resistant to stress-corrosion cracking even in the form of thin-section products such as sheet, thin extrusions and tube prevalent in aircraft of the earlier portion of the period. Development of the chromium-containing alloy 7075–T6 in the 1940s provided a solution to this problem for the thin-section products. As experience accumulated in extending the use of 7075–T6 to larger and thicker components produced from plate, thick-section extrusions or forgings, transverse direction stress-corrosion cracking was encountered along with decreasing strength and an undesired degree of mechanical anisotropy, particularly ductility. Alloy 7079, which had more uniform mechanical properties, enjoyed a transient period of popularity but proved to be even more deficient in resistance to stress-corrosion. The latter problem was eliminated with development of 7075–T73 involving use of a two-stage precipitation treatment with an accompanying significant strength penalty. Despite the strength trade-off, this material is extensively used in medium section thickness components of current large American aircraft and has an eminent success record.

More highly alloyed compositions of this alloy family, 7178 and 7001, capable of developing higher strength than 7075–T6, appeared in the 1950s. The former is used in compressively stressed portions of wing structure but is not highly favoured for newer designs because of somewhat lower fracture toughness. The latter was considered deficient in this attribute and consequently not adopted for aircraft usage.

Further developments based on 7075 include the T76 temper with high resistance to exfoliation corrosion and mechanical properties intermediate to those of the T6 and T73 temper products. Alclad 7075 employing a high strength cladding is used in certain large aircraft. More closely controlled variants of 7075, i.e. 7175 and 7475, coupled with special fabricating practices developed within the last 10 years are characterized by higher fracture toughness leading to increasing application. The newer alloys 7049 and 7050, introduced within the last 5 years, provide further advance in combinations of mechanical properties and environmental resistance resulting in adoption for retrofit programmes and inclusion in newly designed aircraft.

Development and successful application of these materials in today's aircraft is a result of the persistent and systematic efforts of many scientists and engineers employing practical experimental and analytical methodology consistent with the Rosenhain tradition. The good record of reliability which these vehicles and materials have accumulated, and may confidently be expected to continue to accumulate, is a tribute to these people and their talents and dedication.

The invitation to participate in the 1975 Rosenhain Memorial Conference is appreciated greatly by your speaker and the company he represents. The Conference planning committee suggested a presentation on high strength aluminum-base alloys for aircraft structures application with emphasis on those of the Al-Zn-Mg-Cu type, in particular the American alloys and viewpoint. The appropriateness of the subject derives from the involvement of Dr Rosenhain and the National Physical Laboratory in the early and continuing development of aluminum alloys for aircraft. The general concurrence of the birth and growth of the aluminum industry with the age of powered flight has been noted many times. In the maturing period of these industries, over the last 30 years, development and application of the high strength Al-Zn-Mg-Cu alloys has played an important role in making possible the increased size and high reliability of today's commercial aircraft.

Alloys of the Al-Cu-Mg type, with lineage from the 1906 duralumin and age hardening discoveries of Wilm, continue to play a prominent part in not only supersonic aircraft, where mechanical properties at elevated temperature are the basis for their selection, but also in components of subsonic commercial aircraft of latest design. Alloy and temper selection, whether naturally aged or precipitation heat treated, depends upon service requirements, whether emphasis is on fracture toughness or on higher strength and stability at elevated operating temperatures. The well-known 4 Cu, 2 Ni, 1.5 Mg 'Y' alloy (2218)[†], developed at the N.P.L. under Rosenhain, as well as the more readily fabricated variant 2018 have been used for many years principally in engine applications and can be regarded as precursors of RR58 (2618), which was selected for major Concorde structural applications.

While lineage of the currently-used Al-Zn-Mg-Cu alloys cannot be traced with continuity to the N.P.L. activity under Rosenhain's direction, the information published in 1921 (Rosenhain, Archbutt & Hanson 1921) undoubtedly influenced and stimulated the later developments. As recounted in this publication, Al-Zn-Mg and Al-Zn-Cu-Mg alloys were made as early as 1911 in the N.P.L. programmes as an extension of work with Al-Zn-Cu. The addition of Mg greatly increased fabricating difficulties but by 1917, compositions such as alloy 'E' (20 Zn, 2.5 Cu, 0.5 Mg, 0.5 Mn) were successfully extruded and rolled after extrusion. Ultimate tensile strength values ranging to 580 MPa were reported for this material naturally aged after water quenching. This compared with about 420 MPa for duralumin. Evaluation of alloy 'E' and the other experimental alloys included a broad scope of mechanical tests, including notch-impact and fatigue as well as corrosion and stress-corrosion tests. Results of these tests raised questions concerning suitability of the stronger alloys for structural use. The observations concerning compositional effects on metallurgical structures, extent of recrystallization, grain shape, etc., and their correlation with properties were astute and consistent with current views.

It would not be appropriate to recount the subsequent developments without recognizing the influence of Rosenhain's many publications advocating observance of a scientifically based, physical metallurgy approach in alloy development and of those which exemplify his keen observations and interpretations of relations between metallurgical structures and the functional properties of alloys. Specifically, his insistence on the necessity for accurate information on equilibrium relations, his concern with the nature of grain boundaries and plastic deformation as well as with mechanisms of corrosion and stress-corrosion cracking have had considerable

[†] The four-digit numbers cited throughout this paper are those of the current Aluminum Association designation system adopted in 1952.

impact on subsequent research and on development of many aluminum alloys. His later interest in gas removal, impurity effects, grain refinement and ingot structures has influenced both their compositions and processing.

Earliest references generally cited as direct ancestry for the modern 7XXX alloys are those resulting from work of the German investigators Sander and Guertler, who demonstrated the exceptionally high tensile properties possible with fully heat treated ternary Al–Zn–Mg alloys based on the Al–MgZn₂ section (Sander 1926). Their work was followed up in Germany, and concurrent development was proceeding in Great Britain, France and Japan during the 1930–40 decade. Space limitations do not permit tracing these developments, and they had only modest direct influence on the American activity, which is the primary focus of this review.

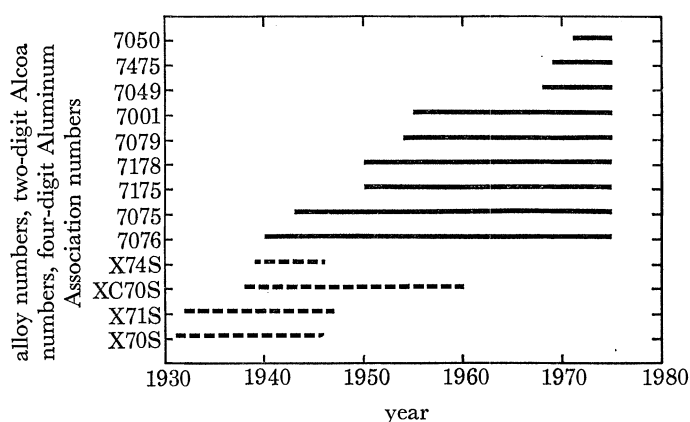


FIGURE 1. Chronology of American Al–Zn–Mg–Cu alloys. Arranged by year number assigned and showing period of listing.

1930–1940: early American Al–Zn–Mg–Cu alloys

Development activity on Al–Zn–Mg–Cu alloys, not on an urgent or intensive basis, was undertaken in the late 1920s by Alcoa's laboratories. Following preliminary production and testing in the laboratories and trials in the plants during 1930–31, two compositions were given the first of the two-digit Alcoa numbers for wrought alloys starting with 7, which denoted zinc as the alloying element present in greatest mass percentage. The first two alloys (table 1*a*, figure 1) were identified by the designations X70S and X71S, the 'X' indicating in both cases the experimental status. The former was proposed for producing forgings and had moderately high strength. For forging, emphasis was to achieve ease of hot working because of then-existing plant equipment limitations. Development activity continued with numerous variants of X70S, including higher strength compositions, such as XC70S, which were predecessors of 76S (now 7076), and more forgeable, less quench-sensitive alloys such as X73S, which was used for a short time to produce radial aircraft engine crankcases.

X71S, intended for producing sheet and extrusions, had higher alloy content and considerably higher strength (U.S. patent 1924729, 1933). Alloys of this series, of which only two are listed in table 1*a*, were considered for construction of dirigibles for the U.S. Navy, but their relative susceptibility to stress-corrosion cracking (s.c.c.), the high ratios of yield strength to tensile strength, lower elongations than for naturally-aged Al–Cu–Mg alloys, poor forming characteristics, and difficulty of fabrication were deterrents to acceptance. If high static strength were the only criterion of engineering value, it is evident that X71S, or alloys based

on the ternary Al–Zn–Mg system that readily attain equal or higher strengths, would have been accepted and used much earlier than was actually the case.

In 1937 information was published on equilibrium phase relations in the ternary Al–Zn–Mg system with Al solid solution limits from 200 to 400 °C (Fink & Willey 1937). Extension of this work to the quaternary system was not undertaken for several years, but availability of the ternary data provided a basis for some fresh approaches.

TABLE 1. AMERICAN Al–Zn–Mg–Cu ALLOYS
(a) PARTIAL LIST OF EARLY ALCOA EXPERIMENTAL ALLOYS

alloy number	year listed	nominal composition, %				typical tensile properties†			product
		Zn	Mg	Cu	Mn	u.t.s.	t.y.s.	el.,	
						MPa	MPa	% in 50 mm	
X70S	1931	10.0	0.4	1.0	0.7	400	310	18	forgings
XC70S‡	1938	8.0	1.1	0.5	0.8	480	420	17	forgings
X71S	1932	10.0	2.0	2.0	1.0	585	560	10	sheet
XB71S‡	1932	9.0	2.7	2.0	1.0	595	565	10	sheet
X73S	1935	5.2	0.9	0.5	—	370	290	20	forgings
X74S	1939	5.2	2.1	1.5	0.4	510	440	12	sheet

(b) COMMERCIAL AMERICAN ALLOYS

alloy number	year listed	nominal composition – mass %						typical tensile properties¹			products²
		Zn	Mg	Cu	Mn	Cr	Zr	u.t.s.	t.y.s.	el.	
								MPa	MPa	% in 50 mm	
7076³	1940	7.5	1.6	0.7	0.6	—	—	530	490	14	F³
7075⁴	1943	5.6	2.5	1.6	—	0.25	—	570	500	11	S, P, F, E
7175⁵	1950	5.6	2.5	1.6	—	0.25	—	620	560	11	F
7178	1950	6.8	2.7	2.0	—	0.25	—	600	540	10	S, P, E
7079	1954	4.3	3.3	0.6	0.2	0.2	—	540	470	14	S, P, F, E
7001	1955	7.4	3.0	2.1	—	0.25	—	675	625	9	E
7049⁶	1968	7.7	2.5	1.6	—	0.16	—	540	470	11	F, E
7475⁴	1969	5.7	2.3	1.6	—	0.22	—	560	490	12	S, P, F
7050⁷	1971	6.2	2.3	2.25	—	—	0.12	530	460	11	P, F, E

† Different heat treating conditions were employed for different compositions and products; some alloys not heat treated to maximum strength temper.

‡ Letter prefixes used to distinguish related compositions established sequentially. These would not qualify as modifications of the original under current Aluminum Association rules.

1, Properties for forged or rolled products. Extruded products have higher strengths. 2, S-sheet, P-plate, F-forgings, E-extrusions. 3, Propeller blade forgings only. 4, See text concerning T73-type and T76 tempers. 5, 'Premium Strength' forgings. 6, Properties listed for T73-type temper. 7, Properties listed for T736-type temper.

The alloy development programmes were spurred to some degree by the appearance of RR77 early in 1937 and by emerging military needs for higher strength materials. Considerable emphasis was placed on resistance to s.c.c., and many hundreds of tests were performed on a large number of compositions fabricated in the laboratories. By mid-1938 a composition had been selected for sheet that was considered to have a combination of the major solute elements providing the most favourable balance of strength, resistance to s.c.c. and other characteristics including fabricating properties. This Al–Zn–Mg–Cu–Mn alloy was designated X74S, and a programme was initiated to produce sheet in the production mill for evaluation by potential users.

The account by Dix (1950) discloses the disappointment encountered in trials of X74S parts which had been subjected to plastic deformation and fit-up stresses. Under these conditions early s.c.c. was encountered, and this effectively sent the development 'back to the laboratory' for several years.

1940-1950: alloys 7076 and 7075

Introduced in 1940, alloy 7076 (table 1*b*) is the oldest continuously-produced American alloy of the 7XXX type, figure 1. The composition was selected to produce forged aircraft propeller blades meeting urgent military needs for a more abrasion- and fatigue-resistant material than then-standard 2025. The amounts produced have never been great because of its specialized application, but this alloy is still produced today.

After the late-1939 observations of s.c.c. in formed sheet components, X74S was eliminated from further consideration and laboratory work intensified to develop more predictive s.c.c. testing procedures and to improve the alloy performance. Among many other compositional and heat treating factors evaluated was addition of small amounts of several sparsely-soluble, high-melting elements, Zr, Cr, V, Mo, W. These were added individually in amounts of 0.15 to 0.25 % to X74S, which in all cases contained 0.4 % Mn. As results from boiling 6 % NaCl solution tests of deformed sheet specimens of these alloys were completed, it was apparent that outstandingly better s.c.c. performance was observed with the Cr-containing composition. Immediately a series of compositions was prepared with 0.05 to 0.45 % Cr, and from these it was concluded that 'the presence of 0.2 to 0.35 % Cr definitely prevents cracking' (U.S. patent 2240940, 1941).

Some time was required to gain further assurance by confirming and repeated tests, to establish that Mn could be omitted when Cr was present, and to ascertain that d.c. ingots could be produced and fabricated in the mill. By March 1941 the first sheet conforming to a low alloy content within the 7075 specification range had been made in the mill and tested. This composition, corresponding to X74S but with Cr in place of Mn, was identified as XA75S. When higher strength was requested, the Zn, Mg and Cu contents were all adjusted upward. This composition was initially designated XB75S. By early 1943 all laboratory work with XB75S was highly favourable, and it was then established as 75S (now 7075).

Broad evaluation programmes were conducted by the aircraft industry, some involving substitution of the higher strength alloy in existing designs, others involving new designs, all at this time for military purposes. The alloy was developed first as a sheet material and soon found suitable for extrusions. First production of hand forgings was in mid-1945, and by early 1946 thick (to 90 mm) plate had been rolled from 200 mm × 1000 mm d.c. ingots and tested. The adverse effect of the Cr addition on development of strength with low quenching rates was noted early. This tended to limit application of the alloy to products having section thicknesses of 75 mm or less. Early production, fabrication and forming problems were solved, and industry acceptance resulted in rapid growth of production so that by 1950 this alloy accounted for 22 % of Alcoa's shipments of strong, heat treatable aircraft products (Nock 1953).

Among the alloys of table 1*b*, 7075 has been produced and used in the greatest quantity by a very wide margin, and high mechanical reliability has been experienced in modern aircraft employing this basic composition in a variety of structural components and fabricated forms. The engineering trade-offs considered in establishing the proportions of the major solute elements, Zn, Mg and Cu were those of continuing importance today, and this basic composition has been found particularly adaptable to new temper and compositional controls to meet

more stringent requirements for fracture toughness and resistance to exfoliation and s.c.c. For about 15 years from its original introduction, the only tempers applied were those of the T6 type. Resistance to s.c.c. in moderately thick products in the T6 temper was inadequate, but the alloy was found to have highly favourable response to the higher temperature T7X type precipitation treatments employed later to greatly increase resistance to s.c.c. and exfoliation corrosion. Increasing base metal purity coupled with special processing provided substantial gains in fracture toughness. This basic alloy and the variants of composition and temper developed since its introduction are expected to retain a continuing role in aircraft applications for some time.

Although only incidental in relation to the development story, it seems opportune to report in connexion with this conference† some previously unpublished information from Alcoa Laboratories on the properties and structure of 7075-W (as-quenched, naturally aged) sheet that was tested and examined after 25 years natural ageing. The changes in tensile properties of this early production lot were monitored over the 25 year period, figure 2, and additional changes are apparently continuing at a very low rate at the present time.

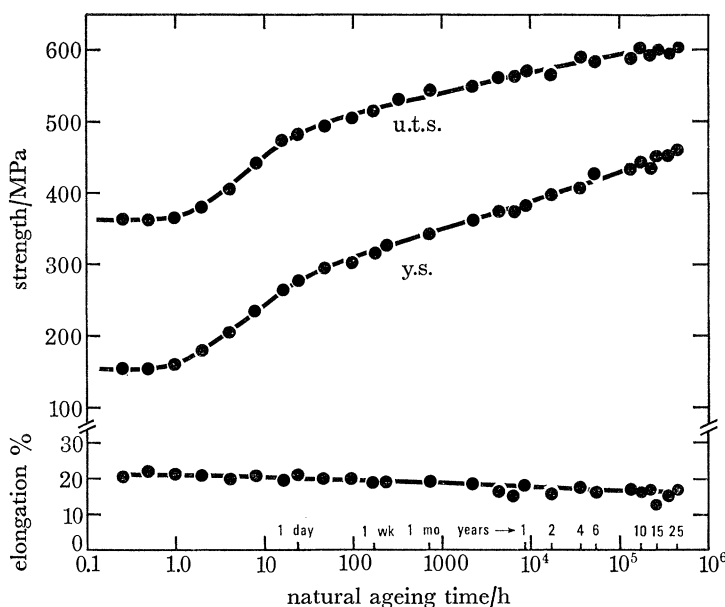


FIGURE 2. Tensile properties transverse to rolling direction of 1.6 mm thick 7075-W sheet as a function of natural (room temperature) ageing time. Solution heat treated at 466 °C, cold water quenched, flattened by stretching, and naturally aged.

The transmission electron microstructure (figure 3, plate 23) of the material examined after 25 years gave unmistakable evidence of the quasi-spherical g.p. zones normally observed in the precipitation heat treated T6 temper. Zones formed during the stages of natural ageing in which the greatest strength changes occur have not been resolved by t.e.m. nor are those formed in long-time naturally aged 2XXX alloys. Estimated mean zone diameter in the W (25 years) material was 1.2 nm and approximate range < 1.0 to about 1.5 nm. The estimated zone density was $4 \times 10^{18}/\text{cm}^3$. A brief experiment showed that a zone structure of similar apparent size and

† The 1921 Eleventh Report to A.R.C. (Rosenhain *et al.* 1921) contains comparative short time and 10-year natural ageing data for some German-produced duralumin.

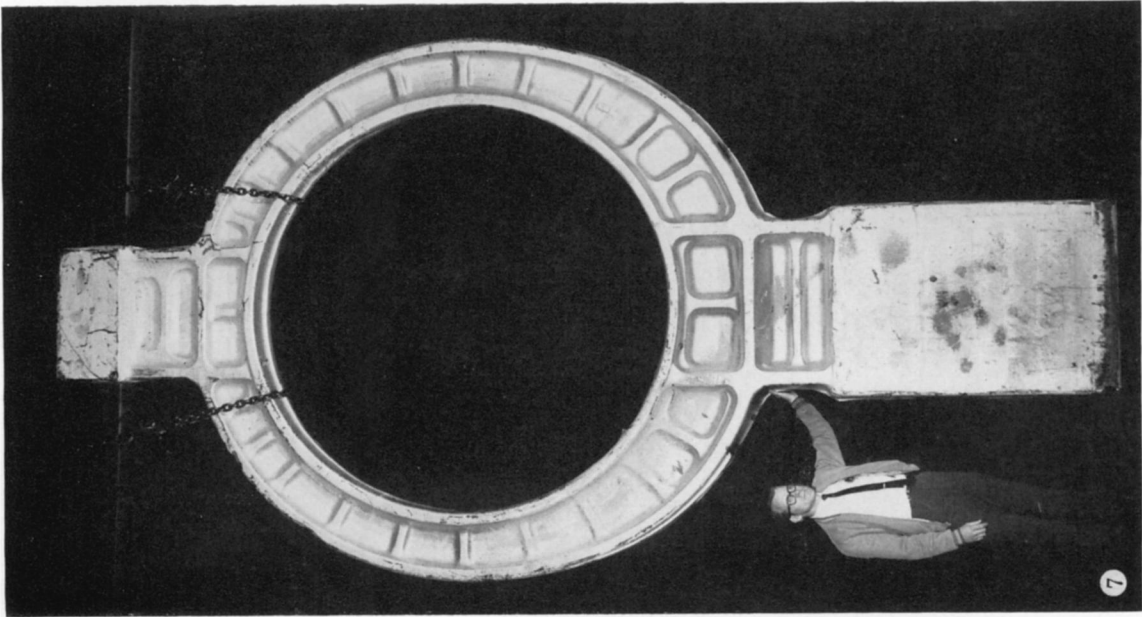


FIGURE 7. Die-forged 7075-T73 integral centre engine support and vertical stabilizer spar for McDonnell-Douglas DC-10. Four similar forgings are used in each plane.

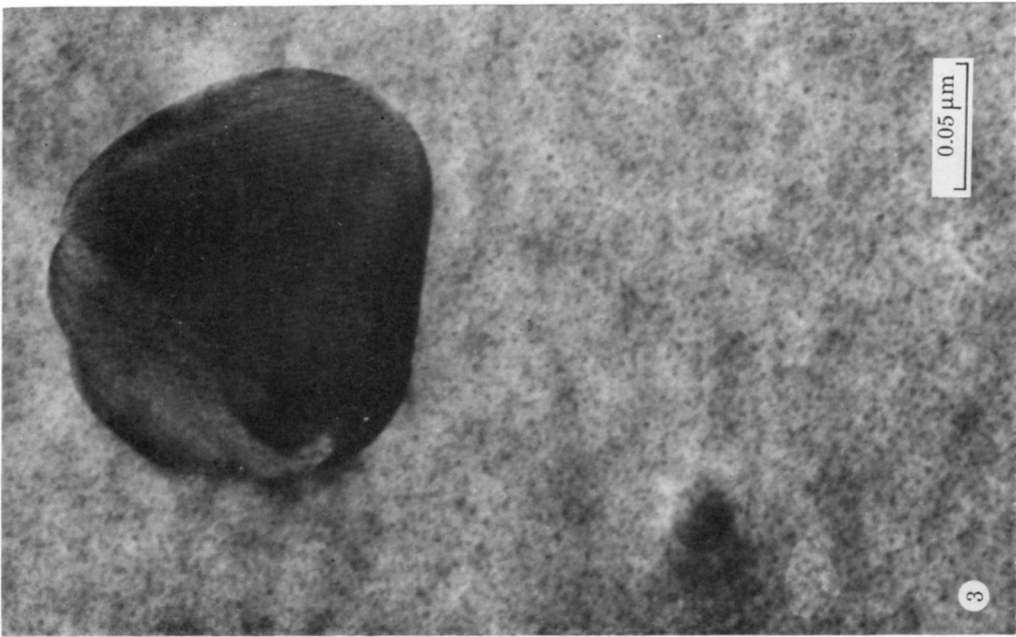


FIGURE 3. Transmission electron microstructure of 7075-W sheet after 30 years natural (room temperature) ageing. Estimated mean size of g.p. zones about 1.5 nm. Large dark area is $Al_{12}Mg_2Cr$ precipitate.

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density was produced by a precipitation treatment on freshly quenched material of 20 min at 120 °C, but the yield strength of the material so treated was only 265 MPa compared with the 25 year r.t. material's 460 MPa.

1950–1960: 7178, 7001, 7175, 7079, 7075–T73

7178. In response to insistent industry and military demand for an alloy with higher strength than 7075–T6 (target yield strength $\sim 10\%$ higher), the 7178 composition was introduced in 1950 (see Nock 1953). This alloy has been used on a selective basis principally for compressively stressed members of both military† and commercial‡ planes but has fallen into general disfavour for newer designs as consideration for damage tolerance and fracture toughness assumed greater prominence.

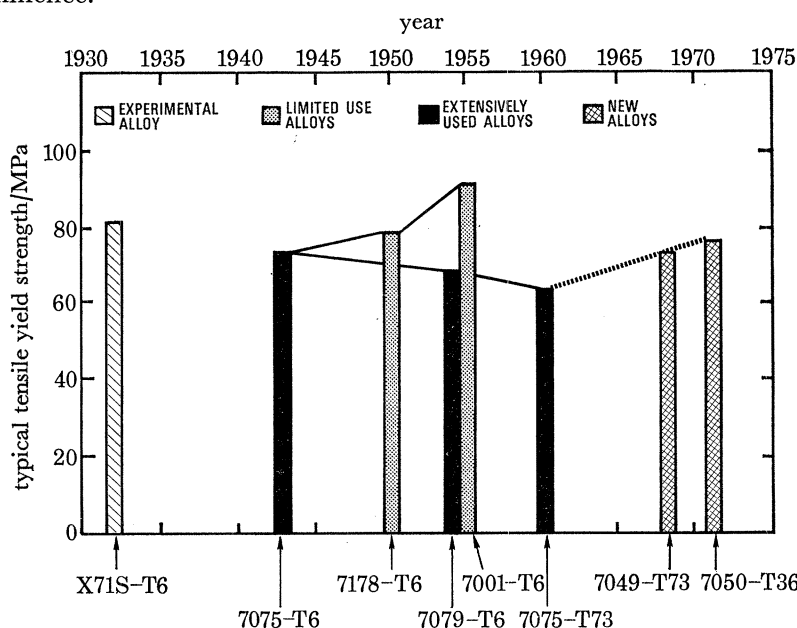


FIGURE 4. Typical tensile yield strengths (0.2 p.s.) of American Al-Zn-Mg-Cu alloys related to year of introduction.

7001. In 1955 the 7001 composition was introduced providing another step on the ladder of increasing static strength capability (Harvey Aluminum 1955). The even more limited ductility and toughness of this alloy in T6 type tempers were considered inadequate to permit its use in major aircraft structure components. The T75 temper developed later offered significantly improved resistance to s.c.c. and exfoliation corrosion. However, 7001–T75 has also not been accepted, mainly because of its fracture toughness characteristics, which are intermediate to those of 7075–T6 and 7178–T6 (ATC 1966).

As evidenced in figure 4, following the appearance of these two higher strength alloys, primary effort was directed towards improving properties other than static strength in the absence of stress-concentrating features. Later alloys are characterized by lower yield and ultimate strengths, representing trade-off of strength to upgrade tolerance for environmental, geometric and manufacturing factors, fatigue or other damage in service.

† Examples: McDonnell-Douglas F4 – alclad fuselage sheet (originally T6, now T76); Boeing B52 extruded wing stiffeners (upper surfaces) and original production upper wing skins; Lockheed C130 main wing plates.

‡ Examples: Lockheed Electra main wing plates; Boeing 727 upper wing skins.

7175. Concern with the degree of mechanical anisotropy of moderately large and thick-section 7075-T6 products in the late 1940s led to establishment in 1950 of the 7175 composition modification. This differed from 7075 only in specifying lower maximum limits on the impurity elements Fe and Si coupled with slight adjustments in maximum Mn and Cr contents in recognition of beneficial ductility effects from reduced volume fraction of undissolved phases. The 7175 specification was applied generally in producing parts of large cross section to meet more stringent elongation requirements in directions perpendicular to that of major extension in working.

With even tighter composition controls and special fabricating procedures 7175 is used to produce Alcoa's 'Premium Strength' forgings, which originated considerably later. These products, available today in two tempers, T66 and T736, provide superior fracture toughness at yield strength levels 10 to 15 % higher than those of comparable 7075 forgings in equivalent tempers (Kauffman, Schilling & Nelson 1969).

7079. With installation of much larger and higher capacity forging and extrusion presses (1952-4), fabrication of much larger shaped components became possible. The mechanical properties of very thick 7075-T6 and 7175-T6 products – forged, rolled or extruded – were lower than was desired because of the high sensitivity of this basic composition to rate of cooling in quenching from the solution heat treatment. Developmental programmes conducted in both laboratories and plants led to selection of alloy 7079† in 1954 for producing the larger parts. This alloy developed strengths higher than the 7075 types and about equivalent to those of the previously most popular forging alloy, 2014-T6, in sections of 100 mm or greater thickness. Its ductility, as measured by percent elongation and reduction in area and by notched specimen properties, was superior to that of the previous alloys, 2014-T6 and 7075-T6, and directionality of these properties was much less pronounced.

Resistance to s.c.c. of 7079-T6, based on accelerated, presumably predictive tests and natural environment tests of duration accumulated up to the time of its commercial introduction, was indicated to be not inferior to 2014-T6 or 7075-T6, which had not offered serious problems in service. Accordingly, this alloy, which also had attractive fabricating characteristics, appeared to meet the requirements considered important at that time exceptionally well. Through the latter 1950s and into the early 1960s 7079 was specified and produced in high volume for a variety of new and older aircraft designs in a wide range of product forms for primary structure elements as well as some skin sheet (which exhibited higher fracture toughness than equivalent 7075-T6 sheet).

By the latter part of the 1950s, it became apparent that this alloy was susceptible to s.c.c. under environmental conditions of service, and a disturbing frequency of cracking was encountered with products of thick and non-uniform section such as die forgings, some extrusions and parts produced by programme-machining thick plate. This was undoubtedly influenced by part size and configuration and in some cases by new service parameters applicable to such parts as hydraulic landing gear cylinders. The greater susceptibility relative to the previously-used alloys had not been indicated by laboratory tests. This apparent divergence between 'accelerated' tests and service experience was found to be associated with alloys containing either no Cu or less than about 1 % of this element, whereas good correlation was found with alloys having Cu contents equal that of 7075 (1.6 % Cu) or higher (Lifka & Sprowls 1967).

† Alloy 7079 is essentially the same as the German alloy identified originally as Hy43 (ISO type AlZn4Mg3Cu; DIN 3.4345).

As a consequence of the service experience with 7079-T6 and inadequacy of 7075-T6 in large, thick-section products, work previously underway was intensified to find a basic materials solution to the s.c.c. problem.

7075-T73. Highest strengths are developed in alloys of the type under discussion by precipitation heat treatment at temperatures lower than 150 °C, a conventional T6 treatment being 24 h at 120 °C. Alternate two-stage treatments, such as 4 h at 96 °C followed by 8 h at 157 °C, were developed and used starting in the 1940s to reduce the total cycle time in producing 7075-T6 sheet (Nock & Wooll 1947).

A broad experimental programme was undertaken with 7075 in which the effects on resistance to s.c.c. of many variants of isothermal and two-stage ageing were evaluated. This work culminated in development of 7075-T73 products, which were introduced commercially in 1960. It was established that precipitation treatment at temperatures higher than 160 °C progressively improved s.c.c. resistance. Higher strength (figure 5) was obtained when the treatment at these temperatures was preceded by a lower temperature stage to nucleate a finer, higher-density precipitate dispersion, and the resulting combinations of strength and s.c.c. resistance were superior. The T73 temper is therefore produced by applying a two-stage precipitation treatment.

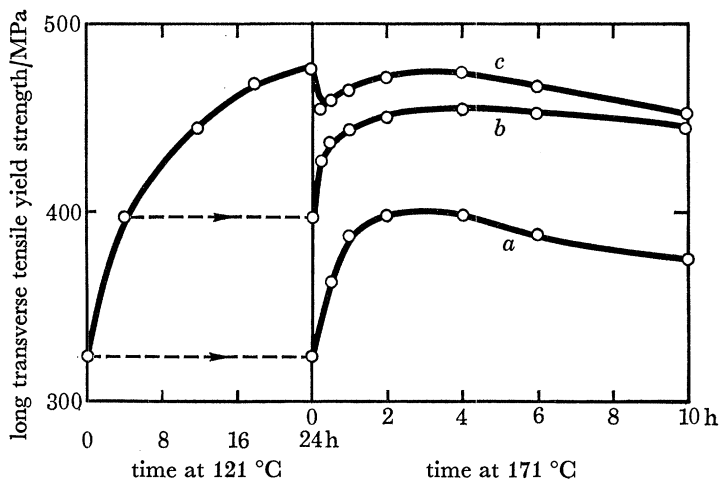


FIGURE 5. Comparison of yield strengths (0.2 p.s.) of 7075 plate resulting from *a* isothermal (171 °C) and *b*, *c* two-stage (121/171 °C) precipitation heat treatments.

Based on many hundreds of accelerated (chloride solutions) and natural environment tests employing smooth surface specimens stressed in the short-transverse direction either in a bending mode or in direct tension, properly processed 7075-T73 material resists s.c.c. at tension stresses of at least 300 MPa. By comparison, the same materials in T6 temper would not be expected to resist s.c.c. consistently in the accelerated tests at stresses over about 50 MPa (Brown *et al.* 1972). Although most of the accumulated information involved in alloy development and proof testing has been obtained from smooth specimen tests, there has been recent interest in measurements of stress-corrosion crack velocity as a function of the crack tip stress intensity. Such data have been obtained from fatigue-precracked double cantilever beam specimens, and comparative values for short transverse direction tests of several of the alloys and tempers discussed here are illustrated in figure 6 (Speidel & Hyatt 1972; Hyatt 1970; Davies, Nordmark & Walsh 1974).

The 7075-T73 development is considered a highly significant contribution to the technology of the Al-Zn-Mg-Cu alloys, enhancing their service reliability and engineering value in the form of large aerospace structural components. Several million kilograms of products in this alloy/temper have now been in service for several years with no reported s.c.c., and it is anticipated that there will be continued large volume usage of this material. One of the largest and most interesting die forgings currently being produced in 7075-T73 is the integral engine mount and vertical stabilizer spar for the McDonnell-Douglas DC-10†, figure 7, plate 23 (Turley & Gassner 1972).

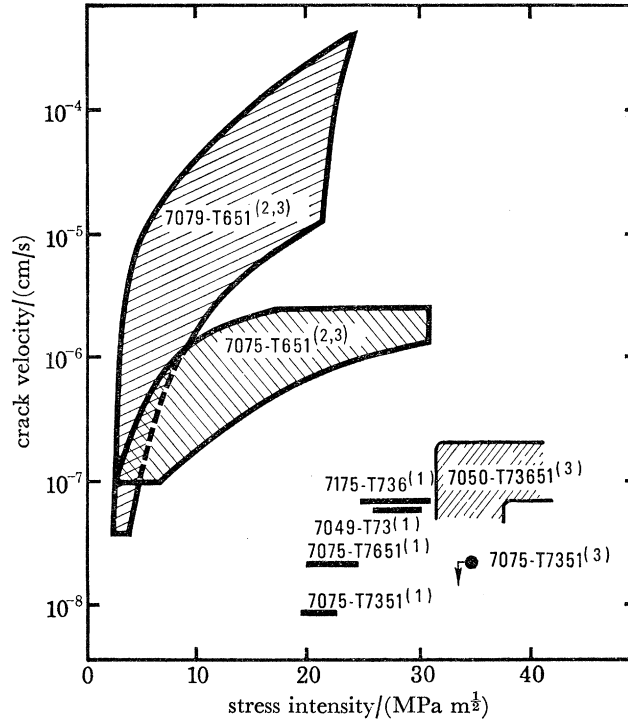


FIGURE 6. Effect of stress intensity on stress-corrosion crack propagation rates in Al-Zn-Mg-Cu alloys tested with short-transverse direction pre-cracked double cantilever beam specimens immersed in saturated aqueous NaCl solution (1, Speidel & Hyatt 1972; 2, Hyatt 1970; 3, Davies *et al.* 1974).

Various hypotheses have been proposed to explain the decisive effect of overageing on s.c.c. resistance and to relate it to microstructural features including the presence of precipitates in grain boundaries; precipitate-free zones (p.f.z.); electrochemical relationships between grains, precipitate-containing boundaries and p.f.z.; dislocation pile-up at grain boundaries; and matrix dislocation-precipitate interactions. The theory attributing s.c.c. in T6 temper to selective dissolution of an anodic p.f.z. accelerated by tension stress, and the higher resistance of T73 temper to equalization of the electrochemical potentials is consistent with some of the experimental observations (Sprowls & Brown 1969; Lifka & Sprowls 1972). The principal microstructural difference between the T6 and T73 tempers, however, is in the intragranular precipitate, which changes in structure from predominantly coherent g.p. zones of T6 to the metastable η' ,‡ of T73. The change in precipitate structure is the basis for the dislocation-

† As-forged length 5.33 m diameter 2.77 m mass 2360 kg.

‡ Equilibrium phase field extends across quaternary diagram joining MgZn_2 and AlCuMg . Evidence from electrochemical solution potential changes indicates Cu precipitates in the transition form of this phase during the higher temperature stage of the T73 treatment.

matrix precipitate interaction hypothesis (Speidel 1971). Although a variety of evidence can be cited as supporting one or the other of the mechanisms suggested, the question must be regarded as incompletely resolved at the present time.

1960-1975: 7075-T76, 7175-T736, 7049, 7475, 7050

Exfoliation corrosion – 7075-T76

Exfoliation corrosion had been a problem with 7075-T6 in certain aircraft components in which the resistance to s.c.c. had been entirely adequate because of part configuration and the nature and direction of stresses applied. Although 7075-T73 provided very high resistance to exfoliation, designers were reluctant to accept the strength reduction, about 15% for this temper. It was determined that less-drastic overageing than required for T73 produced high resistance to exfoliation. Based on this, the exfoliation-resistant T76 temper was evolved in mid-1960s providing strengths about midway between those of the T6 and T73 tempers. In some cases, it has been possible to apply T76 temper material in place of T6 without redesign or increase in weight. 7075-T76 products are highly resistant to s.c.c. at short-transverse direction stress levels of 170 MPa.

Higher strength with high s.c.c. resistance – 7175, 7049, 7050

Application of 7075-T73 in place of the higher strength T6 temper material in some cases required part redesign with an associated weight penalty. Reaction from the aircraft industry was a request to restore the T6 temper strength or higher combined with very high s.c.c. resistance. This motivated new research efforts on a wide geographic scale with more intensive evaluations of Zn, Mg and Cu amounts and ratios; variations in the supplementary elements for grain control and s.c.c. improvement – Cr, Mn and Zr; other additions, most prominently Ag; and numerous processing variants. Silver additions (0.3–0.4 mass % Ag) were widely investigated and generally considered to offer improved resistance to s.c.c. In the U.S. the benefits were not regarded as commensurate with the added cost, and the limited supply of Ag was considered a strong deterrent to extensive commercial use in an aluminum alloy.

7175-T736 Forgings. In 1968 forged products having significantly higher strength than 7075-T73 were developed employing special composition controls within the 7175 specification and non-conventional thermal processing and fabricating processes. The static strengths of these 'Premium Strength Forgings' are comparable with those of 7075-T6, and the resistance to s.c.c. is high although not considered equal to that of 7075-T73 products. Alloy 7175-T736 forgings have been used for a variety of specialized applications.

7049-T73. Alloy 7049 (Luhan & Summerson 1970) also introduced in 1968, differs from 7075 principally in its higher Zn and lower Cr contents, with somewhat more restrictive limits on the ranges of major and impurity elements. The higher Zn:Mg ratio was found to provide higher strength/s.c.c. resistance combinations, the lower Cr reduced quench sensitivity, and Cu, Fe and Si were controlled to improve ductility and fracture toughness at the higher strength level.† This alloy in T73 temper provides high resistance to s.c.c. coupled with strengths comparable with those of 7079-T6 in parts having a section thickness of 75 mm or over. Alloy 7049-T73 forgings have been in service for several years, prominently in landing gear retrofit programmes replacing 7079-T6 in F4, F111, B52 and C141 aircraft. Forgings of this alloy are

† A modification of 7049 with lower maximum limits on Fe (0.20% max.) and Si (0.15% max.) has been established as 7149.

also being specified for selected applications on newer aircraft such as DC10 and F5E (Anon 1974). Extrusions in T73511 and T76511 tempers are available, and design data are being developed for their application.

7050. An alloy of more recent vintage, 7050 (Staley, Hunsicker & Schmidt 1971), represents a greater departure in composition. Designed with particular reference to production of thick section parts in which lower quenching rates are unavoidable and for application of the T73 and T76-type precipitation treatments, 7050 has higher Zn and Cu contents relative to 7075, Zr in place of Cr, and more restrictive limits on the major elements as well as impurity elements. Zr, functioning as the high-temperature precipitate-forming addition to suppress recrystallization, reduces the strength loss with decreasing quench cooling rate characteristic of the Cr- or Mn-containing alloys. The ZrAl_3 phase formed by Zr avoids combination with principal alloying element species, in contrast with Cr which forms $\text{Al}_{12}\text{Mg}_2\text{Cr}$ or Mn which forms $\text{Al}_{20}\text{Cu}_2\text{Mn}_3$. The Zr-containing phase is present in smaller volume fraction as compact, nearly spherical particles. These factors were considered as favouring higher fracture toughness.

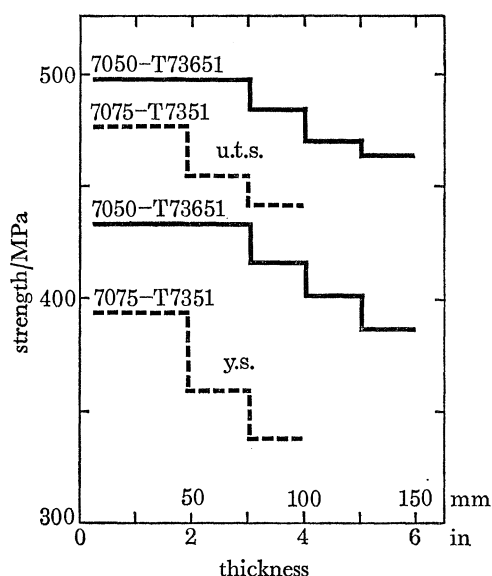


FIGURE 8. Comparative long-transverse direction design mechanical properties (ultimate tensile strength and tensile yield strength) of 7075-T7351 and 7050-T73651 plate as function of thickness.

The higher Cu content, close to the limit of solid solubility in combination with Zn and Mg, increases strength developed at the temperature of the second-stage precipitation treatment (Staley 1974), and, following trends of earlier observations, provides higher strength for a specific level of resistance to s.c.c. (Staley 1973). Higher Cu contents than those of the 7075 range are not similarly effective in compositions containing Cr such as 7178.

Selection of second-stage precipitation conditions for 7050, as with 7075 and 7049, involves a trade-off compromise between strength and resistance to s.c.c. Tempers designated by types T76, T736 and T73 are established. The resistance to s.c.c. of 7050-T736 is considered to be at least equivalent to that of 7175-T736, which has been used for several years with satisfactory experience, but does not equal the 'virtual immunity' of 7075-T73 or 7050-T73.

The strength advantage of 7050-T73651 compared with 7075-T7351, increasing with thickness, is apparent in the design yield strength values for plate charted as a function of

thickness in figure 8. Beyond the thickness shown for 7075-T7351, the strengths of this alloy are generally too low to be of important structural value, while up to 150 mm the Zr-containing alloy has nearly as high strength as the thinnest 7075-T7351 product.

Evaluation of resistance to s.c.c. of 7050 involved hundreds of tests principally with small tensile specimens, 3.2 mm diameter in the stressed section, machined from a variety of products. Emphasis was placed on the performance of specimens oriented to allow stress application in the short transverse direction relative to the wrought structure. Exposures included industrial atmosphere and seacoast environment as well as recognized accelerated laboratory tests.

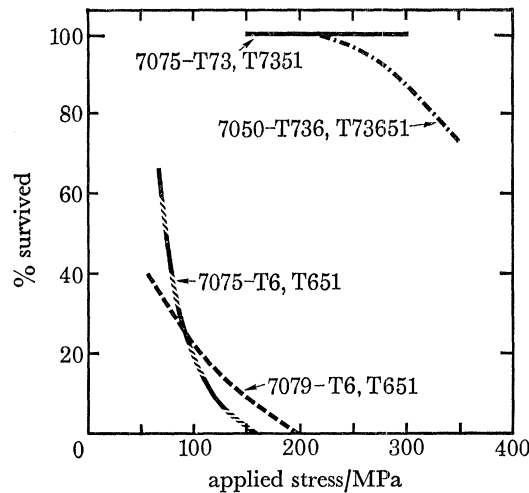


FIGURE 9. Results of industrial atmosphere stress-corrosion tests of 50 to 63 mm thick plate during four years exposure at New Kensington, Pa., U.S.A.

Results from four years exposure in the industrial atmosphere of specimens from development-stage plate complying with the 7050-T736 specification and from similar tests of the earlier alloys provided the information for figure 9. Performance in the seacoast exposure was comparable to that in the industrial environment. Specimens removed from plate produced in a commercial run have been exposed for over 16 months at stresses up to 310 MPa with no s.c.c. failures in either industrial or seacoast environment. These observations, coupled with those from numerous accelerated exposure (NaCl – alternate immersion) tests support a high ranking of resistance to s.c.c. at stress levels to at least 240 MPa for plate and other 7050-T736 products.

TABLE 2. REPRESENTATIVE (MEAN) VALUES OF PLANE-STRAIN STRESS INTENSITY FACTOR, K_{Ic} , FOR PLATE IN 25 TO 50 MM THICKNESS RANGE†

alloy-temper	$K_{Ic}/(\text{MPa m}^{1/2})$, in direction indicated		
	l-t	t-l	s-l‡
7075-T651	29.7	25.3	19.8
7475-T651	44.0	38.5	29.7
7075-T7651	28.6	26.4	20.9
7475-T7651	47.3	41.8	31.9
7075-T7351	34.1	29.7	22.0
7475-T7351	55.0	49.5	36.3
7050-T73651	38.2	31.7	26.9

† Compact-tension specimens per ASTM E399.

‡ Values for 50 mm thickness only.

Fracture toughness tests were performed on a range of commercially-fabricated 7050-T736 products to determine applicable values of plane-strain stress intensity factor, K_{Ic} . The values obtained exceeded those determined for the same types of products in 7075-T73 (e.g. plate data, table 2). This is significant because the normal accompaniment of higher strength, if derived through change in precipitation treatment or increased alloy content maintaining the same purity, is reduced fracture toughness.

Alloy 7050 plate and forgings are being produced currently for several retrofit programmes replacing the originally specified material which was considered inadequate in resistance to either exfoliation corrosion or s.c.c. Several applications in new designs have been made and increased use of this material is expected.

Other alloys related to 7050

Research and development programmes in the U.S.A. and other countries have led to other compositions that are related to 7050 in that Zr is employed in place of Cr or Mn, or in combination with smaller additions of these elements (DiRusso 1969; DiRusso & Buratti 1974; Hyatt & Schimmelbusch 1970; Thompson 1972; Fischer, Lynker & Markworth 1972; Reynolds & Harris 1974). These compositions are characterized by lower Cu content than 7050, in one case coupled with an addition of Ag. The most recently described composition, Alcan GB X3066 (DTD 5120) with a nominal Cu content of 1.7% is reported to have an advantage over 7050 from its lower Cu content in ingot casting and fabricating characteristics (Reynolds & Harris 1974). Additional comparative evaluations of these Zr-containing alloys will assuredly be made resulting in eventual selection of one or another of the present specifications or perhaps an evolutionary compromise.

Higher fracture toughness alloy 7475

Although 7075-T6 aircraft structure components are considered as having an excellent performance record from the viewpoint of mechanical reliability and damage tolerance, additional assurance against the probability of critical stress failure in flawed, poorly machined, service damaged or fatigue cracked parts is regarded as highly desirable. About the same time that more quantitative techniques for measuring fracture toughness were evolving from the basic concepts of linear elastic fracture mechanics, programmes were initiated to achieve superior fracture toughness at high strength levels.

From early work (Nock & Wooll 1963) it was concluded that the Al-Zn-Mg-Cu alloys exhibited an inherent fracture toughness/strength combination advantage over Al-Cu-Mg types in the elevated temperature precipitation strengthened tempers. This is considered as related in part to differences in deformation mechanisms associated with the coherent type precipitate of the former contrasted with the non-deforming precipitate of the latter although other factors are also contributory. For either type alloy, changes in content of major alloying elements or in state of precipitation produce inversely related changes in strength and fracture toughness, a trade-off relation not uncommon to other metal-base alloys. The improvement desired, however, was increased toughness with no sacrifice in strength.

Our early investigations, since confirmed by published studies of many other investigators, demonstrated that highly significant improvements in fracture toughness could be obtained by reducing the volume fraction of insoluble (and undissolved soluble) intermetallic-phase particles present in the alloy structure. Work was concentrated on the 7075 composition since gains were modest in higher strength alloys, such as 7178.

The 7475 composition specification differs from 7075 in the more restrictive limits placed on the impurity elements, principally Fe and Si (0.12% and 0.10% max., respectively) which form sparsely soluble phases, and includes minor but significant adjustments in Mg and Cu ranges. The composition changes were coupled with modified process and mill controls found to provide supplemental toughness improvement through changes in grain size, homogeneity, and size and interparticle spacing of the $Al_{12}Mg_2Cr$ precipitate.

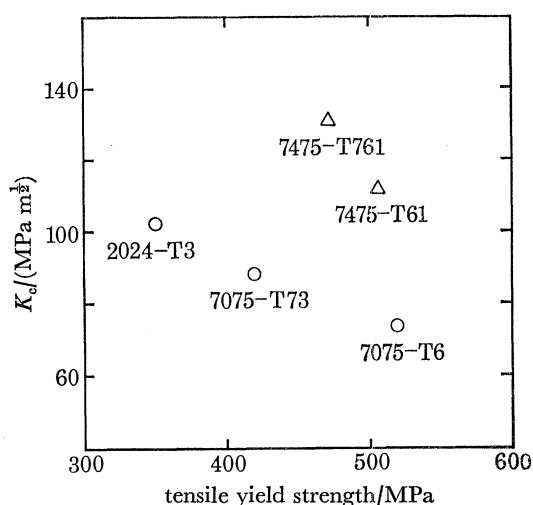


FIGURE 10. Fracture toughness (stress intensity factor, K_{IC} , values) in relation to tensile yield strengths for sheet of five alloy/temper combinations. Mean values from 400 mm wide centre-cracked panels tested without anti-buckling guides.

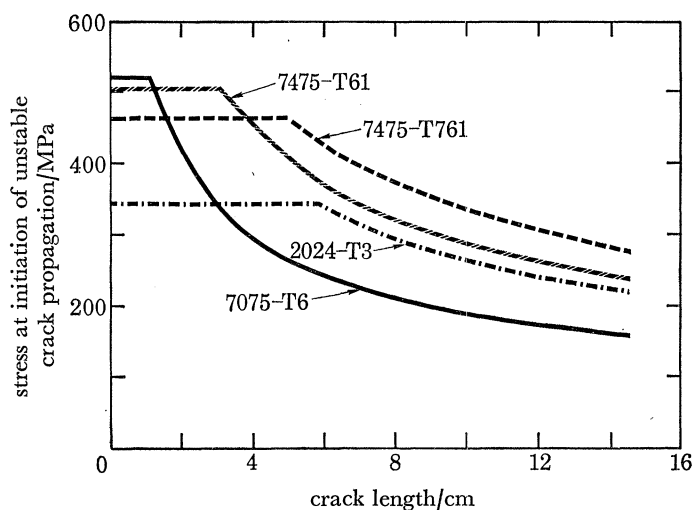


FIGURE 11. Gross section stress at initiation of unstable crack propagation plotted against crack length for wide sheet panels of four alloy/temper combinations.

Combinations of fracture toughness and tensile yield strength for 7475 sheet in two tempers produced in accordance with the process specifications are compared with those of sheet of conventional commercial grade alloys in figure 10. Alloy 2024-T3 (naturally aged temper) has been the American aircraft designers historical preference for those structural members in which high fracture toughness is most essential. The aim with 7475 was to couple a level of

toughness at least equal to that of 2024-T3 with higher strength, and this was achieved with 7475. This is illustrated further by curves of residual strength of large cracked sheet panels as a function of crack size in figure 11. It is apparent that cracks which develop by fatigue, for example, in 7475-T61 or 2024-T3 may grow to about double the size of those in 7075-T6 before the stress becomes critical and unstable fracture occurs. The exfoliation-resistant 7475-T761 has higher crack-length tolerance. Comparative plane-strain fracture toughness data for plate of alloys 7075, 7475 and 7050 in applicable tempers are listed in table 2.

Fatigue crack growth rates at high stress intensity levels are reduced with the higher toughness material. This advantage coupled with the increased flaw size, crack length and damage tolerance are characteristics considered highly attractive for new aircraft designs as well as retrofit programmes, and appreciable quantities of 7475 rolled products have been supplied to both American and European aircraft manufacturers. (New versions of Al-Cu-Mg and Al-Cu alloys (2124, 2048, 2419) having higher fracture toughness in precipitation heat treated tempers preferred for supersonic aircraft have also been made available in recent years. The improvement in fracture toughness of these materials involved application of basic principles similar to those outlined for 7475 products.)

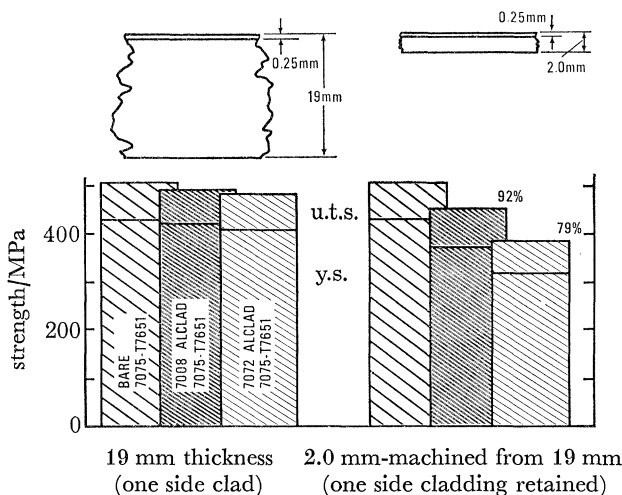


FIGURE 12. Effect of high- (7008) and low- (7072) strength cladding alloys on design strengths of composite (one-side alclad) plate before and after machining from non-clad surface. Data for 2 mm final thickness applies to thin portions of 'sculptured' Lockheed L-1011 wing skins.

High strength alclad sheet and plate

Conventional alclad 7075 has one or both surfaces clad with a thin layer (1.5 to 4% of thickness specified by thickness range) of electrochemically anodic 7072 (1% Zn alloy) to provide galvanic corrosion protection. This surface layer is relatively soft, not highly resistant to abrasion, and measurably reduces strength and fatigue resistance.

The high strength alclad sheet and plate of this alloy family, exemplified by 7008 alclad 7075, was conceived, produced in both laboratory and plant, and evaluated in the late 1940s. However, it was not used commercially for about twenty years until development of the Lockheed L1011 (Frisbee 1974). The cladding alloy is of the Cu-free Al-Zn-Mg type, develops strength by precipitation and is anodic to the 7075 core affording galvanic protection.

In sheet gauges the strength advantage amounts to only about 5% but this is coupled with higher resistance to abrasion and to fatigue. In the wing panels of the L1011 produced from

plate, the higher strength cladding provides a static strength advantage of as much as 15% (figure 12) because of the greater proportionate thickness of cladding to core alloy in the thin webs formed by removal of metal from one surface. The harder surface layer also improves contour forming by shot peening (Brandel & Klass 1971).

Prospectus

Considerable advance has been made in overcoming early disadvantages of the Al-Zn-Mg-Cu alloy products – inadequate resistance to s.c.c. and exfoliation corrosion. Fracture toughness has been improved at high static strength levels. At this time these properties are not considered as restricting the design and application of these alloys for subsonic aircraft. On the other hand, the structures and components in many cases are regarded as fatigue-critical from a design stress viewpoint, and increased resistance to fatigue crack initiation and decreased growth rate would enhance the value of aluminum alloys for these applications. Many mechanistic studies of fatigue have been made, and these efforts are being intensified. However, they have not yet provided unequivocal guidelines to the alloy designer concerning direction or paths to seek increased fatigue performance. The higher toughness materials have shown some advantage in reduced growth rates at high stress intensities, but there is evidence that delayed initiation and growth retardation in earlier stages may occur in compositions and structures characterized by features that are quite in opposition to those favouring high toughness. These matters deserve additional study and further definition of basic principles that can be applied towards improvement.

Unique metallurgical structures developed by unconventional thermomechanical treatments have been reported to increase ductility, fracture toughness and resistance to fatigue. Treatments applied in final stages of processing (f.t.m.t.) generally consist of a sequence of precipitation treatments interspersed with cold or warm working operations. These procedures are intended to develop precipitate-stabilized, dense, uniform dislocation distributions.

More recently another type of thermomechanical treatment (i.t.m.t.) has been proposed to be applied much earlier during fabrication and intended for application to hot-worked products that are normally characterized by unrecrystallized structures. The i.t.m.t. processes are designed to develop either a final recrystallized structure of unusually fine grain size or a fine-grained recrystallized structure that is modified by a moderate amount of final hot working.

Evaluation of materials processed by these treatments has not been sufficiently complete to demonstrate that they fulfill all engineering property requirements or that the possible advantage in one characteristic is not offset by an untenable loss in another. Some of the treatments involve considerable additional cost and are difficult to control on a commercial scale. These treatments are not now applied commercially, but investigations are continuing to determine their possible merit, feasibility and cost effectiveness.

The development and successful application of these materials based on the quaternary Al-Zn-Mg-Cu system is the result of the efforts of many scientists and engineers presently and formerly associated with industry, government agencies and academia. The good record of service and reliability which these present-day materials have accumulated is a tribute to these people and their efforts. Although further advances in engineering properties will undoubtedly be made with this family of materials, the products available today should very satisfactorily fill the needs of the aircraft industry for the next several years.

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