

# High-Strength Aluminum P/M Alloys

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POWDER METALLURGY (P/M) technology provides a useful means of fabricating net-shape components that enables machining to be minimized, thereby reducing costs. Aluminum P/M alloys can therefore compete with conventional aluminum casting alloys, as well as with other materials, for cost-critical applications. In addition, P/M technology can be used to refine microstructures compared with those made by conventional ingot metallurgy (I/M), which often results in improved mechanical and corrosion properties. Consequently, the usefulness of aluminum alloys for high-technology applications, such as those in aircraft and aerospace structures, is extended. This article describes and reviews the latter of these two areas of aluminum P/M technology where high strength and improved combinations of properties are obtained by exploiting the inherent advantages of P/M for alloy design.

The metallurgical reasons for the microstructural refinement made possible by P/M are discussed. The two broad high-strength P/M technologies—rapid solidification (RS) and mechanical attrition (mechanical alloying/dispersion strengthening)—are described. The various steps in aluminum P/M technology are explained to produce an appreciation for the interrelationship between powder processing and resultant properties. Finally, the major thrust areas of P/M alloy design and development are reviewed and some properties of the leading aluminum P/M alloys discussed. No attempt is made to provide design-allowable mechanical properties, and the data presented for the various P/M alloys may not be directly comparable because of differences in product forms. Nevertheless, the properties presented will enable the advantages of aluminum P/M alloys to be appreciated. Greater details of aluminum P/M alloys are provided in other reviews (Ref 1 to 7). Conventional pressed and sintered aluminum P/M alloys for less demanding applications are described in the Appendix to this article.

## Advantages of Aluminum P/M Technology

Aluminum alloys have numerous technical advantages that have enabled them to be

one of the dominant structural material families of the 20th century. Aluminum has low density ( $2.71 \text{ g/cm}^3$ ) compared with competitive metallic alloy systems, good inherent corrosion resistance because of the continuous, protective oxide film that forms very quickly in air, and good workability that enables aluminum and its alloys to be economically rolled, extruded, or forged into useful shapes. Major alloying additions to aluminum such as copper, magnesium, zinc, and lithium—alone, or in various combinations—enable aluminum alloys to attain high strength. Designers of aircraft and aerospace systems generally like using aluminum alloys because they are reliable, reasonably isotropic, and low in cost compared to more exotic materials such as organic composites.

Aluminum alloys do have limitations compared with competitive materials. For example, Young's modulus of aluminum (about 70 GPa, or  $10 \times 10^6$  psi) is significantly lower than that of ferrous alloys (about 210 GPa, or  $30 \times 10^6$  psi) and titanium alloys (about 112 GPa, or  $16 \times 10^6$  psi). This lower modulus is almost exactly offset by the density advantage of aluminum compared to iron- and titanium-base alloys. Nevertheless, designers could exploit higher-modulus aluminum alloys in many stiffness-critical applications.

Although aluminum alloys can attain high strength, the strongest such alloys have often been limited by stress-corrosion cracking (SCC) susceptibility in the highest-strength tempers. For example, the high-strength 7xxx alloys (Al-Zn-Mg and Al-Zn-Mg-Cu) can have severe SCC susceptibility in the highest-strength (T6) tempers. To remedy this problem, overaged (T7) tempers have been developed that eliminate SCC susceptibility, but with a 10 to 15% strength penalty.

The melting point of aluminum, 660 °C (1220 °F), is lower than that of the major competitive alloy systems: iron-, nickel-, and titanium-base alloys. As might be expected, the mechanical properties of aluminum alloys at elevated temperatures are often not competitive with these other systems. This limitation of aluminum alloys is

of particular concern to designers of aircraft and aerospace structures, where high service temperatures preclude the use of aluminum alloys for certain structural components.

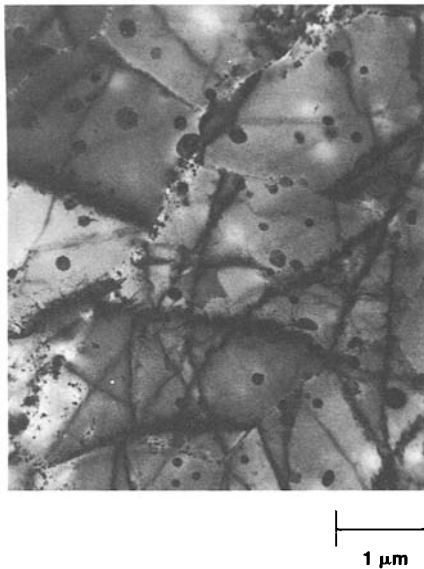
The number of alloying elements that have extensive solid solubility in aluminum is relatively low. Consequently, there are not many precipitation-hardenable aluminum alloy systems that are practical by conventional I/M. This can be viewed as a limitation when alloy developers endeavor to design improved alloys. Aluminum P/M technology enables the aforementioned limitations of aluminum alloys to be overcome to various extents, while still maintaining most of the inherent advantages of aluminum.

**Structure/Property Benefits.** The advantages of P/M stem from the ability of small particles to be processed. This enables:

- The realization of RS rates
- The uniform introduction of strengthening features, that is, barriers to dislocation motion, from the powder surfaces

The powder processes of rapid solidification and mechanical attrition lead to microstructural grain refinement and, in general, better mechanical properties of the alloy. Specifically, the smaller the mean free path between obstacles to dislocation motion, the greater the strengthening. In addition, finer microstructural features are also less apt to serve as fracture-initiating flaws, thereby increasing toughness.

The RS rates made possible by P/M enable microstructural refinement by several methods. For example, grain size can be reduced because of the short time available for nuclei to grow during solidification. Finer grain size results in a smaller mean free path between grain boundaries, which are effective barriers to dislocation motion, leading to increased "Hall-Petch" strengthening. In addition, RS can extend the alloying limits in aluminum by enhancing supersaturation and thereby enabling greater precipitation hardening without the harmful segregation effects from overalloyed I/M alloys. Moreover, elements that are essentially insoluble in the solid state, but have

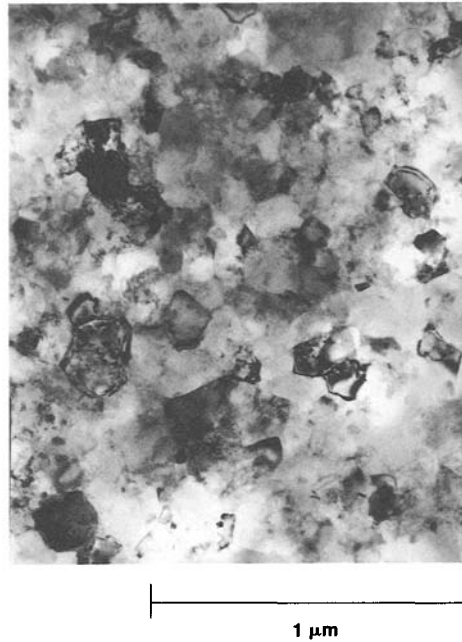


**Fig. 1** Experimental Al-Zn-Mg-Cu-Co RS alloy that contains a fine distribution of spherical  $\text{Co}_2\text{Al}_9$  particles and fine grain size (2 to 5  $\mu\text{m}$ ). Courtesy of L. Christodoulou, Martin Marietta Laboratories

significant solubility in liquid aluminum, can be uniformly dispersed in the powder particles during RS. This can lead to the formation of novel strengthening phases that are not possible by conventional I/M, while also suppressing the formation of equilibrium phases that are deleterious to toughness and corrosion resistance. The photomicrograph shown in Fig. 1 exemplifies the microstructural refinement possible by RS-P/M processing. This experimental Al-6.7Zn-2.4Mg-1.4Cu-0.8Co alloy has a grain size of 2 to 5  $\mu\text{m}$  and a fine distribution of  $\text{Co}_2\text{Al}_9$  particles of about 0.1 to 0.4  $\mu\text{m}$ . Comparable I/M alloys have a grain size that is almost an order of magnitude larger. In addition, making this alloy by I/M would lead to very coarse cobalt-containing particles because of the low solid-solubility of cobalt in aluminum. These coarse particles would significantly degrade toughness. Thus, RS processes constitute one of the two major classes of high-strength P/M technology.

The other class of high-strength P/M technology relies on the introduction of strengthening features from powder surfaces, which can be accomplished on a fine scale because of the high surface-area-to-volume ratio of the powder particles. Most, but not all, such processes involve ball milling and are called mechanical attrition processes.

Oxides can be easily introduced from powder surfaces by consolidating the powder and hot working the product (see the section "Mechanical Attrition Process" in this article). The fine oxides can improve strength by oxide dispersion strengthening (ODS), and in addition, by substructural strengthening resulting from the disloca-



**Fig. 2** Experimental mechanically alloyed Al-1.5Li-0.9O-0.6C alloy featuring an ultrafine grain/subgrain size. Courtesy of J.R. Pickens, unpublished research

tions created during working, which are generated by dislocation-oxide interactions. However, ODS can be increased by mechanically attriting the powder particles to more finely disperse the oxides. Ball milling in the presence of organic surfactants allows carbides to be dispersed in a similar fashion. Finally, ball milling aluminum and other powders can enable fine intermetallic dispersoids to form during milling or subsequent thermomechanical processing.

The interplay between dispersed oxides, carbides (or other ceramics), intermetallics, and dislocations during hot, warm, or cold working can enable great refinement in grain and subgrain size. This can result in significant strengthening because grain and subgrain boundaries can be effective barriers to dislocation motion. For example, an experimental Al-1.5Li-0.9O-0.6C mechanically alloyed material had an extremely fine grain/subgrain size of 0.1  $\mu\text{m}$  (Fig. 2). The ultrafine oxides and carbides, coupled with this fine grain size, enabled the alloy to have a tensile strength of nearly 800 MPa (115 ksi) with 3% elongation. An I/M alloy containing this much oxygen and carbon that is effectively contained in finely dispersed particles is not viable.

P/M processing, including both rapid solidification and mechanical attrition processes, provides alloy designers with additional flexibility resulting from the inherent advantage of working with small bits of matter. In addition to refinement of strengthening features, different metallic powders can be mixed to form duplex microstructures. Furthermore, ceramic powders may be mixed with aluminum alloy

powders to form metal-matrix composites (MMCs). However, the fine particles of matter must be consolidated and formed into useful shapes. The potential benefits of P/M technology can be realized, or lost, in the critical consolidation-related and forming processes.

## Aluminum P/M Processing

There are several steps in aluminum P/M technology that can be combined in various ways, but they will be conveniently described in three general steps:

- Powder production
- Powder processing (optional)
- Degassing and consolidation

Powder can be made by various RS processes including atomization, splat quenching to form particulates, and melt spinning to form ribbon. Alternatively, powder can be made by non-RS processes such as by chemical reactions including precipitation, or by machining bulk material. Powder-processing operations are optional and include mechanical attrition (for example, ball milling) to modify powder shape and size or to introduce strengthening features, or comminution such as that used to cut melt-spun ribbon into powder flakes for subsequent handling.

Aluminum has a high affinity for moisture, and aluminum powders readily adsorb water. The elevated temperatures generally required to consolidate aluminum powder causes the water of hydration to react and form hydrogen, which can result in porosity in the final product, or under confined conditions, can cause an explosion. Consequently, aluminum powder must be degassed prior to consolidation. This is often performed immediately prior to consolidation at essentially the same temperature as that for consolidation to reduce fabrication costs. Consolidation may involve forming a billet that can be subsequently rolled, extruded, or forged conventionally, or the powder may be consolidated during hot working directly to finished-product form. The various steps in aluminum P/M technology will now be discussed in greater detail.

### Powder Production

**Atomization** is the most widely used process to produce aluminum powder. Aluminum is melted, alloyed, and sprayed through a nozzle to form a stream of very fine particles that are rapidly cooled—most often by an expanding gas. Atomization techniques have been reviewed extensively in Ref 8 and 9 and in *Powder Metallurgy*, Volume 7 of the 9th Edition of *Metals Handbook*. Cooling rates of  $10^3$  to  $10^6$  K/s are typically obtained.

**Splat cooling** is a process that enables cooling rates even greater than those obtained in atomization. Aluminum is melted

and alloyed, and liquid droplets are sprayed or dropped against a chilled surface of high thermal conductivity—for example, a copper wheel that is water cooled internally. The resultant splat particulate is removed from the rotating wheel to allow subsequent droplets to contact the bare, chilled surface. Cooling rates of  $10^5$  K/s are typical, with rates up to  $10^9$  K/s reported.

**Melt-spinning techniques** are somewhat similar to splat cooling. The molten aluminum alloy rapidly impinges a cooled, rotating wheel, producing rapidly solidified product that is often in ribbon form. The leading commercial melt-spinning process is the planar flow casting (PFC) process developed by M.C. Narisimhan and co-workers at Allied-Signal Inc. (Ref 10). The liquid stream contacts a rotating wheel at a carefully controlled distance to form a thin, rapidly solidified ribbon and also to reduce oxidation. The ribbon could be used for specialty applications in its PFC form but is most often comminuted into flake powder for subsequent degassing and consolidation.

In the aforementioned RS powder-manufacturing processes, partitionless (that is, no segregation) solidification can occur, and with supersaturation RS can ultimately lead to greater precipitation strengthening. Furthermore, with the highest solidification rates, crystallization can be suppressed. The novelties of RS aluminum microstructures and nonequilibrium phase considerations have been reviewed (Ref 11, 12).

**Other Methods.** Powder can also be made from machining chips or via chemical reactions (Ref 1, 13). Such powders should be carefully cleaned before degassing and consolidation.

### Mechanical Attrition Process

Mechanical attrition processes often involve ball milling in various machines and environments. Such processes can be used to control powder size and distribution to facilitate flow or subsequent consolidation, introduce strengthening features from powder surfaces, and enable intermetallics or ceramic particles to be finely dispersed. The two leading mechanical attrition processes today—mechanical alloying in the United States, and reaction milling in Europe—are improvements on sinter-aluminum-pulver (SAP) technology developed by Irmann in Austria (Ref 14, 15).

**SAP Technology.** In 1946, Irmann and co-workers were preparing rod specimens for spectrographic analysis by hot pressing mixtures of pure aluminum and other metal powders. They noticed the unexpectedly high hardness of the resulting rods. Mechanical property evaluations revealed that the hot-pressed material had strength approaching that of aluminum structural alloys. Based on microstructural evaluations, Irmann attributed the high strength of the hot-pressed compacts to the breakdown of

the surface oxide film on the powder particles during hot pressing. Irmann performed mechanical tests at elevated temperatures and showed that these alloys not only had surprisingly high elevated-temperature strength, but retained much of their room-temperature strength after elevated-temperature exposure.

Irmann also ball milled aluminum powder and noticed that it was similar to the flaky powder used in paint pigment. Hüttig studied the formation of the flaky powder during ball milling and noticed the competition between comminution and welding of the powder particles (Ref 16). This fracture and welding of the powder particles caused the surface oxide film to be ruptured and become somewhat dispersed in the powder particles. Irmann called these hot-pressed materials sinter-aluminum-pulver (SAP), later referred to as sintered aluminum powder or sintered aluminum product in the United States. Various SAP alloys were developed in the 1950s that displayed excellent elevated-temperature properties (Ref 1).

**The mechanical alloying process** is a high-energy ball-milling process that employs a stirred ball mill called an attritor, a shaken ball mill, or a conventional rotating ball mill (Ref 17, 18). The process is performed in the presence of organic surfactants, for example, methanol, stearic acid, and graphite, to control the cold welding of powder particles and provide oxygen and carbon for dispersion strengthening (Ref 19, 20). Mechanical alloying claims the advantage of milling under “dry” conditions—that is, not in mineral oil, as in some SAP processing, which must be removed after milling—although dry SAP ball milling in the presence of stearic acid was reported in the 1950s (Ref 1).

Elemental powders may be milled with aluminum powders to effect solid-solution strengthening or to disperse intermetallics. The dispersed oxides, carbides, and/or intermetallics create effective dislocation sources during the milling process and suppress dislocation annihilation during subsequent working operations, which result in greatly increased dislocation density. Thus, mechanical alloying enables the effective superimposition of numerous strengthening mechanisms (Ref 2, 19-21), including:

- Oxide dispersion
- Carbide dispersion
- Fine grain size
- High dislocation density and substructure
- Solid-solution strengthening

Consequently, high strength can be obtained without reliance on precipitation strengthening, which may introduce problems such as corrosion and SCC susceptibility, and propensity for planar slip. Nevertheless, the aforementioned five strengthening contributions can be augmented by precipitation strengthening as well as intermetallic dispersion strengthening.

**Reaction milling** is another mechanical attrition process derived from SAP technology (Ref 22-24). It is extremely similar to mechanical alloying, and subtle differences between the two processes appear to have little appreciable effect on the compositions that can be processed and the resulting microstructures produced. Investigations in Europe have also successfully superimposed numerous strengthening features in aluminum P/M alloys, as described above.

### Powder Degassing and Consolidation

The water of hydration that forms on aluminum powder surfaces must be removed to prevent porosity in the consolidated product. Although solid-state degassing has been used to reduce the hydrogen content of aluminum P/M wrought products, it is far easier and more effective to remove the moisture from the powder. Degassing is often performed in conjunction with consolidation, and the most commonly used techniques are described below. The various aluminum fabrication schemes are summarized in Fig. 3.

**Can Vacuum Degassing.** This is perhaps the most widely used technique for aluminum degassing because it is relatively non-capital intensive. Powder is encapsulated in a can, usually aluminum alloys 3003 or 6061, as shown schematically in Fig. 4. A spacer is often useful to increase packing and to avoid safety problems when the can is welded shut. The author has found that packing densities are typically 60% of theoretical density when utilizing this method on mechanically alloyed powders. Care must be used to allow a clear path for evolved gases through the spacer to prevent pressure buildup and explosion.

To increase packing density, the powder is often cold isostatically pressed (CIP) in a reusable polymeric container before insertion into the can. Powder densities in the CIPed compact of 75 to 80% theoretical density are preferred because they have increased packing density with respect to loose-packed powder, yet allow sufficient interconnected porosity for gas removal. At packing densities of about 84% and higher, effective degassing is not possible for several atomized aluminum-alloy powders (Ref 25, 26). Furthermore, one must control CIP parameters to avoid inhomogeneous load transfer through the powder, which can lead to excessive density in the outer regions of the cylindrical compact and much lower densities in the center. Such CIP parameters often must be developed for a specific powder and compact diameter (see the article “Cold Isostatic Pressing of Metal Powders” in Volume 7 of the 9th Edition of *Metals Handbook*).

The canned powder is sealed by welding a cap that contains an evacuation tube as shown in Fig. 4. After ensuring that the can contains no leaks, the powder is vacuum

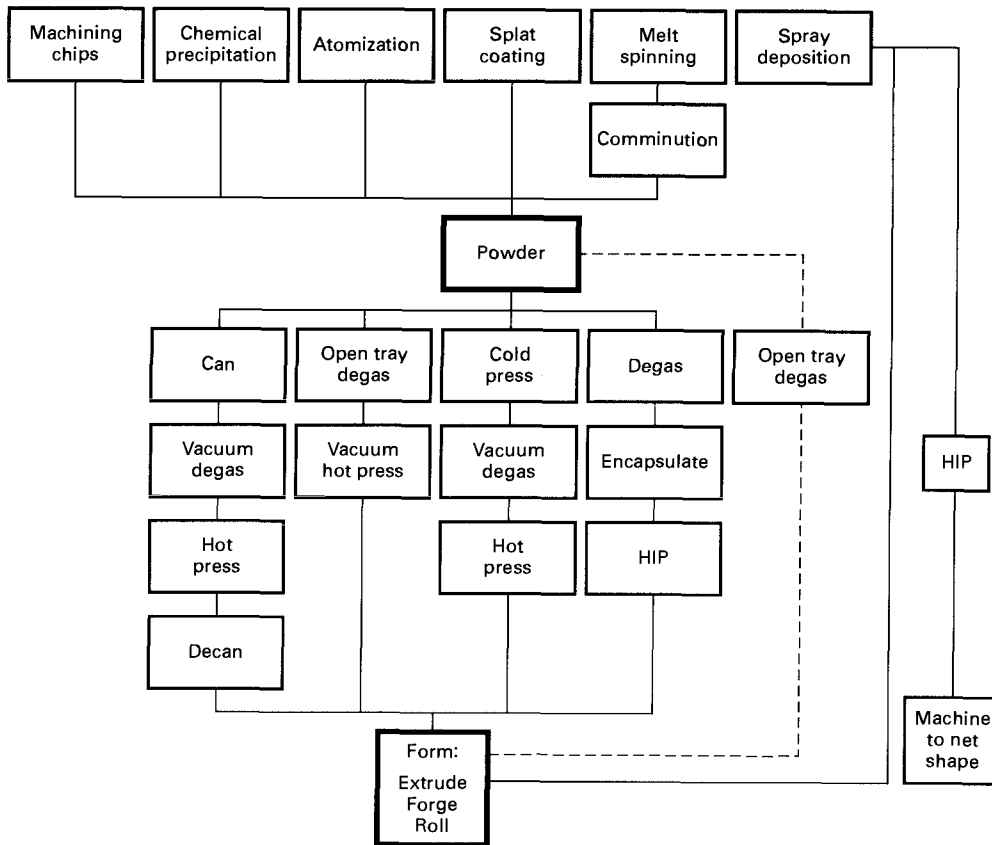


Fig. 3 Aluminum P/M fabrication schemes

degassed while heating to elevated temperatures. The rate in gas evolution as a function of degassing temperature depends on powder size, distribution, and composition. The ultimate degassing temperature should be selected based on powder composition, considering tradeoffs between resulting hydrogen content and microstructural coarsening. For example, an RS-P/M precipitation-hardenable alloy that is to be welded would likely be degassed at a relatively high temperature to minimize hydrogen content (hotter is not always better). Coarsening would not significantly decrease the strength of the resulting product because solution treatment and aging would be subsequently performed and provide most of the strengthening. On the other hand, a mechanically attrited P/M alloy that relies on substructural strengthening and will serve in a mechanically fastened application might be degassed at a lower temperature to reduce the annealing out of dislocations and coarsening of substructure.

When a suitable vacuum is achieved (for example,  $<5$  millitorr), the evacuation tube is sealed by crimping. The degassed powder compact can then be immediately consolidated to avoid the costs of additional heating. An extrusion press using a blind die (that is, no orifice) is often a cost-effective means of consolidation.

**Dipurative Degassing.** Roberts and co-workers at Kaiser Aluminum & Chemical Corporation have developed an improved degassing method called "dipurative" degassing (Ref 27). In this technique, the vacuum-degassed powder, which is often canned, is backfilled with a dipurative gas (that is, one that effectively removes water of hydration) such as extra-dry nitrogen, and then re-evacuated. Several backfills and evacuations can be performed resulting in lower hydrogen content. In addition, the degassing can often be performed at lower temperatures to reduce microstructural coarsening.

**Vacuum Degassing in a Reusable Chamber.** The cost of canning and decanning adversely affects the competitiveness of aluminum P/M alloys. This cost can be alleviated somewhat by using a reusable chamber for vacuum hot pressing. The powder or CIPed compact can be placed in the chamber and vacuum degassed immediately prior to compaction in the same chamber. Alternatively, the powder can be "open tray" degassed, that is, degassed in an unconfined fashion, prior to loading into the chamber. The processing time to achieve degassing in an open tray can be much less than that required in a chamber or can, thereby increasing productivity. Care must be exercised to load the powder into the

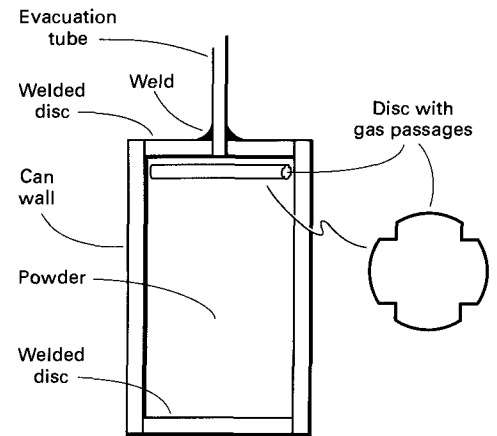


Fig. 4 Degassing can used for aluminum P/M processing. Source: Ref 2

chamber using suitable protection from ambient air and moisture. The compacted billet can then be formed by conventional hot-working operations.

**Direct Powder Forming.** One of the most cost-effective means of powder consolidation is direct powder forming. Degassed powder, or powder that has been manufactured with great care to avoid contact with ambient air, can be consolidated directly during the hot-forming operation. Direct powder extrusion and rolling have been successfully demonstrated numerous times over the past two decades (Ref 28, 29). It still remains an attractive means for decreasing the cost of aluminum P/M.

**Hot Isostatic Pressing (HIP).** In HIP, degassed and encapsulated powder is subjected to hydrostatic pressure in a HIP apparatus (see the article "Hot Isostatic Pressing of Metal Powders" in Volume 7 of the 9th Edition of *Metals Handbook*). Can vacuum degassing is often used as the precursor step to HIP. Furthermore, net-shape encapsulation of degassed powder can be used to produce certain near net-shape parts. Relatively high HIP pressures ( $\sim 200$  MPa, or 30 ksi) are often preferred. Unfortunately, the oxide layer on the powder particle surfaces is not sufficiently broken up for optimum mechanical properties. A subsequent hot-working operation that introduces shear-stress components is often necessary to improve ductility and toughness.

**Rapid Omnidirectional Consolidation.** Engineers at Kelsey-Hayes Company have developed a technique to use existing commercial forging equipment to consolidate powders in several alloy systems (Ref 30). Called rapid omnidirectional consolidation (ROC), it is a lower-cost alternative to HIP. In ROC processing, degassed powder is loaded into a thick-walled "fluid die" that is made of a material that plastically flows at the consolidation temperature and pressure, and which enables the transfer of hydrostatic stress to the powder. Early fluid dies were made of mild steels or a Cu-10Ni alloy,

**Table 1** Nominal compositions of aluminum P/M alloys for ambient-temperature service

Alloy	Composition, wt%									
	Zn	Mg	Cu	Co	Zr	Ni	Cr	Li	O	C
7090	8.0	2.5	1.0	1.5	...	...	...	...	...	...
7091	6.5	2.5	1.5	0.4	...	...	...	...	...	...
CW67	9.0	2.5	1.5	...	0.14	0.1	...	...	...	...
7064	7.4	2.4	2.1	0.75	0.3	...	0.15	...	0.2	...
Al-9052	...	4.0	...	...	...	...	...	...	0.5	1.1
Al-905XL	...	4.0	...	...	...	...	...	1.3	0.4	1.1

**Table 2** Longitudinal tensile properties from experimental extrusions of aluminum P/M alloys designed for ambient-temperature service

Alloy	Temper	Yield strength, MPa (ksi)	Ultimate tensile strength, MPa (ksi)	Elongation, %	Reference
7091	T6E192	600 (87)	640 (93)	13	41(b)
	T7	545 (79)	595 (86)	11	41(b)
7090	T6511	640 (93)	675 (98)	10	41(b)
	T7	580 (84)	620 (90)	9	41(b)
CW67	T7X1	580 (84)	614 (89)	12	40(c)
7064(a)	T6	635 (92.1)	683 (99.0)	12	42
	T7	621 (90.0)	650 (95)	9	
Al-9052	F	380 (55.0)	450 (65.0)	13	21(c)
	F	630 (91.0)	635 (92.0)	4	41(d)

(a) Formerly called PM-64. (b) Pilot production extruded bar purchased from Alcoa. (c) Typical values. (d) High-strength experimental 23 kg (50 lb) heat made by carefully controlled processing

with subsequent dies made from ceramics, glass, or composites.

The preheated die that contains powder can be consolidated in less than 1 s in a forging press, thereby reducing thermal exposure that can coarsen RS microstructures. In addition, productivity is greatly increased by minimizing press time. Depending upon the type of fluid die material being used, the die can be machined off, chemically leached off, melted off, or designed to "pop off" the net-shape component while cooling from the consolidation temperature. For aluminum-alloy powders, unconfined degassing is critical to optimize the cost effectiveness of the ROC process. An electrodynamic degasser was developed for this purpose.

Just as in the case of HIP, the stress state during ROC is largely hydrostatic. Consequently, there may not be sufficient shear stresses to break the oxide layer and disperse the oxides, which can lead to prior powder-particle boundary (PPB) failure. Consequently, a subsequent hot-working step of the ROC billet is often necessary for demanding applications. More detailed information on this technique can be found in the article "Rapid Omnidirectional Compaction" in Volume 7 of the 9th Edition of *Metals Handbook*.

**Dynamic Compaction.** Various ultrahigh strain-rate consolidation techniques, that is, dynamic compaction, have been developed and utilized for aluminum alloys (Ref 31, 32). In dynamic compaction, a high-velocity projectile impacts the degassed powders that are consolidated by propagation of the resultant shock wave through the powder. The bonding between the powder particles is believed to occur by melting of a very thin

layer on the powder surfaces, which is caused by the heat resulting from friction between the powder particles that occurs during impact. The melted region is highly localized and self-quenched by the powder interiors shortly after impact. Thus, dynamic compaction has the advantage of minimizing thermal exposure and microstructural coarsening, while breaking up the PPBs.

**New Directions in Powder Consolidation.** Perhaps the most promising area in powder consolidation to have matured over the past decade is actually a bridge between RS-P/M technology and I/M technology. Spray-forming techniques such as the Osprey process (Ref 33), liquid dynamic compaction (Ref 34, 35), and vacuum plasma structural deposition (Ref 36) build consolidated product directly from the atomized stream. Solidification rates greater than those in conventional I/M are attained, but they are not as high as those in RS-P/M processes such as atomization or splat cooling. Spray-forming technology, which will be discussed briefly later, is also described in the article "Spray Deposition of Metal Powders" in Volume 7 of the 9th Edition of *Metals Handbook*.

## Alloy Design Research

Aluminum P/M technology is being used to improve the limitations of aluminum alloys and also to push the inherent advantages of aluminum alloys to new limits. The alloy design efforts can be described in three areas (Ref 4):

- High ambient-temperature strength with improved corrosion and SCC resistance

- Improved elevated-temperature properties so aluminum alloys can more effectively compete with titanium alloys
- Increased stiffness and/or reduced density for aluminum alloys to compete with organic composites

This third area includes aluminum-lithium alloys, aluminum-beryllium-lithium alloys, and metal-matrix composites.

## High Ambient-Temperature Strength and Improved Corrosion/SCC Resistance

**RS Alloys.** Rapid-solidification processing has been used to develop improved aluminum alloys for ambient-temperature service for over 25 years. Much of the early work has been conducted by investigators at the Aluminum Company of America (Alcoa) (Ref 37-39). The most successful work in this area was in the Al-Zn-Mg alloy subsystem (7xxx alloys), where more highly alloyed 7xxx alloy variants with dispersed transition-metal intermetallic phases were investigated. The leading alloys developed have been cobalt-containing alloys 7091 and 7090 (Ref 39), and a subsequent nickel- and zirconium-containing alloy CW67 (Ref 40). Compositions of aluminum alloys for ambient-temperature service are provided in Table 1. Mechanical properties, of course, depend upon mill product form and thermo-mechanical history. Tensile properties of extrusions are provided in Table 2 to allow an appreciation for the strengths that are possible.

The SCC resistance of 7090 and 7091 is improved with respect to I/M 7xxx alloys. In fact, SCC resistance generally increases with cobalt content from 0 to 1.6 wt%, but general corrosion resistance, as assessed by weight-loss tests, decreases (Ref 41). For example, in the peak-strength condition, alloy 7091 with 0.4 wt% Co has a similar SCC plateau crack velocity to I/M 7075, but with 1.5 wt% Co, 7090 has a lower plateau velocity (Fig. 5). Alloy CW67 also has improved combinations of strength and SCC resistance with respect to conventional 7xxx I/M alloys and it has replaced 7090 and 7091 as Alcoa's leading RS-P/M aluminum alloy for ambient-temperature service (Ref 40).

The effect of P/M processing on the fatigue behavior of aluminum alloys is complex. In general, resistance to fatigue-crack initiation is improved by P/M processing, in part because of the refinement of constituent particles at which fatigue cracks can nucleate. On the other hand, resistance to fatigue-crack growth can be decreased by P/M processing because of the refinement in grain size. For fine-grain P/M aluminum alloys, the plastic zone size may span several grains, thereby enabling the transfer of deformation across grain boundaries. In coarser-grain I/M alloys, the plastic zone may be contained within one grain.

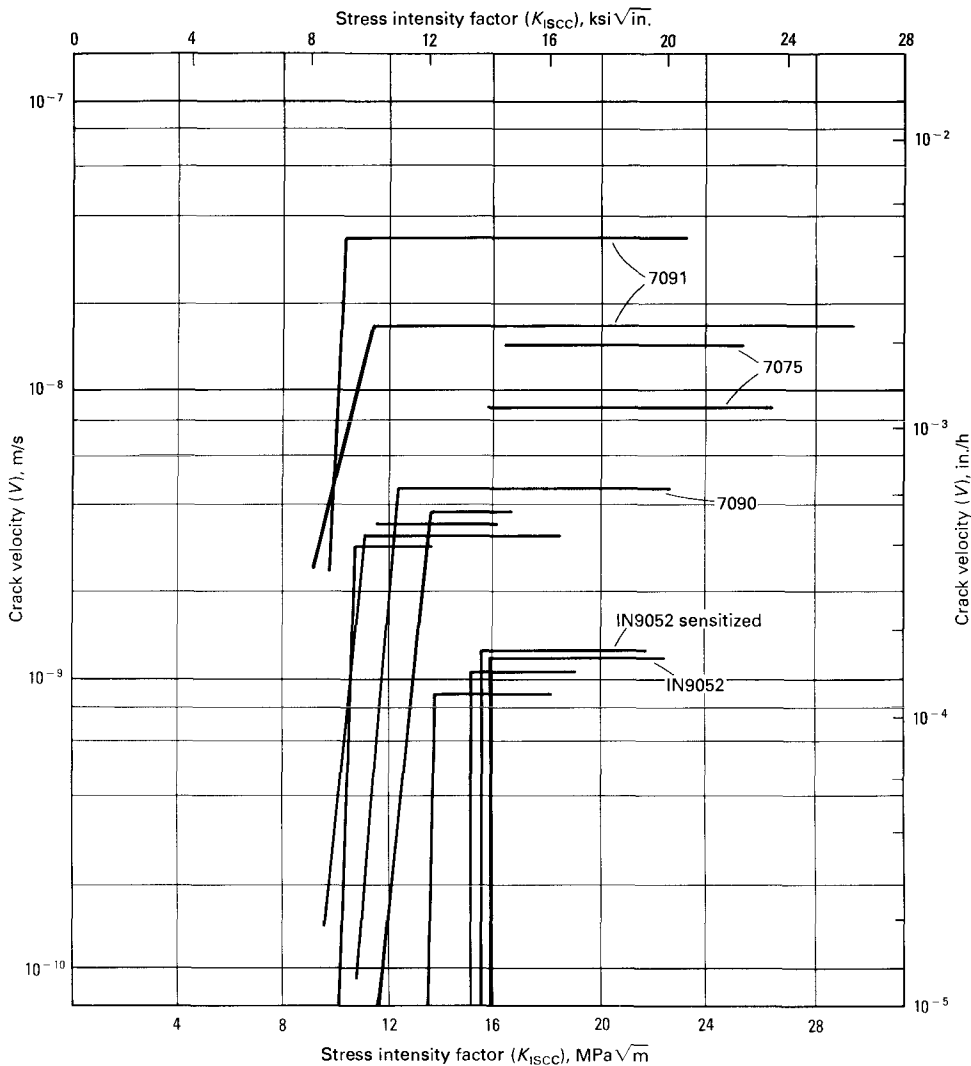


Fig. 5 Stress-corrosion cracking characteristics for the three P/M alloys and I/M 7075, all in their highest-strength (peak aged) conditions. Source: Ref 41

Voss compared the fatigue-crack growth rate under constant load amplitude of 7075-T6510 made by I/M and P/M and enhanced-purity I/M alloy of similar composition, 7475-T651 (Ref 42). Typical results are shown in Fig. 6. The I/M alloys generally displayed better resistance to fatigue-crack propagation in laboratory air. This behavior can change under spectrum fatigue testing, that is, testing designed to simulate service conditions by altering amplitude, where P/M aluminum alloys can perform better than I/M counterparts. The prospective user should exercise care in selecting the proper fatigue testing for aluminum alloys by performing tests that are relevant to service.

Scientists at Kaiser Aluminum & Chemical Corporation have also developed RS-P/M 7xxx alloys (Ref 43). The leading such alloy, 7064 (formerly called PM-64), is in part strengthened by zirconium-, chromium-, and cobalt-containing dispersoids (see Table 1). It also displays attractive combinations of strength and SCC resistance, and

in addition, has been shown to be superplastically formable.

**Wear-Resistant RS Alloys.** Japanese researchers have also been developing alloys for ambient-temperature service by RS-P/M. However, they have been more interested in wear resistance than corrosion resistance. For example, Honda Motor Company, Ltd. (Ref 44) has been developing duplex powder alloys that contain atomized aluminum-silicon or aluminum-iron "hard alloy" powders with more ductile aluminum powders (Ref 44). The wear resistance is provided by the hard alloy powders, and good forgeability is provided by the more ductile powders.

Japanese researchers at Sumitomo Electric Industries and Kobe Steel are also looking at RS-P/M Al-Si alloys for wear-resistant applications (Ref 45, 46). For example, Kobe Steel has patented an Al-Si-Cu-Mg family of alloys that contain dispersoid-forming elements for compressor pistons and connecting rods. The lead-

ing composition is an Al-20Si-5Fe-2Cu-1Mg-1Mn (wt%) alloy.

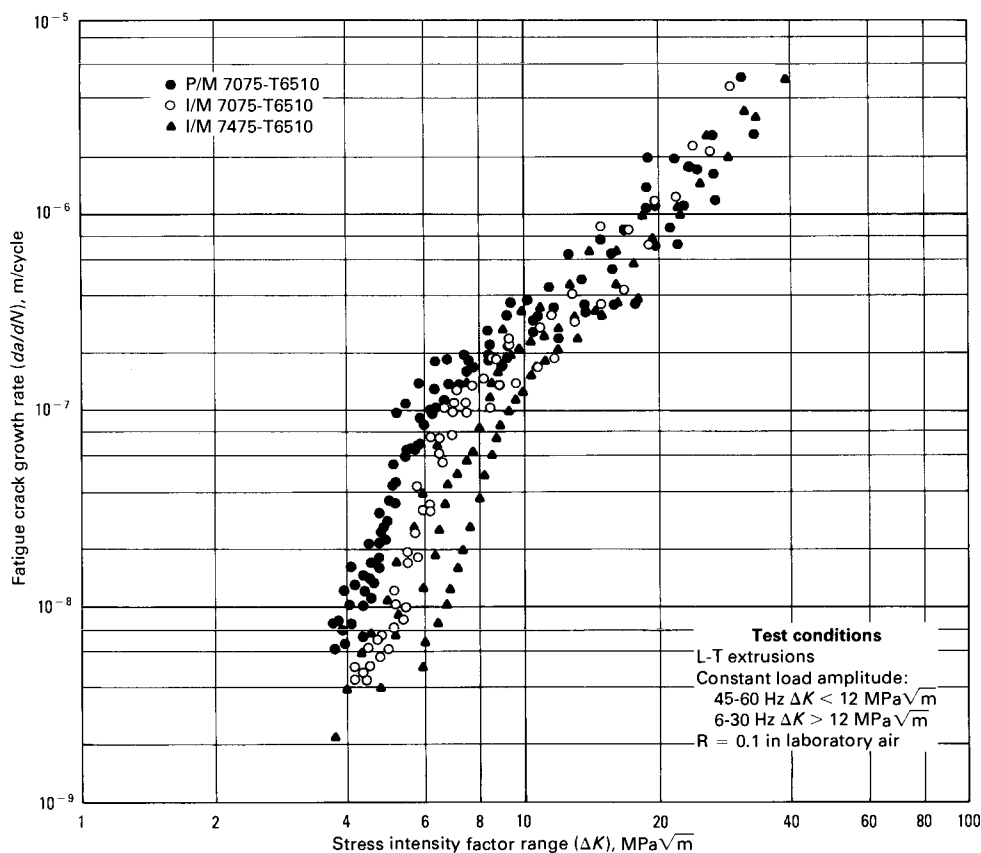
Researchers at Tokoku University in Japan have developed RS amorphous alloys that can be produced in films, fibers, plates, or pipes (Ref 47). These alloys contain various combinations of yttrium, nickel, and lanthanum, producing unprecedented strengths as high as 1140 MPa (165 ksi) on, presumably, small samples. Although it is doubtful that such alloys could be scaled up to large wrought products that have such strengths, the alloys are aimed at wear-resistant applications in pistons, valves, gears, brakes, bearings, and rotors (Ref 47).

**Mechanically Alloyed Materials.** Mechanical attrition processes have also been used to develop improved alloys for ambient-temperature service. Mechanically alloyed Al-4Mg-1.1C-0.5O alloy Al-9052 (formerly IN-9052) (Ref 19-21) and Al-4Mg-1.3Li-1.1C-0.4O alloy Al-905XL (formerly IN-905XL) (Ref 48) were designed to obtain high strength from dispersion and substructural strengthening, while minimizing corrosion and SCC susceptibility that might be introduced by certain precipitates. In addition, Al-905XL was designed to have increased specific stiffness, introduced by the lithium addition. Each of these alloys has been fabricated over a wide range of strength levels by controlling dispersoid content and thermomechanical processing to vary substructural strengthening. In general, these alloys are hot worked at relatively high values of temperature-compensated strain rate, referred to as the Zener Hollomon parameter, which is defined by:

$$Z = \dot{\epsilon} \exp(Q/RT) \quad (\text{Eq 1})$$

where  $\dot{\epsilon}$  is the mean strain rate,  $Q$  is the activation energy of the rate-controlling process in the deformation mechanism,  $R$  is the universal gas constant, and  $T$  is absolute temperature. Such strain rates lead to increased substructural strengthening. For example, extruding at a relatively low temperature and high speed produces a high  $Z$ , which can lead to a fine array of subgrains. Hot working at too high a value of  $Z$  can result in sufficient stored energy to cause recrystallization. However, the relatively high-volume fraction of finely dispersed oxides and carbides tends to stabilize the fine grain or subgrain structure and inhibit the formation of coarse recrystallized grains.

Alloy Al-9052 is dispersion strengthened by magnesium oxides, aluminum oxides, and aluminum carbide; solid-solution strengthened by magnesium; and fine grain/subgrain strengthened (Ref 19, 20). Alloy Al-905XL is strengthened by similar features, although it does display a slight artificial aging response from lithium-containing precipitates (Ref 49-51). The SCC behavior of a high-strength variant of Al-9052 was compared with RS-P/M 7091 and 7090 in their



**Fig. 6** Fatigue crack growth behavior of equivalent yield strength P/M 7075-T6510, I/M 7075-T6510, and I/M 7475-T6510 in a laboratory air environment. Source: Ref 43

peak strength (T6) conditions (Ref 41). The nonheat-treatable mechanically alloyed material displayed the lowest susceptibility in these peak strength conditions (see Fig. 5). However, the RS-P/M alloys were immune to SCC in the overaged T7 conditions, which resulted in a 10% decrease in strength. The higher cobalt-containing RS-P/M alloy 7090 displayed lower susceptibility and higher strength than 7091, although the higher-volume fraction of cobalt-containing dispersoids did increase pitting susceptibility with respect to 7091. The three P/M alloys showed clear SCC resistance advantages over competitive I/M alloys (Fig. 5). This work concluded with the following generalization concerning the SCC susceptibility of aluminum P/M alloys.

**SCC Susceptibility.** P/M alloys made by RS or by mechanical alloying achieve excellent combinations of high strength and superb SCC resistance because they enable the uniform introduction of microstructural features that either improve SCC resistance (for example,  $\text{Co}_2\text{Al}_9$  intermetallic particles), or increase strength without degrading SCC resistance (for example, oxide and carbide dispersion strengthening) (Ref 41).

Al-905XL is currently the leading mechanically alloyed material under commercialization. The alloy is primarily aimed at

forging applications where its attractive strength (Table 3), low density, and good corrosion and SCC resistance offer advantages over conventional aluminum alloys. A mechanically alloyed 2xxx alloy, Al-9021, is also under development, but its corrosion resistance is not as good as that of Al-9052 and Al-905XL.

**Concluding Remarks.** Attractive alloys have been developed by both RS-P/M and mechanical attrition processes for ambient-temperature service. Their primary advantage is improved combinations of strength and SCC/corrosion resistance. However, their advantage over high-strength aluminum I/M alloys is often viewed as not significant enough to effect alloy changes in existing applications. Furthermore, the extensive, successful efforts in developing aluminum-lithium I/M alloys has provided additional competition for these P/M alloys.

#### P/M Alloys With Improved Elevated-Temperature Properties

Powder metallurgy technology is inherently suited to alloy design for elevated-temperature service. Rapid solidification technology allows formation of finely dispersed intermetallic strengthening phases that resist coarsening and are not practical, or in some cases are not even possible, by

**Table 3** Typical mechanical properties of Al-905XL P/M forgings

Material property	Direction	
	Longitudinal	Transverse
Ultimate tensile strength		
MPa .....	517	483
ksi .....	75	70
Yield strength (0.2% offset)		
MPa .....	448	414
ksi .....	65	60
Elongation, % .....		
	9	6
Fracture toughness, $K_{Ic}$		
MPa $\sqrt{\text{m}}$ .....	30	30
ksi $\sqrt{\text{in.}}$ .....	27	27
Modulus of elasticity		
GPa .....	80	...
$10^6$ psi .....	11.6	...

Source: Ref 50

conventional I/M. In addition, mechanical attrition processes can similarly disperse intermetallics and also disperse oxides and carbides on an extremely fine scale. These oxides and carbides are extremely resistant to coarsening at the service temperatures envisioned for aluminum alloys.

Early efforts focused on extending the service temperature of aluminum alloys to 315 to 345 °C (600 to 650 °F). In the mid-1980s, a U.S. Air Force initiative sought to extend the possible service temperature to an extremely challenging 480 °C (900 °F).

**RS-P/M Alloys.** Much of the development in this area sought to disperse slow-diffusivity transition metals as intermetallic phases in aluminum. Most of the successful alloys developed contain iron.

Scientists at Alcoa (Ref 52-54) investigated numerous such RS-P/M alloys, and Al-Fe-Ce alloys CU78 and CZ42 were the leading alloys developed (compositions of elevated-temperature service alloys are given in Table 4; a typical microstructure is shown in Fig. 7). The alloys displayed good tensile strength up to 315 °C (600 °F) as indicated in Tables 5 and 6, good room-temperature properties after elevated-temperature exposure (Fig. 8), and surprisingly good resistance to environmentally assisted cracking (Ref 54). No failures were observed in 180 days in a 3.5% NaCl solution under conditions of alternate immersion at a stress of 275 MPa (40 ksi). Pratt & Whitney has found that molybdenum additions to RS-P/M Al-Fe produce attractive properties also (Ref 56).

Perhaps the leading RS-P/M aluminum alloys developed are the Al-Fe-V-Si alloys by Allied-Signal Inc. that are made using planar flow casting (Ref 57, 58). For example, alloys FVS-0812 and FVS-1212 display exceptionally high strengths at 315 °C (600 °F) and usable strengths at 425 °C (800 °F) (see Tables 7 and 8). Furthermore, the alloys contain high-volume fractions of silicides that increase modulus. Both alloys have good corrosion resistance in salt-fog environments and good resistance to SCC

**Table 4 Nominal compositions of aluminum P/M alloys for elevated-temperature service**

Alloy	Supplier	Composition, wt%(a)								
		Fe	Ce	Cr	V	Si	Zr	Mn	Mo	Ti
CU78	Alcoa	8.3	4.0	...	...	...	...	...	...	...
CZ42	Alcoa	7.0	6.0	...	...	...	...	...	...	...
...	Pratt & Whitney	8.0	...	...	...	...	...	...	2	...
FVS-0812	Allied-Signal	8.5	...	...	1.3	1.7	...	...	...	...
FVS-1212	Allied-Signal	11.7	...	...	1.2	2.4	...	...	...	...
FVS-0611	Allied-Signal	6.5	...	...	0.6	1.3	...	...	...	...
...	Alcan	...	...	5	...	...	2	...	...	...
...	Alcan	...	...	5	...	...	2	1	...	...
...	Inco	...	...	...	...	...	...	...	...	4(b)
...	Inco	...	...	...	...	...	...	...	...	8(b)
...	Inco	...	...	...	...	...	...	...	...	12(b)

(a) Oxygen also present. (b) Oxide- and carbide-dispersion strengthening also present

in saline environments. The dispersed silicides are so fine that they contribute greatly to strength over a wide temperature range (Fig. 9), but do not degrade corrosion and SCC resistance. The silicides are apparently so stable that alloy FVS-0812 can be exposed to 425 °C (800 °F) for up to 1000 h without degradation of room-temperature properties.

Alloy FVS-0812 has been forged into aircraft wheels and is now under consideration as a replacement for conventional alloy 2014, the mainstay aircraft wheel alloy. In addition, the stability of the alloy after elevated-temperature exposure, coupled with its high strength at 300 °C (575 °F), is enabling the alloy to challenge titanium alloys in certain engine components.

Allied-Signal, Inc. is also developing RS-P/M alloy FVS-0611 for applications where good formability is necessary. It is lower in alloying content (Table 4) than the other PFC alloys for elevated-temperature service and is aimed for applications such as rivets and thin-walled tubing.

Scientists at Alcan International Limited took a different approach to designing RS-P/M alloys for elevated-temperature service (Ref 59). They pursued the Al-Cr-Zr system to develop an alloy that could attain attractive elevated-temperature properties from powder that is less sensitive to cooling rate and also has better hot workability. Both chromium and zirconium can produce thermally stable solid solutions in aluminum at

modest solidification rates (10<sup>3</sup> K/s). The consolidated billet could then be extruded or rolled with a relatively low flow stress at the lower end of the hot-working range and then aged at a higher temperature to form stable intermetallics that provide elevated-temperature strength.

This alloy design approach provides a potential advantage over RS-P/M Al-Fe-X alloys, which often attain their highest hardness just after RS (see Fig. 10). Hardness, and consequently yield strength, can be lost in each thermomechanical processing step from degassing/consolidation through hot working. In addition, elevated-temperature flow stress of the as-compacted billet can make extrusion, forging, and rolling difficult and expensive. Another advantage over RS-P/M Al-Fe-X alloys is that the leading Al-Cr-Zr compositions have lower density.

Palmer *et al.* (Ref 59) present analysis by Ekvall *et al.* (Ref 60) and other data comparing the modulus and density of various RS-P/M alloys for elevated-temperature service (Table 9). They compare two parameters that are related to weight-savings ability—specific modulus  $E \div \rho$  and the parameter  $E^{1/3} \div \rho$ . Specific modulus is the weight-controlling parameter where the predominant failure mode involves aeroelastic stiffness, and the parameter  $E^{1/3} \div \rho$  applies in certain buckling-limited applications. The leading Al-Cr-Zr alloys, Al-5Cr-2Zr and Al-5Cr-2Zr-1Mn, compare favorably to other RS-P/M alloys and have a substantial ad-

**Table 5 Modified property goals (minimum values) for shaped extrusions of P/M alloy CZ42**

Material property	Value
<b>Tensile strength, MPa (ksi)</b>	
At room temperature	448 (65)
At 166 °C (330 °F)	365 (53)
At 232 °C (450 °F)	310 (45)
At 260 °C (500 °F)	283 (41)
At 316 °C (600 °F)	221 (32)
<b>Yield strength, MPa (ksi)</b>	
At room temperature	379 (55)
At 166 °C (330 °F)	345 (50)
At 232 °C (450 °F)	296 (43)
At 260 °C (500 °F)	262 (38)
At 316 °C (600 °F)	200 (29)
<b>Modulus of elasticity, GPa (10<sup>6</sup> psi)</b>	
At room temperature	78.6 (11.4)
At 166 °C (330 °F)	68.9 (10.0)
At 232 °C (450 °F)	64.1 (9.3)
At 260 °C (500 °F)	62.1 (9.0)
At 316 °C (600 °F)	56.5 (8.2)
<b>Elongation, at room temperature to 316 °C (600 °F), %</b>	
	5.0
<b>Fracture toughness (<math>K_{Ic}</math>), MPa√m (ksi√in.)</b>	
L-T at room temperature	23 (21)
T-L at room temperature	18 (16)

Source: Ref 54

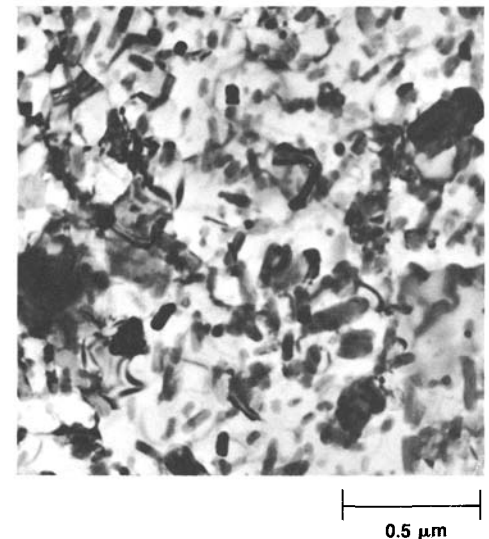
vantage over leading I/M alloys such as 2219.

**Mechanically Attrited Alloys.** Reaction milling is an SAP-derived process developed by Jangg (Ref 22, 23) and co-workers in Europe. It is claimed to be an attritor milling process that mills without a process control agent (PCA) (Ref 24), such as the stearic acid often used in mechanical alloying. Reaction milling is often performed in the presence of lampblack, or graphite, the latter of which the author has found to act as a PCA (Ref 19, 20) that introduces carbon. Oxygen is introduced by careful con-

**Table 6 Tensile and fracture properties of Al-Fe-X alloy thin-sheet specimens tested in the L-T orientation**

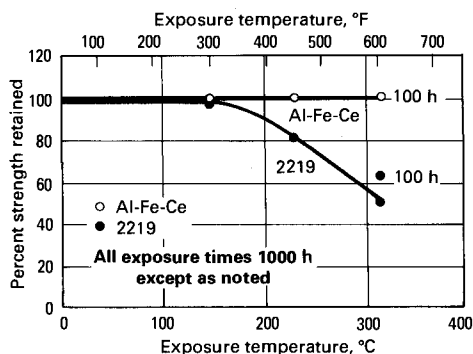
Material	Temperature		Yield strength		Tensile strength		Elongation, %	Fracture toughness ( $K_{Ic}$ )	
	°C	°F	MPa	ksi	MPa	ksi		MPa√m	ksi√in.
Al-8Fe-7Ce	25	77	418.9	60.8	484.9	70.3	7.0	8.5	7.7
	316	600	178.1	25.8	193.8	28.1	7.6	7.9	7.2
Al-8Fe-2Mo-1V	25	77	323.5	46.9	406.6	59.0	6.7	9.0	8.2
	316	600	170.0	24.6	187.5	27.2	7.2	8.1	7.4
Al-10.5Fe-2.5V	25	77	464.1	67.3	524.5	76.1	4.0	5.7	5.2
	316	600	206.3	29.9	240.0	34.8	6.9	8.1	7.4
Al-8Fe-1.4V-1.7Si	25	77	362.5	52.6	418.8	60.7	6.0	36.4	33.1
	316	600	184.4	26.7	193.8	28.1	8.0	14.9	13.6

Note: Tensile and compact-tension (CT) specimens were prepared according to ASTM specifications E399 and E813. The tensile specimens were 203 mm (8.0 in.) in total length, 50.8 mm (2 in.) in gage length, 12.7 mm (0.5 in.) in width, and 1.27 mm (0.05 in.) in thickness. The CT specimens were 38.1 mm (1½ in.) in width and 7.6 to 10.2 mm (0.3 to 0.4 in.) in thickness. Source: Ref 55



**Fig. 7** TEM micrograph of an RS-P/M Al-8Fe-4Ce alloy showing fine Al-Fe-Ce phases that strengthen the 1 mm (0.040 in.) thick sheet. Courtesy of G.J. Hildeman and L. Angers, Alcoa Laboratories



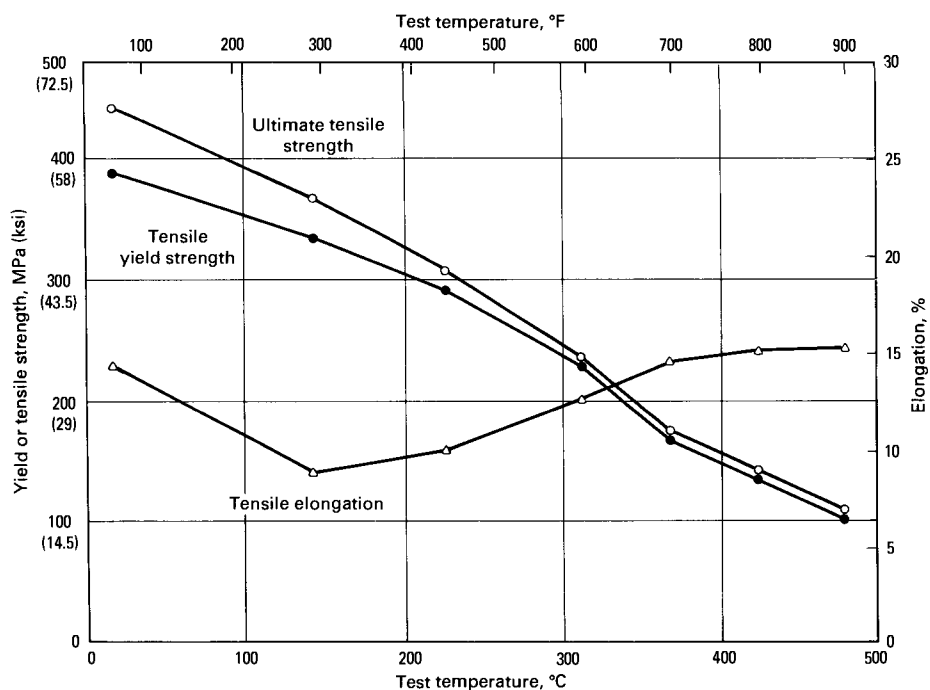


**Fig. 8** Percent of room-temperature strength retained after elevated-temperature exposure of Al-Fe-Ce alloys

trol of the milling atmosphere and forms oxides that are finely dispersed by the reaction milling process. Oxygen in the milling atmosphere may be considered a PCA by reducing the degree of cold welding when powder particles are impacted during ball milling.

The leading reaction-milled alloys (DISPAL alloys) are not as strong as most other elevated-temperature P/M alloys at ambient or intermediate temperatures (Ref 24). However, their decrease in strength with increasing temperature is rather flat, so they have reasonable strength and also good stability at higher temperatures, for example, 150 MPa (22 ksi) tensile strength at 400 °C (750 °F).

Several mechanically alloyed aluminum alloys for elevated-temperature service were designed in 1978 by the author while working for the International Nickel Company (now Inco Alloys International, Inc.) (Ref 61). Various low-solid solubility transition-metal elements were mechanically alloyed to potentially form stable dispersoids, as were electropositive solid-solution elements (for example, lithium and magnesium) to potentially form stable ox-



**Fig. 9** Tensile properties of FVS-0812 alloy extrusions

ides. The leading alloy developed in the program was an Al-4Ti alloy that contained carbon and oxygen from the mechanical alloying process. A yield strength of 159 MPa (23 ksi) and a tensile strength of 186 MPa (27 ksi) were attained at 345 °C (650 °F) on the first heat investigated. In addition, room-temperature strengths (325 MPa, or 47 ksi yield strength and 383 MPa, or 56 ksi ultimate tensile strengths) were essentially unchanged after 100 h at 345 °C (650 °F). In recent work in the mechanically alloyed Al-Ti system by Wilsdorf *et al.* (Ref 62), higher titanium levels and yttria additions were investigated that improved strength at 345 °C (650 °F). In addition, yield and tensile strengths at 425

°C (800 °F), as high as 113 MPa (16 ksi) and 133 MPa (19 ksi), respectively, were attained.

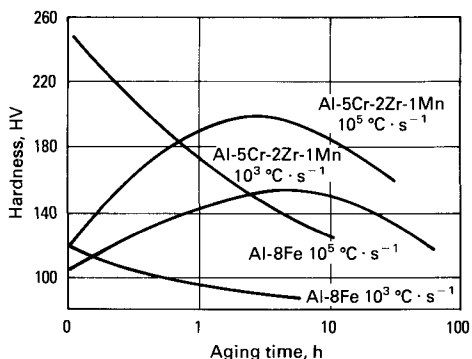
**Concluding Remarks.** This is the area where aluminum P/M has perhaps the best chance for replacing conventional aluminum or titanium alloys. In fact, the attractive room-temperature properties of these alloys and their surprisingly good corrosion resistance will enable them to compete for warm- and ambient-temperature applications. RS-P/M alloys generally attain higher strength than those made by mechanical attrition processes up to about 350 °C (660 °F), with the Allied-Signal Inc. FVS series alloys showing particular promise. The mechanically attrited materials display superb stability—that is, room-temperature strength after elevated-temperature exposure—and high strengths for aluminum alloys at 425 °C (800 °F).

**Table 7** Room- and elevated-temperature tensile properties of planar flow cast alloy FVS-0812

Test temperature °C	°F	Yield strength		Ultimate tensile strength		Elongation, %	Young's modulus	
		MPa	ksi	MPa	ksi		GPa	10 <sup>6</sup> psi
24	75	413	60	462	67	12.9	88.4	12.8
149	300	345	50	379	55	7.2	83.2	12.0
232	450	310	45	338	49	8.2	73.1	10.6
316	600	255	37	276	40	11.9	65.5	9.5
427	800	138	20	155	22	15.1	61.4	8.9

**Table 8** Mechanical properties of planar flow cast alloy FVS-1212

Test temperature °C	°F	Yield strength		Ultimate tensile strength		Elongation, %	Young's modulus	
		MPa	ksi	MPa	ksi		GPa	10 <sup>6</sup> psi
24	75	531	77	559	81	7.2	95.5	13.9
150	300	455	66	469	68	4.2	...	...
230	450	393	57	407	59	6.0	...	...
315	600	297	43	303	44	6.8	...	...



**Fig. 10** Variation of microhardness with aging time at 400 °C for Al-5.2Cr-1.9Zr-1Mn alloy compared to Al-8Fe alloy. Source: Ref 59

**Table 9 Elastic modulus, density, and weight-saving parameters for thermally stable RS-P/M aluminum alloys**

Alloy designation	Composition	Elastic modulus ( <i>E</i> )		Density ( $\rho$ ), g/cm <sup>3</sup>	<i>E</i> / $\rho$ MN · m · kg <sup>-1</sup>	<i>E</i> <sup>1/3</sup> / $\rho$ N <sup>1/3</sup> m <sup>-1/3</sup> kg <sup>-1</sup>
		GPa	10 <sup>6</sup> psi			
.....	Al-5Cr-2Zr	80.8	11.72	2.82	28.7	1.53
.....	Al-5Cr-2Zr-1Mn	86.5	12.54	2.86	30.2	1.55
.....	CU78	79.6	11.54	2.95	27.0	1.46
.....	CZ42	80.0	11.60	3.01	26.6	1.43
.....	Al-8Fe-2Mo	86.2	12.50	2.91	29.6	1.52
.....	FVS-0812	88.4	12.82	3.02	29.3	1.48
.....	FVS-1212	95.5	13.85	3.07	31.1	1.49
.....	RAE 72	89.0	12.91	2.89	30.8	1.54
.....	I/M 2219	73.0	10.59	2.86	25.5	1.46

Note: *E*/ $\rho$ , specific modulus; *E*<sup>1/3</sup>/ $\rho$ , buckling parameter. Source: Ref 60

**Table 10 Properties of RS-P/M aluminum-beryllium-lithium alloys**

Composition, wt%	Temper	Yield strength		Ultimate tensile strength		Elongation, %	Young's modulus		Density g/cm <sup>3</sup>
		MPa	ksi	MPa	ksi		GPa	10 <sup>6</sup> psi	
Al-20.5Be-2.4Li	As-quenched	321	46.5	451	65	6.4	123	18	2.298
	Peak aged	483	70	531	77	3.3	123	18	2.298
Al-29.6Be-1.3Li	Underaged	434	63	494	72	5.0	142	21	...
	Peak aged	497	72	536	78	2.6	142	21	...

Source: Ref 66

### High-Modulus and/or Low-Density Alloys

Aluminum P/M is useful to attain stiffer and/or lighter alloys in several ways. First, the alloying limits of elements that increase modulus (*E*) or decrease density ( $\rho$ ) can be extended by either RS or mechanical attrition processes. Lithium and beryllium are the only elements that generally do both in aluminum, and RS aluminum-lithium and aluminum-lithium-beryllium alloys have been made that contain alloying levels not practical by I/M. In addition, attempts to mechanically alloy aluminum-lithium alloys have been fruitful at modest lithium levels, but not so at very high levels (3.5 to 4 wt%) (Ref 48). Powder metallurgy technology is also useful in producing metal-matrix composites (MMCs).

The development of P/M alloys that have increased stiffness and/or decreased density is not receiving as much emphasis as it did less than a decade ago (early 1980s). This is largely due to the success of aluminum-lithium alloys made by I/M, which have the advantage of being producible using modified conventional equipment (see the article "Aluminum-Lithium Alloys" in this Volume). In addition, several technologies have arisen that enable MMCs to be made by I/M—for example, Alcan Aluminum Limited's Dural technology for SiC in aluminum, and Martin Marietta's XD technology for TiB<sub>2</sub> or TiC in aluminum. Several high *E* ÷  $\rho$  aluminum P/M alloys will be briefly described to show how P/M technology can be utilized in this area.

**Al-Li and Al-Li-Be Alloys.** Scientists at Allied-Signal Inc. investigated 3 to 4 wt% Li aluminum-lithium alloys made by the PFC

technique. The rapid cooling rate allows lithium levels to be attained that are higher than those practical by conventional I/M using commercial direct-chill casting technology. The practical limit for lithium in aluminum by commercial-scale direct-chill casting is about 2.7 wt%. The decreased thermal conductivity caused by lithium additions results in slower cooling rates, which lead to unacceptable segregation at high lithium levels. In addition, the potential explosive hazard increases with lithium content.

Several of the high-lithium-containing PFC alloy compositions were assessed in an Air Force program in cooperation with The Boeing Company (Ref 63). The leading such alloy, 644B, is an Al-3.2Li-1Cu-0.5Mg-0.5Zr alloy with a density of 2.537 g/cm<sup>3</sup>. The high zirconium content, which is over 3 times the effective maximum (about 0.14 wt%) in commercial-scale direct-chill casting, was added to form composite Al<sub>3</sub>Zr-Al<sub>3</sub>Li precipitates that would hopefully be less shearable than Al<sub>3</sub>Li ( $\delta'$ ).

The alloy attained a mean yield strength of 424 MPa (62 ksi), tensile strength of 521 MPa (76 ksi), with 7% elongation in the longitudinal direction in extruded form. Unfortunately, toughness was low, so subsequent alloy variations contained lower lithium content (Ref 64). The alloy also displayed SCC susceptibility.

As mentioned earlier, mechanically alloyed Al-905XL attains moderate to high strength and good corrosion resistance. The 1.3 wt% Li in the alloy coupled with lithium-, magnesium-, and aluminum-containing oxides also increases elastic modulus, which is 80 GPa (11.6 × 10<sup>6</sup> psi). Lithium levels of up to 4% were investigated in the

early development of mechanically alloyed aluminum-lithium alloys (Ref 48), but oxygen contamination caused ductility to be low. More recently, various other lithium-containing compositions were integrated into the aforementioned Air Force program (Ref 63).

As pointed out by Perepezko, the aluminum-beryllium alloy system is one that can benefit greatly from the RS-P/M approach (Ref 65). The solid-solubility limit of beryllium in aluminum is low (0.3 at%), as is the eutectic composition (2.5 wt%), so the extension of solubility by RS-P/M techniques is particularly useful.

Lewis and co-workers at Lockheed Corporation have examined Al-Be-Li alloys made by RS-P/M and have attained values of specific modulus that are significantly higher than those attained in conventional aluminum alloys (Ref 66). They used melt spinning to examine Al-Be-Li and Al-Be alloys designed to be age hardenable and to have ultralow density. The alloys were designed to be strengthened by  $\delta'$  and possibly  $\alpha$ -Be dispersoids. The ability of RS-P/M technology to extend solubility limits can be appreciated by the Al-20.5Be-2.4Li and Al-29.6Be-1.3Li alloys that displayed perhaps the best properties (Table 10).

**Metal-Matrix Composites.** A significant amount of work has been undertaken to produce aluminum MMCs using P/M technology (Ref 2-6). Much of this work has emphasized SiC for reinforcement, although work with Al<sub>2</sub>O<sub>3</sub>, TiB<sub>2</sub>, and other ceramics has received attention. P/M technology enables unique matrix alloys to be designed to accept specific ceramic particles. The alloy designer has the flexibility to select compositions that might not be practical by I/M to improve wettability and/or the nature of the ceramic matrix interface. In addition, P/M allows fine-scale mixing of ceramic and alloy powder (or powders for duplex matrices) and provides the option of "slushy state" consolidation, that is, at a consolidation temperature between the solidus and liquidus of the matrix alloy. DWA Corporation and Arco Metals (formerly Silag Corporation) have extended significant effort in this area throughout the past decade and each produced alloys with high strength and stiffness. Some of the more interesting aluminum MMC work has involved reinforcing high-strength P/M alloy matrices. For example, Agarwala *et al.* examined 7091-SiC composites and found attractive mechanical properties, but decreased corrosion resistance (Ref 67). Unfortunately, no one alloy has gained widespread commercial acceptance.

Mechanical alloying can also be used to make MMCs. A low-volume fraction of Al<sub>2</sub>O<sub>3</sub> can be easily dispersed by the attrition process (Ref 19, 20). More recently, Nieh *et al.* demonstrated that 15 vol% SiC could be introduced into Al-4Cu-1.5Mg-1.1C-0.80 al-

loy Al-9021 and produce an MMC with a modulus of 81.8 GPa ( $11.8 \times 10^6$  psi) (Ref 68).

It appears that MMC research by RS-P/M is losing favor to spray-deposition forming technologies, which, as mentioned earlier, are a bridge between P/M and I/M.

**Spray Deposition.** In spray-deposition processes, the atomized stream of alloyed particles is deposited onto a chilled substrate and builds up a compact directly. With proper control of atmosphere, this obviates degassing and consolidation, thereby reducing costs. For more demanding applications, a subsequent hot-working step is often necessary. The as-sprayed compact can be hot worked conventionally into useful shapes that exhibit improved ductility. The details of the leading such processes, Osprey (Ref 33) and liquid dynamic compaction (LDC) (Ref 34, 35, 69), have been described in detail.

Very high strengths have been obtained on LDC-processed 7075 containing 1% Ni and 0.8% Zr. For extrusions, an ultimate tensile strength of 816 MPa (118 ksi) and yield strength of 740 MPa (107 ksi) with 9% elongation have been achieved. More recently, preliminary LDC work on the Al-4-6.3Cu-1.3Li-0.4Ag-0.4Mg-0.14Zr alloy (Weldalite 049)—which was designed to be a greater than 700 MPa (102 ksi) tensile strength I/M alloy (Ref 70)—showed extremely high ductility in very underaged tempers by refinement of constituent particles (42% elongation and 484 MPa, or 70 ksi, tensile strength).

The vacuum plasma structural deposition process is currently being explored by Pechiney Aluminum Company to produce properties equivalent to those in isothermally forged products. The process is being used for titanium alloys and titanium aluminum matrix systems aimed for elevated-temperature applications.

Scientists at Alcan Aluminum Limited have combined Osprey Corporation technology with proprietary ceramic powder-spray technology to make MMCs (Ref 71). They call their technology COSPRAY, and have introduced SiC or B<sub>4</sub>C into aluminum-lithium matrix alloy 8090 and I/M alloy 2618. The strength of the 8090 + 10 vol% SiC was slightly higher than that of unreinforced 8090, but ductility decreased. However, modulus increased from 79.5 to 95.9 GPa ( $11.5$  to  $14 \times 10^6$  psi). Similar behavior was observed in 2618 + 14 vol% SiC, but with a greater strength increment. Unfortunately, the B<sub>4</sub>C reacted with aluminum, and possibly with lithium, and degraded mechanical properties.

**Concluding Remarks.** Aluminum-lithium I/M alloys and new techniques of making MMCs by I/M are now competing with high  $E \div \rho$  alloys made by P/M. It is likely that research and development resources will be decreased in this area. However, spray-forming technologies like Osprey and liquid

dynamic compaction will probably receive increased research and development resources and may be cost competitive for near net-shape components.

## New Directions in Aluminum P/M Research

**Intermetallics.** The need for improved elevated-temperature properties over the range 300 to 1500 °C (570 to 2730 °F) has generated considerable interest in ordered intermetallics. With the inherent low density of aluminum-rich aluminide intermetallics, and some promising compositions that have cubic structures, it is likely that RS and mechanical attrition processes will be extended to aluminides to an increasing extent (see the article "Ordered Intermetallics" in this Volume). In addition, the RS-P/M aluminum alloys have a good

chance to displace conventional aluminum I/M alloys, and perhaps titanium alloys in certain applications because the elevated-temperature properties required by designers likely cannot be attained by I/M. Combinations of P/M processes will likely improve properties such as mechanically alloying RS powder or attrited PFC ribbon, both with and without ceramic reinforcement (Ref 72, 73).

**Superplastic forming (SPF)** is a process attractive for reducing costs in aerospace structures (Ref 74, 75). Superplasticity is generally attainable with a fine array of high-angle grain boundaries that are stabilized by fine dispersoids. Such microstructures have been attained readily by P/M, both with and without ceramic reinforcement. It is likely that research and development in this promising area will continue.

# Appendix: Conventionally Pressed and Sintered Aluminum P/M Alloys

CONVENTIONALLY PRESSED AND SINTERED aluminum powder metal parts have been commercially available for many years. Sintered aluminum P/M parts are competitive with many aluminum castings, extrusions, and screw machine products that require expensive and time-consuming finishing operations. In addition, sintered aluminum P/M parts compete with other metal powder parts in applications where some of the attractive physical and mechanical properties of aluminum can be used.

Commercially available aluminum powder alloy compositions (Table 11) consist of blends of atomized aluminum powders mixed with powders of various alloying elements such as zinc, copper, magnesium, and silicon. The most common heat-treatable grades are comparable to the 2xxx and 6xxx series wrought aluminum alloys. Alloys 201AB and MD-24 are most similar to wrought alloy 2014. They develop high strength and offer moderate corrosion resistance. Alloys 601AB and MD-69 are similar to wrought alloy 6061. These alloys offer high strength, good ductility, corrosion resistance, and can be specified for anodized parts. Alloy 601AC is the same as 601AB, but does not contain an admixed lubricant. It is used for isostatic and die-wall-lubricated compaction. When high conductivity is required, alloy 602AB often is used. Conductivity of 602AB ranges from  $24 \times 10^6$  to  $28 \times 10^6$  S/m (42.0 to 49% IACS), depending on the type of heat treatment selected.

## Aluminum P/M Part Processing

Basic design details for aluminum P/M parts involve the same manufacturing operations, equipment, and tooling that are used for iron, copper, and other metal-powder compositions. Detailed information on P/M design and processing can be found in *Powder Metallurgy*, Volume 7 of the 9th Edition of *Metals Handbook*.

**Compacting.** Aluminum P/M parts are compacted at low pressures and are adaptable to all types of compacting equipment. The pressure density curve, which compares the compacting characteristics of aluminum with other metal powders, indicates that aluminum is simpler to compact. Figure 11 shows the relative difference in compacting characteristics for aluminum and sponge iron or copper.

The lower compacting pressures required for aluminum permit wider use of existing presses. Depending on the press, a larger part often can be made by taking advantage

**Table 11** Compositions of typical aluminum P/M alloy powders

Grade	Composition, %				Lubricant
	Cu	Mg	Si	Al	
601AB	0.25	1.0	0.6	bal	1.5
201AB	4.4	0.5	0.8	bal	1.5
602AB	...	0.6	0.4	bal	1.5
202AB	4.0	...	...	bal	1.5
MD-22	2.0	1.0	0.3	bal	1.5
MD-24	4.4	0.5	0.9	bal	1.5
MD-69	0.25	1.0	0.6	bal	1.5
MD-76	1.6	2.5	...	bal	1.5

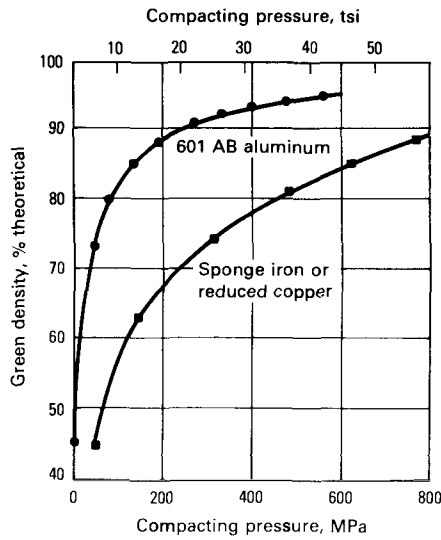


Fig. 11 Relationship of green density and compacting pressure

of maximum press force. For example, a part with a 130 cm<sup>2</sup> (20 in.<sup>2</sup>) surface area and 50 mm (2 in.) depth is formed readily on a 4450 kN (500 ton) press. The same part in iron would require a 5340 kN (600 ton) press. In addition, because aluminum responds better to compacting and moves more readily in the die, more complex shapes having more precise and finer detail can be produced.

**Sintering.** Aluminum P/M parts can be sintered in a controlled, inert atmosphere or in vacuum. Sintering temperatures are based on alloy composition and generally range from 595 to 625 °C (1100 to 1160 °F). Sintering time varies from 10 to 30 min. Nitrogen, dissociated ammonia, hydrogen, argon, and vacuum have been used for sintering aluminum; however, nitrogen is preferred because it results in high as-sintered mechanical properties (Table 12). It is also economical in bulk quantities. If a protective atmosphere is used, a dew point of -40 °C (-40 °F) or below is recommended. This is equivalent to a moisture content of 120 mL/m<sup>3</sup> (120 ppm) maximum.

Aluminum preforms can be sintered in batch furnaces or continuous radiant tube mesh or cast belt furnaces. Optimum dimensional control is best attained by maintaining furnace temperature at ±2.8 °C (±5 °F). Typical heating cycles for aluminum parts sintered in various furnaces are illustrated in Fig. 12.

Mechanical properties are directly affected by thermal treatment. All compositions respond to solution heat treating, quenching, and aging in the same manner as conventional heat-treatable alloys. More detailed information on sintering of aluminum can be found in the article "Production Sintering Practices for P/M Materials" in Volume 7 of the 9th Edition of *Metals Handbook*.

**Re-Pressing.** The density of sintered compacts may be increased by re-pressing. When re-pressing is performed primarily to improve the dimensional accuracy of a compact, it usually is termed "sizing"; when performed to improve configuration, it is termed "coining." Re-pressing may be followed by resintering, which relieves stress due to cold work in re-pressing and may further consolidate the compact. By pressing and sintering only, parts of over 80% theoretical density can be produced. By re-pressing, with or without resintering, parts of 90% theoretical density or more can be produced. The density attainable is limited by the size and shape of the compact.

**Forging** of aluminum is a well-established technology. Wrought aluminum alloys have been forged into a variety of forms, from small gears to large aircraft structures, for many years (see the article "Forging of Aluminum Alloys" in *Forming and Forging*, Volume 14 of the 9th Edition of *Metals Handbook*). Aluminum lends itself to the forging of P/M preforms to produce structural parts.

In forging of aluminum preforms, the sintered aluminum part is coated with a graphite lubricant to permit proper metal flow during forging. The part is either hot or cold forged; hot forging at 300 to 450 °C (575 to 850 °F) is recommended for parts requiring critical die fill. Forging pressure usually does not exceed 345 MPa (50 ksi). Forging

Table 12 Typical properties of nitrogen-sintered aluminum P/M alloys

Alloy	Compacting pressure		Green density %	Green density g/cm <sup>3</sup>	Green strength		Sintered density %	Sintered density g/cm <sup>3</sup>	Temper	Tensile strength(a)		Yield strength(a)		Elongation, %	Hardness	
	MPa	tsi			MPa	psi				MPa	ksi	MPa	ksi			
601AB	96	7	85	2.29	3.1	450	91.1	2.45	T1	110	16	48	7	6	55-60 HRH	
									T4	141	20.5	96	14	5	80-85 HRH	
									T6	183	26.5	176	25.5	1	70-75 HRE	
	165	12	90	2.42	6.55	950	93.7	2.52	T1	139	20.1	88	12.7	5	60-65 HRH	
									T4	172	24.9	114	16.6	5	80-85 HRH	
									T6	232	33.6	224	32.5	2	75-80 HRE	
	345	25	95	2.55	10.4	1500	96.0	2.58	T1	145	21	94	13.7	6	65-70 HRH	
									T4	176	25.6	117	17	6	85-90 HRH	
									T6	238	34.5	230	33.4	2	80-85 HRE	
602AB	165	12	90	2.42	6.55	950	93.0	2.55	T1	121	17.5	59	8.5	9	55-60 HRH	
									T4	121	17.5	62	9	7	65-70 HRH	
									T6	179	26	169	24.5	2	55-60 HRE	
	345	25	95	2.55	10.4	1500	96.0	2.58	T1	131	19	62	9	9	55-60 HRH	
									T4	134	19.5	65	9.5	10	70-75 HRH	
									T6	186	27	172	25	3	65-70 HRE	
	201AB	110	8	85	2.36	4.2	600	91.0	2.53	T1	169	24.5	145	24	2	60-65 HRE
										T4	210	30.5	179	26	3	70-75 HRE
										T6	248	36	248	36	0	80-85 HRE
180		13	90	2.50	8.3	1200	92.9	2.58	T1	201	29.2	170	24.6	3	70-75 HRE	
									T4	245	35.6	205	29.8	3.5	75-80 HRE	
									T6	323	46.8	322	46.7	0.5	85-90 HRE	
413		30	95	2.64	13.8	2000	97.0	2.70	T1	209	30.3	181	26.2	3	70-75 HRE	
									T4	262	38	214	31	5	80-85 HRE	
									T6	332	48.1	327	47.5	2	90-95 HRE	
202AB	180	13	90	2.49	5.4	780	92.4	2.56	T1	160	23.2	75	10.9	10	55-60 HRH	
									T4	194	28.2	119	17.2	8	70-75 HRH	
									T6	227	33	147	21.3	7.3	45-50 HRE	
	Cold-formed parts (19% strain)	180	13	90	2.49	5.4	780	92.4	2.56	T2	238	33.9	216	31.4	2.3	80 HRE
										T4	236	34.3	148	21.5	8	70 HRE
										T6	274	39.8	173	25.1	8.7	85 HRE
										T8	280	40.6	250	36.2	3	87 HRE

(a) Tensile properties determined using powder metal flat tension bar (MPIF standard 10-63), sintered 15 min at 620 °C (1150 °F) in nitrogen

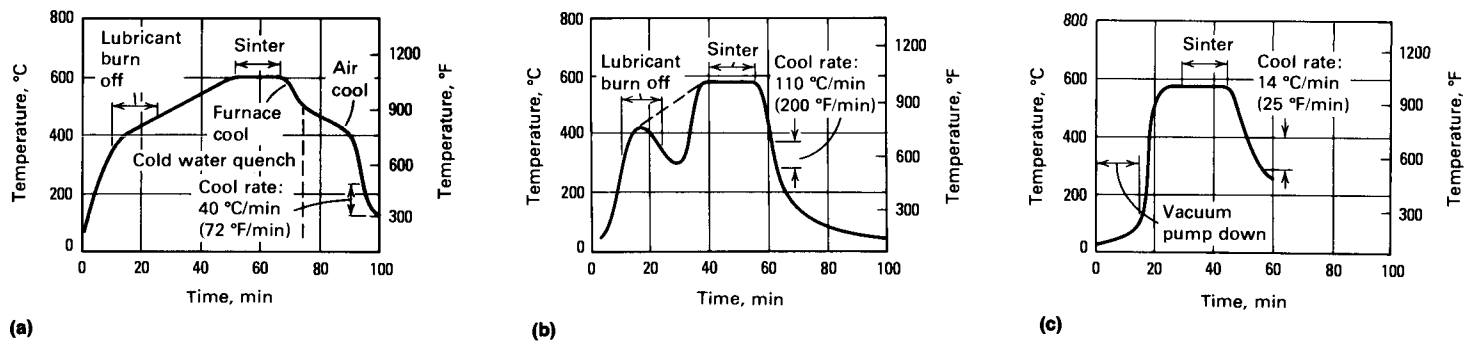


Fig. 12 Typical heating cycles for aluminum P/M parts sintered in (a) A batch furnace. (b) A continuous furnace. (c) A vacuum furnace

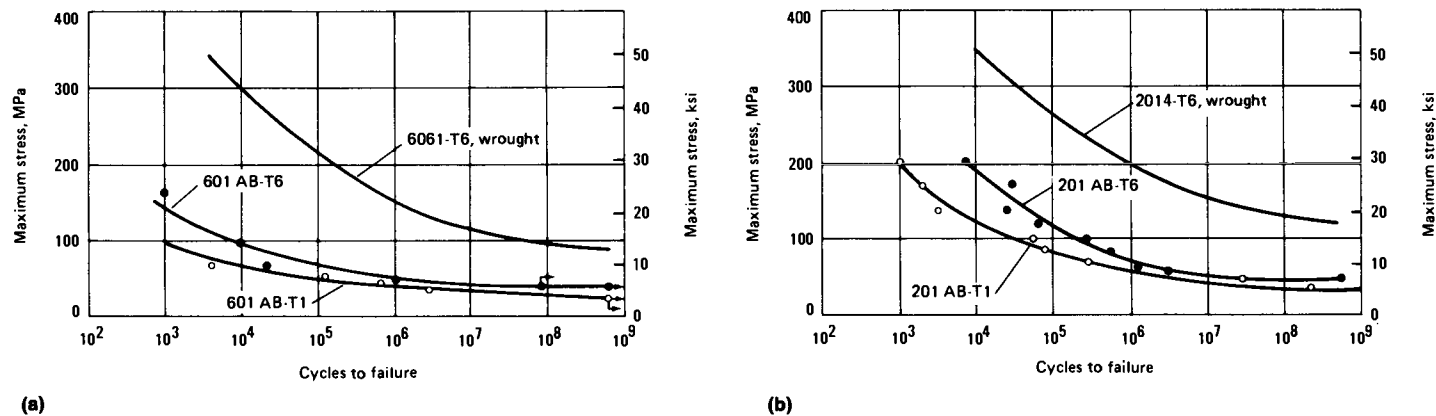


Fig. 13 Fatigue curves for (a) P/M 601AB. (b) P/M 201AB

normally is performed in a confined die so that no flash is produced and only densification and lateral flow result from the forging step. Scrap loss is less than 10% compared to conventional forging, which approaches 50%. Forged aluminum P/M parts have densities of over 99.5% of theoretical density. Strengths are higher than nonforged P/M parts, and in many ways, are similar to conventional forging. Fatigue endurance limit is doubled over that of nonforged P/M parts.

Alloys 601AB, 602AB, 201AB, and 202AB are designed for forgings. Alloy

202AB is especially well suited for cold forging. All of the aluminum powder alloys respond to strain hardening and precipitation hardening, providing a wide range of properties. For example, hot forging of alloy 601AB-T4 at 425 °C (800 °F) followed by heat treatment gives ultimate tensile strengths of 221 to 262 MPa (32 to 38 ksi), and a yield strength of 138 MPa (20 ksi), with 6 to 16% elongation in 25 mm (1 in.).

Heat treated to the T6 condition, 601AB has ultimate tensile strengths of 303 to 345 MPa (44 to 50 ksi). Yield strength is 303 to

317 MPa (44 to 46 ksi), with up to 8% elongation. Forming pressure and percentage of reduction during forging influence final properties.

Ultimate tensile strengths of 358 to 400 MPa (52 to 58 ksi), and yield strengths of 255 to 262 MPa (37 to 38 ksi), with 8 to 18% elongation, are possible with 201AB heat treated to the T4 condition. When heat treated to the T6 condition, the tensile strength of 201AB increases from 393 to 434 MPa (57 to 63 ksi). Yield strength for this condition is 386 to 414 MPa (56 to 60 ksi), and elongation ranges from 0.5 to 8%.

Properties of cold-formed aluminum P/M alloys are increased by a combination of strain-hardened densification and improved interparticle bonding. Alloy 601AB achieves 257 MPa (37.3 ksi) tensile strength and 241 MPa (34.9 ksi) yield strength after forming to 28% upset. Properties for the T4 and T6 conditions do not change notably between 3 and 28% upset. Alloy 602AB has moderate properties with good elongation. Strain hardening (28% upset) results in 221 MPa (32 ksi) tensile and 203 MPa (29.4 ksi) yield strength. The T6 temper parts achieve 255 MPa (37 ksi) tensile strength and 227 MPa (33 ksi) yield strength. Highest cold-formed properties are achieved by 201AB. In the as-formed condition, yield strength increases from 209 MPa (30.3 ksi) for 92.5% density, to 281 MPa (40.7 ksi) for 96.8% density.

Table 13 Typical heat-treated properties of nitrogen-sintered aluminum P/M alloys

Heat-treated variables and properties	Grades			
	MD-22	MD-24	MD-69	MD-76
<b>Solution treatment</b>				
Temperature, °C (°F).....	520 (970)	500 (930)	520 (970)	475 (890)
Time, min .....	30	60	30	60
Atmosphere.....	Air	Air	Air	Air
Quench medium.....	H <sub>2</sub> O	H <sub>2</sub> O	H <sub>2</sub> O	H <sub>2</sub> O
<b>Aging</b>				
Temperature, °C (°F).....	150 (300)	150 (300)	150 (300)	125 (257)
Time, h .....	18	18	18	18
Atmosphere.....	Air	Air	Air	Air
<b>Heat-treated (T<sub>6</sub>) properties(a)</b>				
Transverse-rupture strength, MPa (ksi) .....	550 (80)	495 (72)	435 (63)	435 (63)
Yield strength, MPa (ksi) .....	200 (29)	195 (28)	195 (28)	275 (40)
Tensile strength, MPa (ksi) .....	260 (38)	240 (35)	205 (30)	310 (45)
Elongation, %.....	3	3	2	2
Rockwell hardness, HRE.....	74	72	71	80
Electrical conductivity, %IACS.....	36	32	39	25

(a) T<sub>6</sub>, solution heat treated, quenched, and artificially age hardened

**Table 14** Electrical and thermal conductivity of sintered aluminum alloys, wrought aluminum, brass, bronze, and iron

Material	Temper	Electrical conductivity(a) at 20 °C (68 °F), %IACS	Thermal conductivity(b) at 20 °C (68 °F), cgs units
601AB	T4	38	0.36
	T6	41	0.38
	T61	44	0.41
201AB	T4	32	0.30
	T6	35	0.32
	T61	38	0.36
602AB	T4	44	0.41
	T6	47	0.44
	T61	49	0.45
6061 wrought aluminum	T4	40	0.37
	T6	43	0.40
	Hard	27	0.28
Brass (35% Zn)	Annealed	27	0.28
	Hard	15	0.17
Bronze (5% Sn)	Annealed	15	0.17
	Hot rolled	16	0.18

(a) Determined with FM-103 Magnatester. (b) Converted from electrical conductivity values

Alloy 202AB is best suited for cold forming. Treating to the T2 condition, or as-cold formed, increases the yield strength significantly. In the T8 condition, 202AB develops 280 MPa (40.6 ksi) tensile strength and 250 MPa (36.2 ksi) yield strength, with 3% elongation at the 19% upset level.

### Properties of Sintered Parts

**Mechanical Properties.** Sintered aluminum P/M parts can be produced with strength that equals or exceeds that of iron or copper P/M parts. Tensile strengths range from 110 to 345 MPa (16 to 50 ksi), depending on composition, density, sintering practice, heat treatment, and repressing procedures. Table 12 lists typical properties of four nitrogen-sintered P/M alloys. Properties of heat-treated, pressed, and sintered grades are provided in Table 13.

Impact tests are used to provide a measure of toughness of powder metal materials, which are somewhat less ductile than similar wrought compositions. Annealed specimens develop the highest impact strength, whereas fully heat-treated parts have the lowest impact values. Alloy 201AB generally exhibits higher impact resistance than alloy 601AB at the same percent density, and impact strength of 201AB increases with increasing density. A desirable combination of strength and impact resistance is attained in the T4 temper for both alloys. In the T4 temper, 95% density 201AB develops strength and impact properties exceeding those for as-sintered 99Fe-1C alloy, a P/M material frequently employed in applications requiring tensile strengths under 345 MPa (50 ksi).

Fatigue is an important design consideration for P/M parts subject to dynamic stresses. Fatigue strengths of pressed and

**Fig. 14** Machining chips from a wrought aluminum alloy (right) and from a P/M aluminum alloy (left)

sintered P/M parts may be expected to be about half those of the wrought alloys of corresponding compositions (see comparisons of two P/M alloys with two wrought alloys in Fig. 13). These fatigue-strength levels are suitable for many applications.

**Electrical and Thermal Conductivity.** Aluminum has higher electrical and thermal conductivities than most other metals. Table 14 compares the conductivities of sintered aluminum alloys with wrought aluminum, brass, bronze, and iron.

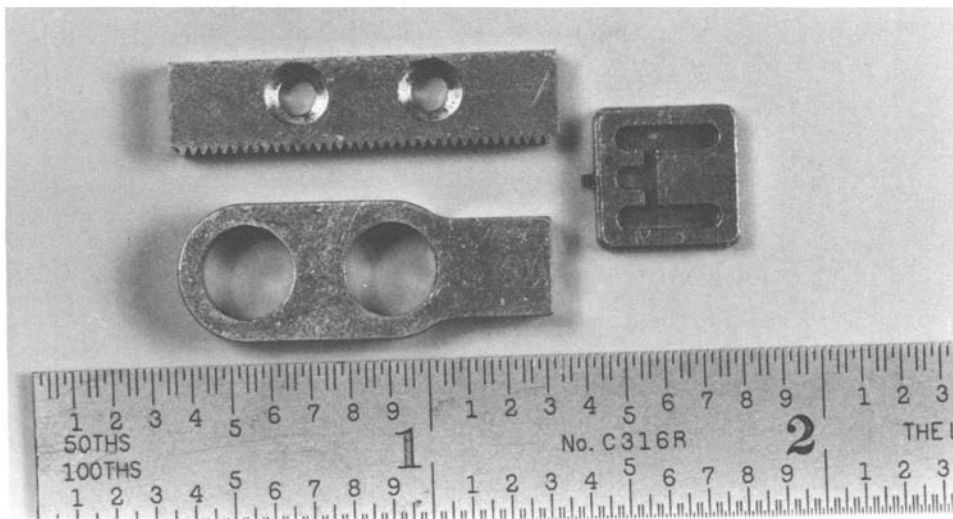
**Machinability.** Secondary finishing operations such as drilling, milling, turning, or grinding can be performed easily on aluminum P/M parts. Aluminum P/M alloys provide excellent chip characteristics; compared to wrought aluminum alloys, P/M chips are much smaller and are broken more easily with little or no stringer buildup, as can be seen in Fig. 14. This results in improved tool service life and higher machinability ratings.

### Applications for Sintered Parts

Aluminum P/M parts are used in an increasing number of applications. The business machine market currently uses the greatest variety of aluminum P/M parts. Other markets that indicate growth potential include automotive components, aerospace components, power tools, appliances, and structural parts. Due to their mechanical and physical properties, aluminum P/M alloys provide engineers with flexibility in material selection and design. These factors, coupled with the economic advantages of this technology, should continue to expand the market for aluminum P/M parts. A variety of pressed and sintered aluminum P/M parts are shown in Fig. 15.

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**Fig. 15** Typical pressed and sintered aluminum P/M parts made from alloy 601AB. Top: gear rack used on a disc drive. Bottom: link flexure used on a print tip for a typewriter. Right: header/cavity block used on a high-voltage vacuum capacitor. Courtesy of D. Burton, Perry Tool & Research Company

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