Intermetallic Compounds for High-Temperature Structural Use

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The next generation of efficient turbines and engines will require materials that can withstand operating temperatures approaching 2000°C. Intermetallic compounds with high melting temperatures are candidates for this application, but the obstacle of their limited ductility must first be overcome. Because the available data on these materials is limited, a survey of the effects of chemistry and crystal structure must be performed. With the use of melting temperature and density as figures of merit, the most likely candidates have been identified for preliminary screening.

INTERMETALLIC COMPOUNDS CONSIST OF METALS IN APPROXImately stoichiometric ratios in ordered crystal structures. We will show that such materials have special mechanical properties that may in the future make it possible for a Boeing 747 to fly from New York to Lima, Delhi to Tokyo, or Sydney to Manila using 64% of the fuel it now does; a commercial supersonic aircraft to fly from San Francisco to Seoul at 80% of current operating costs; or two military fighter planes to be purchased for the present cost of one. These are some of the possibilities if new, strong, high-temperature materials can be identified and processed.

Jet engine performance, like that of all heat engines, increases with the temperature of the working fluid, and this limiting temperature is set by the properties of available materials. The materials must be strong and stiff at elevated temperature and have some ductility at both elevated and ambient temperatures. Oxidation and corrosion resistance also are needed but will not be discussed in this article. The specific accomplishments envisioned above will not be attained solely with the intermetallic compounds that we discuss in this article, but these compounds could provide three quarters of the prospective savings in fuel and cost if they allow jet engine temperatures to be raised from their present maximum of ~1400°C to ~1850°C.

Needs for improved high-temperature structural materials are widespread, with much current attention to jet engines. Although there are abundant aerospace applications such as supersonic airplanes, spacecraft, missiles, rockets, and engines, there are also many earth-bound applications, for example, in turbines and in various types of reactors.

Why Intermetallic Compounds?

The materials currently used in structural components that are subjected to high stresses and elevated temperatures are multiphase alloys where the strengthening phase is an intermetallic compound; therefore, the development of new intermetallic phases that retain strength to even higher temperatures is required for improved performance. Identification of useful intermetallic compounds is a difficult process because easily measured parameters do not provide clear indications of rheological properties. A limited, well-studied sample of intermetallics is now known to have desirable strengths at intermediate temperatures; there are many other compounds that melt at higher temperatures, but their properties are largely unknown.

The most studied intermetallic compound is Ni_3Al, which melts at 1395°C. Its strength varies with temperature (Fig. 1) in a way that is strikingly atypical of most solids (*1*): its yield stress increases as temperature is raised from −200° to 700°C (0.58 of the melting temperature) (*2*). This behavior contrasts with the monotonic decrease of flow stress of most metals, as shown for a Ni-Cu alloy in Fig. 1. Several other compounds that have the same $L1_2(cP4)$ crystal structure as Ni_3Al (*3*) have the same behavior, but this property is by no means confined to this structure. It has been observed in other cubic (*4*) and tetragonal (*5, 6*) crystal types. Another desirable mechanical property of Ni_3Al is that it is ductile in both single crystals (*7*) and polycrystals of properly controlled solute content and stoichiometry (*8*). The unconventional and desirable structural properties of Ni_3Al produced great interest that led to extensive work on intermetallic compounds in earlier decades (*9–11*), and researchers have recently begun to focus on finding other compounds with similar properties (*12–14*).

Half of the melting temperature is generally considered the maximum use temperature for single-phase materials. Materials with high melting temperatures also generally have high strength, high elastic moduli, and good creep resistance. For these reasons intermetallics with melting temperatures higher than that of Ni_3Al are of clear interest, and studies should be done of some of the nearly 300 known binary intermetallic compounds that melt at temperatures above 1500°C (Fig. 2) (*15*). This group of materials is the largely unexplored pool from which workers hope to identify the high-strength, high-temperature materials of the future. In addition to high melting temperature, low specific gravities, ρ, are desired for aerospace applications and for rotating parts, where centrifugal stresses are proportional to ρ. For these uses, the compounds that have the highest melting temperature at any given ρ are the first choices in exploring new materials (*15, 16*) (the ones that plot close to the curve in Fig. 2).

Because of their strong bonding, high-temperature intermetallics generally have high elastic moduli, but values overlap those of pure metals. For the aerospace applications and for rotating parts, the clear superiority of intermetallics appears in the moduli relative to specific gravity. Because many intermetallics have low densities, many also have very high specific stiffnesses. We have measured

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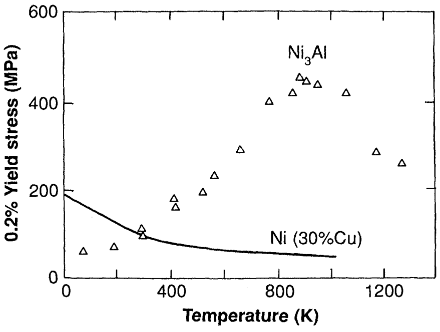

Fig. 1. Yield stress for the ordered intermetallic compound Ni_3Al [data from (*2*)] compared to that of a random solid solution of 30% Cu in Ni (*44*). The increasing stress flow from 77 K to >800 K is an uncommon and desirable feature for high-temperature use.

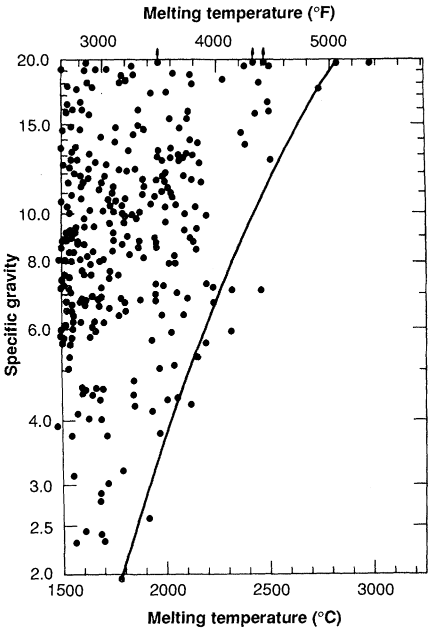

Fig. 2. Melting temperature versus specific gravity for 293 intermetallic compounds. The solid line is an empirical, approximate envelope to the data. The materials that are closest to this line are candidates for future high-temperature structural use.

specific shear stiffnesses for 21 binary intermetallics from those in Fig. 2; most (76%) have stiffnesses that are significantly above those of conventional metals such as Al, Fe, Ni, and Ti (with values ~10 GPa); 43% have stiffnesses exceeding 20 GPa; one, $Be_{12}Ti$, has a specific shear stiffness of 57 GPa. In addition, because most strengthening mechanisms produce hardening that is proportional to moduli, strength and specific strength generally should also be high in these compounds.

What Is an Intermetallic Compound?

The characteristics of intermetallic compounds are derived from their structures and the presence of ordering. We need, therefore, to know the location of the atomic sites (crystal structure) and the occupation of the sites (ordering).

Pure metals typically crystallize into one of the three different lattice structures shown in Fig. 3. The tendency of metals to adopt one of these crystal structures is generally attributed to the electronic structure of the atoms (*10, 17, 18*). The structure of metals varies from body-centered cubic (bcc) to hexagonal close-packed (hcp) to face-centered cubic (fcc) as the outer-shell bonding electron density per atom increases. The face-centered cubic lattice has atoms at the eight corners and six faces of the cube (Fig. 3B). The atoms in the fcc lattice have 12 nearest-neighbor atoms, which give it the densest atomic arrangement for spheres that are all of the same size. An alternate stacking of close-packed planes produces the hcp structure (Fig. 3C). The bcc lattice (Fig. 3A) has atoms occupying the eight cube corners and the cube center. In contrast to the close-packed lattices, the atoms in the bcc structure have only eight nearest-neighbor atoms and therefore form a less dense structure. In a number of cases the looser packing of the bcc structure makes it the preferred structure at high temperature as a result of entropic stabilization. Conversely, metals with the bcc structure commonly transform to a close-packed lattice at elevated pressure.

As a second element is added to the pure metal, the solute atoms substitute for the primary metal atoms in the lattice randomly. For some combinations of similar metals with a common lattice structure, the random substitution can occur across the entire range of composition. For other metal combinations, as the concentration of the solute increases, a new structure develops in which the lattice sites are occupied by specific types of atoms (ordering). Ordering is illustrated for the Ni-Al system (Fig. 4). We use this system as an example because it is one of the most studied, and the crystal structures of the ordered phases are relatively simple. The concepts are then easily generalized to other compositions and crystal structures. The maximum solubility of Al in disordered Ni fcc solid solution is about 20% (Fig. 4, right). At higher Al concentrations, order develops with the Ni atoms residing on the cube faces and the Al atoms located at the cube corners. This crystal structure is designated as $L1_2$ in the Strukturbericht notation or as *cP*4 with Pearson's symbol (*19*). This particular Pearson symbol indicates a *c*ubic *p*rimitive structure with *four* atoms per unit cell. In the perfectly ordered $L1_2$ structure, the atomic ratio is A_3B, and each *B* atom is coordinated by twelve *A* atoms. This notation scheme is applicable to any crystal structure, including the much more complicated *C*15 (*cF*24). This structure consists of several interpenetrating face-centered cubic lattices with 24 atoms per cell. Depending on the constituent elements, some compounds exist over a range of compositions with significant deviations from the ideal stoichiometric ratio. For example, the $L1_2$ nickel aluminide phase field extends for several atomic percent about the ideal ratio (Fig. 4); remarkably, in some binary systems the stoichiometric composition is not even included in the range of composition over which the compound is stable. These deviations can be accommodated by anti-site defects, for example, *A* atoms on *B* sites, by vacancies, or by interstitial atoms.

The driving force for ordering is the greater strength of the *A*-*B* bonds than of the *A*-*A* and *B*-*B* bonds that they replace on ordering. Thus, unlike atoms are favored over like ones as nearest neighbors. If the ordering energy is low, as for alloying of Ni with Fe, then as the temperature increases, the entropic tendency for disorder dominates and causes the lattice to disorder. For other metallic combinations, such as Ni with Al, the ordering energy is high enough that the $L1_2$

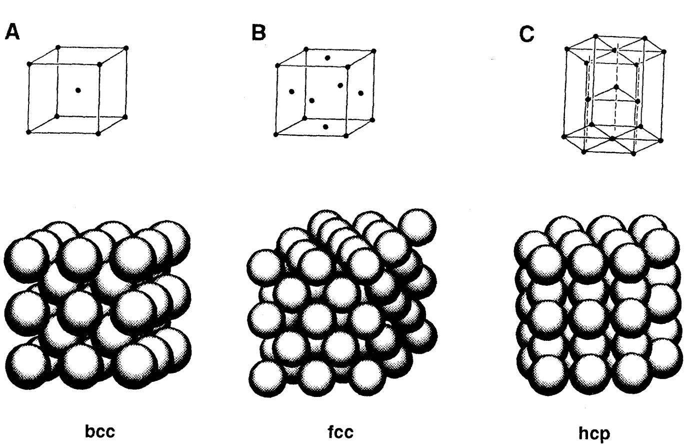

Fig. 3. The three most common crystal lattice structures for pure metals: (**A**) body-centered cubic, (**B**) face-centered cubic, (**C**) hexagonal close-packed.

structure is retained up to the melting temperature.

The phase with compositions near NiAl has the *B*2 (*cP*2) structure. This phase can be thought of as an ordering of the bcc structure with an Al atom in the cube center and Ni atoms at the cube corners. As with the *L*1_2 phase, the *B*2 structure is stable for significant (10%) deviations from the ideal ratio. At higher Al concentrations, other intermetallic phases are observed, including Ni_2Al_3 (an ordered hexagonal phase) and Al_3Ni (an ordered orthorhombic phase).

The Ni-Al binary system illustrates some of the complexities of intermetallic phase formation. Both Ni and Al are fcc metals. When alloyed to form Ni_3Al, an ordered fcc phase results. However, when alloyed to Al_3Ni, an orthorhombic phase occurs. This type of asymmetry in behavior is common, and therefore calculations of binary phase diagrams are difficult, and only a few have been successfully determined. Ternary phase relations are even more difficult to calculate, although progress is being made in this area. Some help comes from considering various empirical relations among intermetallic phases and the chemical properties of the constituent metals (*20*). Over 60 years ago, Hume-Rothery (*17*) noted that the average number of valence electrons per atom has specific ranges for the different observed lattice structures for many binary alloys. Since then metallurgists have recognized that other factors are also important in determining the structure of materials. The difference in the electronegativity of the constituent atoms is a dominating factor in phase stability because it determines the directionality and strength of the bonds. Geometrical factors, primarily atomic size, determine the degree of packing that is consistent with increased symmetry.

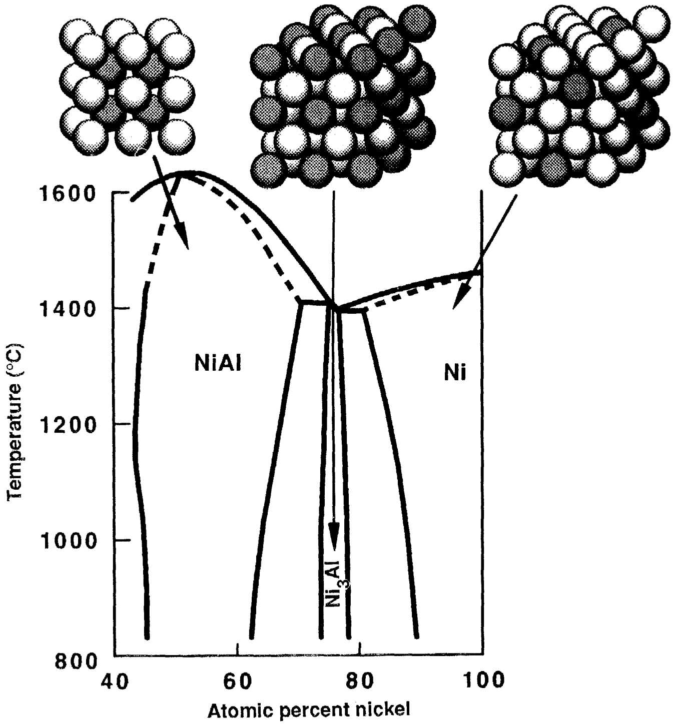

Fig. 4. Partial binary phase diagram for Ni-Al system. The crystal structures for the phases are indicated. The slightly larger atomic diameter of the Al atoms (solid) compared to the Ni atoms (open) is ignored. The unmarked areas that lie between the marked phases contain mixtures of the two horizontally adjacent phases. Ni_3Al melts at 1395°C; the upper phase field therefore is liquid.

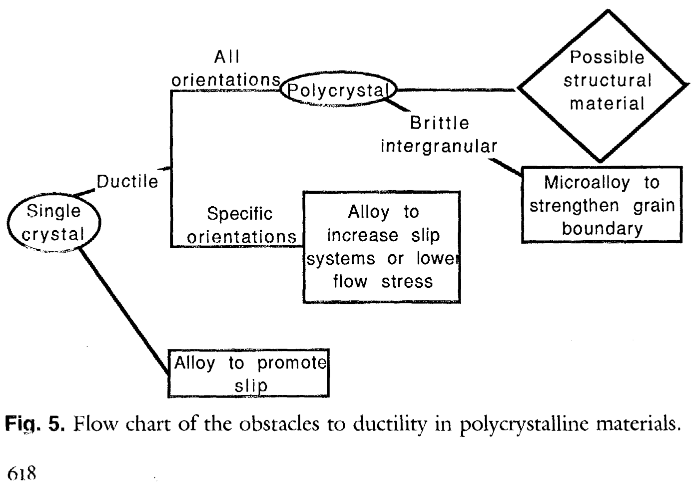

Fig. 5. Flow chart of the obstacles to ductility in polycrystalline materials.

A significant problem in relating the observed crystal structures of intermetallic phases to the properties of the constituent elements is that the proper atomic parameters are not yet recognized. For example, in addition to the well-known Pauling scale, there are several other electronegativity scales that can be used, including Mulliken, S-orbital, Martynov-Batsanov, and Watson-Bennet. There are also several definitions for atomic size. One of the more successful correlations of binary phase crystal structure with atomic parameters is that of Villars (*21*). As the atomic parameters, he used the Martynov-Batsanov electronegativity difference, the sum of the number of valence electrons, and the difference of the Zunger pseudopotential radii sums. In the resulting three-dimensional structural stability diagrams, the 20 *AB* structure types (90% of the more than 1000 known *AB* phases) are separated into contiguous groups with only 22 compounds plotting in the wrong group. The 26 AB_2 structure types are grouped with only 27 violations out of a total of 1011 compounds. Similar results were obtained for the A_3B and A_3B_5 compounds.

Pettifor (*22, 23*) recently proposed a more general approach based on a particular sequence of the elements from the periodic table. He numbered each element primarily by group number rather than by period (atomic number) to reflect the "quantum character" of the atoms. Using this "Mendeleev number" for the axes, he constructed two-dimensional maps of the various A_xB_y compounds. He showed that compounds with the same structure type occupy contiguous regions of the map with few violations. With this approach, modifying the crystal structure with alloying additions can be guided by the use of an average, compositionally weighted Mendeleev number.

An alternative to these empirical relations is the development of tools for predicting phase stability based on first-principle calculations of the electronic structure of the material. Given a particular stoichiometry, the general approach to determining the stable structure is to compare the total energy of different crystal structures. Because the energy differences between competing structures are generally less than 0.1 eV per atom out of a cohesive energy of several electron volts per atom, this approach has only recently become possible with the increased availability of high-speed computers and the combined theoretical and computational approaches that have been developed. For example, Freeman has used an all-electron band structure to calculate phase stability in the Ni-Al system, including the effect of alloying additions on Ni_3Al (*24*). The long-term goal is to combine these calculations of the enthalpic contribution to the energy with calculated entropic effects to predict high-temperature equilibria and therefore entire phase diagrams.

Effect of Ordering on Mechanical Properties

Deformation in metals and alloys occurs by the motion of defects in the crystal lattice. The nature of the defect involved depends on the magnitude of the applied stress and the temperature. At low temperature and high stress, the deformation proceeds via the stress-induced motion of linear defects known as dislocations. As the temperature increases, diffusion of point defects, such as vacancies, dominates. Both of these mechanisms involve high-energy intermediate states in which the local atomic structure is greatly perturbed. Because the atoms in ordered phases prefer specific neighbors and interatomic distances, the activation barrier for the flow process and, hence, the resistance to deformation are high. The barriers can lead to high strength and retention of strength to elevated temperature. However, in many cases this resistance is so high that brittle fracture occurs. Brittleness at ambient temperature is a critical problem for many intermetallic compounds that would otherwise be mechanically useful. Although there are exceptions, most intermetallics are easily fractured at low temperatures. For example, the thermal stresses produced during cooling often cause cast samples to fracture spontaneously; most samples are easily fractured at room temperature by a chisel and a light hammer blow.

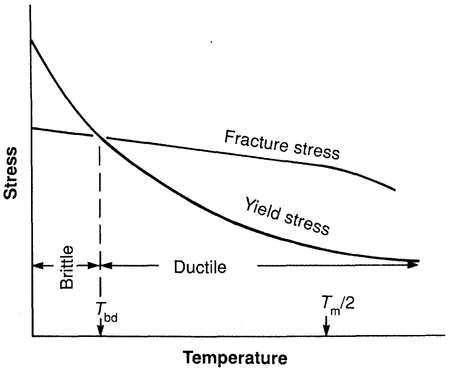

Fig. 6. Schematic variation of fracture stress and flow stress with temperature. The fracture stress decreases slowly with the elastic modulus; the flow stress is in addition decreased by thermally assisted plastic flow. Below the crossover temperature T_{bd} the material is brittle.

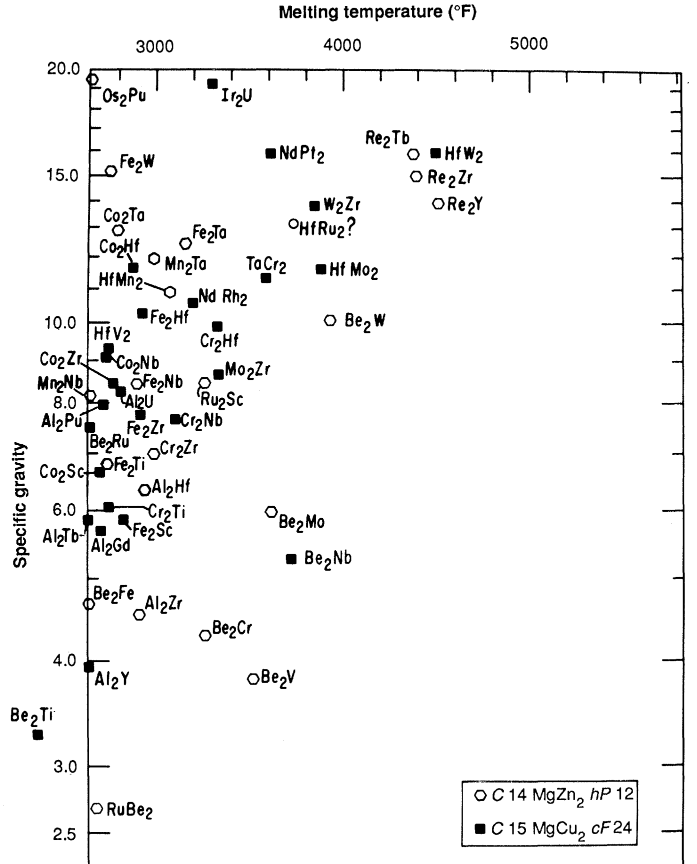

Fig. 7. Melting temperature versus specific gravity for *C*14 and *C*15 Laves phases.

The three major reasons for fracture in order of decreasing severity and frequency are high flow stress relative to cleavage stress, inadequate active slip systems, and grain boundary weakness. The flow chart shown in Fig. 5 illustrates how these problems can surface and affect the sequence of choices in the development of new structural alloys. Consider first a single crystal of the compound. Some materials will exhibit brittle behavior in all orientations, indicating that the flow stress is higher than the fracture stress on all the available slip planes. This situation is the most severe case of brittle behavior and in principle can be overcome only by alloying to lower the ratio of the fracture stress to the flow stress; however, we know of no case where this approach has been successful. In some cases, the single crystal is ductile in only certain orientations. This situation is an indication that too few slip systems are available in the crystal and that the ductility of the polycrystalline form of the material may be restricted. One approach is to alloy in order to increase the number of active slip systems. Alternatively, ductility can be obtained in polycrystalline specimens by controlling grain size and morphology. Finally, some compounds are ductile as single crystals in all orientations but are brittle when prepared in polycrystalline form. This condition is often associated with fracture at the grain boundaries.

Cleavage versus plasticity. The most general explanation of the tendency toward brittleness is largely phenomenological. If brittle failure can occur at a lower stress than plastic deformation, brittleness will be the rule (Fig. 6). With increasing temperature, a brittle-to-ductile transition will occur at a crossover temperature T_{bd} where plasticity is easier than fracture. This schematic description is consistent with the observation that such transitions occur in many materials. In some intermetallics, slip requires such high stresses that cracks are produced even at the highest temperatures that have been tested, for example, to 1300°C for $Re_{24}Ti_5$—a structure that has 58 atoms per unit cell.

Rice and Thomson (*25*) have described the microscopic processes and important parameters that are thought to affect the qualitative relations shown in Fig. 6. They considered that an incipient crack might either be formed at the cost of a surface energy per unit area γ, or it might be suppressed by forming dislocations that serve to blunt the crack at an energy cost that increases with the shear modulus G and the slip vector b of the dislocation that is formed. They predicted brittle failure for $Gb/\gamma > 10$ and ductile behavior for $Gb/\gamma < 7.5$. Therefore, selecting structures that minimize b and maximize γ should aid in identifying ductile compounds. Low values of the elastic moduli would also encourage ductility but at the cost of reducing the high stiffness and plastic strength which are the useful features of the intermetallics.

The parameter Gb/γ provides useful qualitative guidance as to when cracks will appear. There are, however, many effects of crack and slip orientations, and for many compounds needed data are not available. A further complication is that b may decrease as a result of dissociation of full dislocations into partial ones that are bound together by fault surfaces (*26*). Such dissociations have been observed in a few cases, but even for these cases, effects of alloying on fault energies and the extent of separation of partials are unmeasured complications.

The best hope of controlling ductility is through the slip vector b. In many of the intermetallic compounds that have large, complicated unit cells, the minimum lattice translation vector is large. Therefore b is large, and brittleness is more likely for more complicated structures.

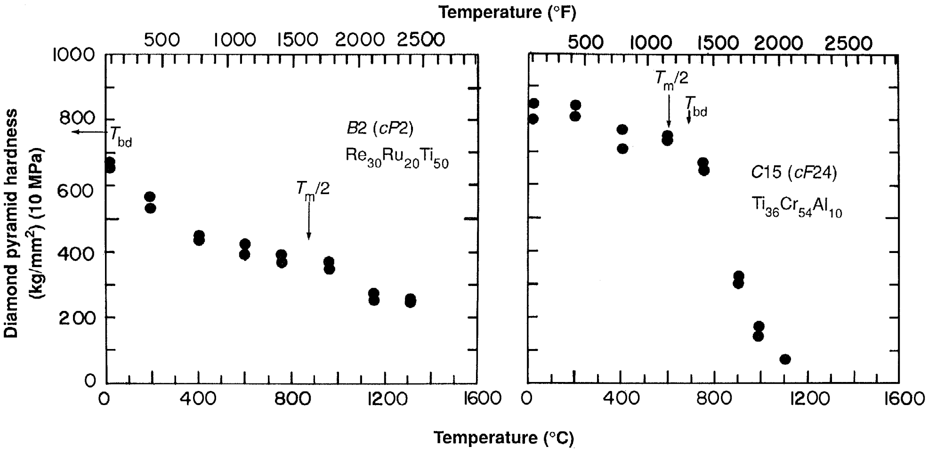

Fig. 8. Diamond pyramid microhardness $h(T)$ versus temperature for a load of 500 g; T_{bd} is the temperature above which cracks are not observed at indentations. T_m is the melting temperature. (**Left**) For one of the simpler structures (only two atoms per unit cell), $Re_{30}Ru_{20}Ti_{50}$, $h(T)$ decreases monotonically, corresponding to thermally assisted plastic deformation. T_{bd} is <23°C. (**Right**) For a more complicated structure (24 atoms per unit cell), $Ti_{36}Cr_{54}Al_{10}$, $h(T)$ decreases negligibly up to half of the melting temperature. Little macroscopic plasticity is expected. T_{bd} is 700°C.

Available slip systems. Von Mises (*27*) noted that uniform, arbitrary deformation by slip requires the action of five independent slip systems, a slip system being a combination of a surface—the crystallographic slip plane—and a direction of slip on that plane. The proof of von Mises' statement is straightforward and rigorous: There are three volume strains (only two of which are independent if volume is conserved) and there are three shear strains, giving five independent quantities that can be specified. To accommodate the imposed five strains, five nonredundant deformation modes are required, that is, five slip systems for materials in which deformation is entirely by slip.

Studies of most of the ordered structures that are typical of the high-temperature intermetallic compounds have shown that at room temperature, fewer than five independent systems normally can act; therefore, extensive plastic deformation at ambient temperatures is often difficult. One undesirable resolution of the difficulty is the opening and propagation of cracks, that is, brittle failure. However, brittleness may occur from other causes. Alternatively, brittleness may not be encountered because the von Mises criterion does not apply if sufficient other modes of deformation (for example, twinning, kinking) are available or if deformation is nonuniform. Locally concentrated stresses may also activate new slip systems that ordinarily do not act at ambient temperatures but are observed at elevated temperatures where thermally activated slip is easier.

The need for five independent slip systems is an inference derived under the assumption that uniform strain occurs throughout a polycrystalline sample. Nonuniform strain relaxes this constraint. At grain boundary between just two crystals, on the basis of von Mises' criterion, activation of only two slip systems is required in order to maintain continuity in each crystal (*28*). And in polycrystalline alpha brass (face-centered cubic 70% Cu, 30% Zn), which has 12 slip systems from which many sets of five independent ones might be activated, slip in the interiors of nearly half of the crystals does not occur on more than four (*29*). At planar grain boundaries, deformation is often accommodated by fewer than five slip systems (*30*), but at grain corners and along curved or jagged boundaries, slip is more constrained (Table 1). As a result, these regions are prime sites for nucleating cracks. Whether those cracks ultimately propagate catastrophically depends on diverse factors.

To summarize, limited numbers of available deformation modes may lead to crack formation and brittleness. The precise requirements for suppressing crack formation are not yet known.

Grain Boundary Fracture

Having several active, independent slip systems is a necessary but not sufficient criterion for macroscopic ductility in polycrystalline materials. For example, even though at room temperature single crystals of Ni_3Al are ductile in all crystal orientations (*13*), the polycrystalline alloy fractures with negligible plastic strain to failure (*31*). The poor ductility results from premature fracture along the grain boundaries that are not strong enough to withstand the stresses required for deformation of the grain interiors.

In pure metals, grain boundary fracture is generally associated with impurities that segregate to the grain boundaries and weaken them (*32*). The most detrimental grain boundary segregants, such as S, P, Sn, and Sb, have high electronegativities and therefore act to decrease the charge that is in the metal-metal bonds at the boundary, thereby reducing the cohesive strength and promoting intergranular fracture (*33*). Intermetallic compounds are also sensitive to this "pest degradation." Sulfur embrittles at room temperature and oxygen embrittles at higher temperatures (*34, 35*). However, certain intermetallic compunds exhibit brittle intergranular fracture even when prepared from high purity metals (*31, 36*). Although the basis of the intrinsic weakness of grain boundaries in certain intermetallic compounds is still in dispute, it is probably related to the severe disruption in the lattice structure that occurs at the boundary.

Aoki and Izumi prevented intergranular fracture in an intermetallic compound by alloying (*8*). They showed that microalloying additions (less than 1 atomic percent) of boron to Ni_3Al resulted in more than 30% plastic strain to failure with fully transgranular fracture compared to the zero ductility and pure intergranular fracture observed for the unmodified compound processed under identical conditions. Although the mechanism for the boron-induced ductility is still in dispute, the phenomenology of the effect has been well established (*37–39*). Boron is effective in small quantities because it concentrates strongly in a thin layer (<5 nm) at grain boundaries where it can have an enhancement of about ten times the bulk concentration. Subsequent studies have shown that ductility is obtained in Ni_3Ga by boron doping and in Ni_3Si by either boron or carbon doping (*40*).

Crystal Structure and Mechanical Properties

Many of the high melting temperature compounds shown in Fig. 2 are now being screened for desirable properties. As a first step, two generally available properties, melting temperature T_m and specific gravity ρ, have been used for preliminary screening. These properties are useful because they are insensitive to processing history and to the presence of minor concentrations of alloying elements, factors that would complicate selecting candidate materials, and T_m and ρ are first-order figures of merit. Compounds with a high T_m generally have high stiffness, high strength, low thermal-expansion coefficient, and low creep rate; low-ρ compounds are

Table 1. Fractions of different regions in which fewer than five independent slip systems act in alpha brass.

Sites	Fraction with <5 independent slip systems
Grain interiors	49 ± 4%
Planar grain boundaries	62 ± 5%
Curved or jagged grain boundaries	30 ± 4%
Grain corners	23 ± 4%

light weight and have high specific strength and specific stiffness (*15, 16*).

The many different types of intermetallic compounds (Fig. 2) are best identifed and separated according to structures (*15*). For example (Fig. 7), among the hexagonal *C*14 (*hP*12) structures, several beryllides and two or three rhenium compounds lie close to the low ρ–high T_m boundary, whereas for the *C*15 (*cF*24) structure, Al_2Y, Al_2Zr, Re_2Nb, or HfW_2 might be selected for first testing if no other factors than T_m and ρ were to be considered. There are, however, many other facets to such decisions. They include prospects for solid solution alloying (which adds new dimensions in adjusting properties), the magnitudes of likely slip vectors (which affect brittleness and the ease of moving dislocations), stacking fault and antiphase domain boundary energies (which control the possibilities for dislocation splitting and ease of slip transferral from one atomic plane to the next), and the possibility of equilibrium with a metallic phase of one of the major constituents (which sometimes can diminish the brittleness of the intermetallic phase).

Definitive testing of even a single composition is a time-consuming and therefore expensive task. For more rapid and economical, but less definitive, screening we have been using microhardness tests as a function of temperature to give an indication of strength versus temperature; the temperature above which cracks no longer occur at the hardness indentations is taken as a brittle-to-ductile transition temperature for the purpose of intercomparison. The dozens of ordered intermetallics that we have studied in this manner show a wide variety of behaviors.

Certain behaviors are so common in many compounds as to suggest that there is a dichotomy in the data (Fig. 8). The *B*2 (*cP*2) structure $Re_{30}Ru_{20}Ti_{50}$ shows almost uniformly decreasing hardness with temperature, and there is a nearly 50% decrease between room temperature and $T_m/2$. This behavior is commonly observed in relatively simple structures with either known or easily envisioned slip systems; that is, plastic deformation is possible. This particular structure is simple cubic with only two atoms per atomic cell. (Ru and Re atoms are mixed in one position, Ti occupies the other.) Thermally assisted activation of deformation is presumed to be responsible for much of the observed decrease in flow stress. Plasticity is often extensive enough at the hardness identations (as it is here) that no cracks are created. Unfortunately this behavior does not insure that the material is ductile on the larger scale that applies in more conventional tension or bending tests.

The behavior of $Ti_{36}Cr_{54}Al_{10}$, which has 24 atoms per unit cell, is typical of more complicated structures. This *C*15 (*cF*24) crystal type is called the cubic Laves phase. The structure is complicated and likely slip systems are not obvious from theory. Suggestions that have been made for slip processes involve special zonal dislocations, which require that different displacements simultaneously occur on differing adjacent atomic planes (*41, 42*). Hardness values drop by less than 20% between room temperature and $T_m/2$. Because 20% is a typical drop in Young's modulus (E) over that temperature range (*43*), the hardness relative to E is essentially unchanged and only drops above $T_m/2$, where (typically) diffusional creep supplies a new mechanism of deformation that makes materials unsuitable for structural use. In short, no macroscopic plasticity is to be expected in the temperature range of interest.

In the case of $Re_{30}Ru_{20}Ti_{50}$ the strength is still substantial at 1300°C, and if single-phase materials are to be found with adequate mechanical properties at this high a temperature, they will most likely come from this class of materials, rather than from those like $Ti_{36}Cr_{54}Al_{10}$.

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