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J. WADSWORTH
G.A. HENSHALL
T.G. NIEH
E.M. TALEFF

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MATERIALS ISSUES IN SOME ADVANCED FORMING TECHNIQUES, INCLUDING SUPERPLASTICITY

J. Wadsworth, G. A. Henshall, and T. G. Nieh
Chemistry & Materials Science Directorate
Lawrence Livermore National Laboratory
Livermore, California

E. M. Taleff
The University of Texas at Austin
Department of Aerospace Engineering and Engineering Mechanics
Austin, Texas

ABSTRACT

From mechanics and macroscopic viewpoints, the sensitivity of the flow stress of a material to the strain rate, i.e. the strain rate sensitivity (m), governs the development of neck formation and therefore has a strong influence on the tensile ductility and hence formability of materials. Values of strain rate sensitivity range from unity, for the case of Newtonian viscous materials, to less than 0.1 for some dispersion strengthened alloys. Intermediate values of $m = 0.5$ are associated with classical superplastic materials which contain very fine grain sizes following specialized processing. An overview is given of the influence of strain rate sensitivity on tensile ductility and of the various materials groups that can exhibit high values of strain rate sensitivity. Recent examples of enhanced formability (or extended tensile ductility) in specific regimes between $m = 1$ and $m = 0.3$ are described, and potential areas for commercial exploitation are noted. These examples include: internal stress superplasticity, superplastic ceramics, superplastic intermetallics, superplastic laminated composites, superplastic behavior over six orders of magnitude of strain rate in a range of aluminum-based alloys and composites, and enhanced ductility in Al-Mg alloys that require no special processing for microstructural development.

INTRODUCTION AND OVERVIEW

At medium-to-high homologous temperatures, in the absence of significant cavitation or grain boundary separation, the tensile ductility of materials is largely controlled by the sensitivity of the flow stress to the strain rate, i.e. the strain rate sensitivity. This is because the strain rate sensitivity governs the evolution or development of necks in tensile tested material or in real manufacturing operations having areas undergoing tensile strain. In some cases, strain hardening can also have an influence. This is because the strain hardening characteristics govern the onset or start of necks. An overview of the influence of strain rate sensitivity on tensile elongation was shown in a classical compilation of data by Woodford[1]; these data are reproduced in Fig. 1, in which elongation is plotted as a function of strain rate sensitivity for a wide range of materials. The elongation is seen to increase to large values as the strain rate sensitivity increases. The simplest measures of strain rate sensitivity and strain hardening are through the exponents, m and N, respectively, in the equation $\sigma = K'\dot{\varepsilon}^m\varepsilon^N$, where σ is the true flow stress, K is a constant, $\dot{\varepsilon}$ is the true strain rate, and ε is the true strain. In this paper, we will for the most part focus on the influence of strain rate sensitivity, i.e. for the case $\sigma = K\dot{\varepsilon}^m$ because in many cases of high temperature plasticity, the role of strain hardening in influencing ductility is relatively minor. [It should be noted that particularly in studies of creep, that this equation is written in the reverse form, i.e. ($\dot{\varepsilon} = K''\sigma^n$; in which n is the stress exponent, and has a value that is the reciprocal of m.]

As shown in Table I, the range of values of strain rate sensitivity found in practical materials is quite broad, i.e. from 1.0 to <0.1. At values of $m = 1$, the flow stress becomes independent of strain rate and the material behaves in a Newtonian viscous manner. Thus, necking is extremely slow to propagate and in principle, therefore, almost unlimited tensile ductility would be expected. Examples of Newtonian-viscous materials include hot glass, tar, and well-masticated chewing gum. Values of $m = 1$ are found in some crystalline materials under certain testing conditions. It is important to note, however, that a high value of m is a necessary but insufficient condition for high elongations. For example, prior to about 1986, many fine grained ceramics were found to exhibit values of $m = 1$ in compression tests; however, they rarely exhibited useful tensile ductilities because of premature grain boundary separation. (We will describe recent examples of large tensile elongations in superplastic ceramics in a subsequent section.) Also, certain anisotropic materials can exhibit values of $m = 1$ under specialized testing conditions of thermal cycling which generate large internal stresses. In these groups of materials, very large tensile elongations (~1000%) can be obtained. This phenomenon is known variously as internal stress, thermal cycling, or transformational superplasticity. It is worth emphasizing that a fine-grained microstructure is not required for this type of superplasticity.

Classical polycrystalline, fine-grained, metallic superplastic materials typically have values of strain rate sensitivity that are about 0.5. Extremely large values of elongation-to-failure (up to 8000%) have been found in such materials, although typical values are about 400-1000% [2].

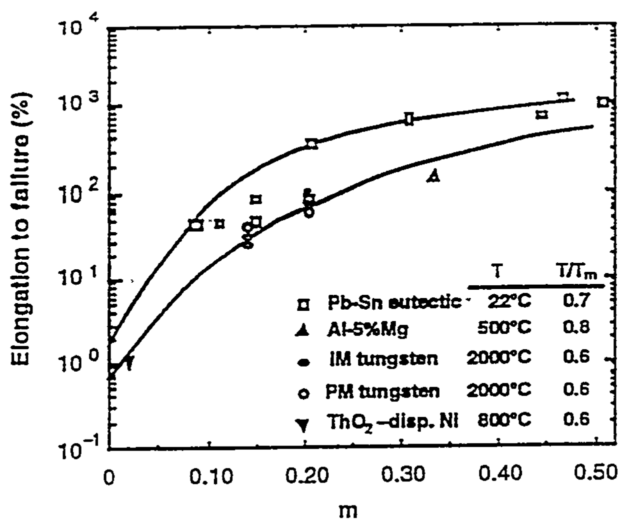

Fig 1. Elongation to failure as a function of strain rate sensitivity from Woodford.[1]

In certain metallic alloys where dislocation motion is controlled by the dislocation glide step in the glide/climb mechanism of high temperature deformation, the strain rate sensitivity is 0.33. For materials in this group, relatively large elongations (up to 400%) can be found that are intermediate between superplastic ($m \geq 0.5$) and non-superplastic ($m \leq 0.2$) materials.[3,4] In contrast, most materials that deform by a dislocation-climb-controlled mechanism exhibit a value of strain rate sensitivity of ≤ 0.2 and total elongations are usually not in excess of 50-80%. In dispersion strengthened materials, the creep stress exponent is extremely high (n~25) and therefore the strain rate sensitivity value is very low (m~ 0.04). In such cases, corresponding tensile ductilities are also extremely low (<5%). A schematic overview of the above groups of materials is shown in a log σ–log $\dot{\varepsilon}$ plot in Fig. 2.

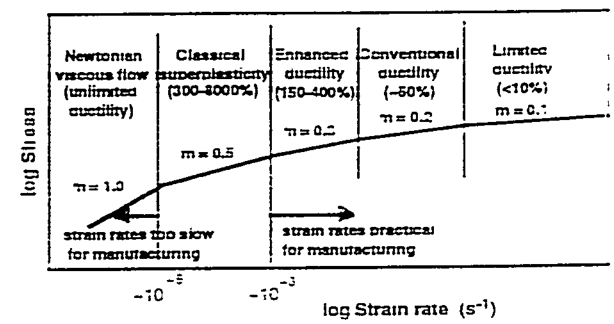

Fig 2. A schematic illustration of log σ–log $\dot{\varepsilon}$ plot illustrating the various strain rate sensitivity regimes, and the associated values of tensile ductility.

TABLE I
STRAIN RATE SENSITIVITY VALUES FOUND IN DIFFERENT CLASSES OF POLYCRYSTALLINE SOLIDS

Strain Rate Sensitivity Value, m	Materials
1	— Fine grained ceramics — Internal stress superplastic materials (α–U, Zn, Al-based and Zn-based composites, Fe, Ti) — Fine grained superplastic materials in which grain boundary sliding is accommodated by dislocation glide in a Class I solid solution — Materials undergoing viscous creep (Nabarro-Herring, Coble, Harper-Dorn)
0.5	— Classical fine grained superplastic materials including metallic alloys, intermetallics and ceramics
0.33	— Class I solid solutions — Some composites at high strain rates — Some coarse-grained intermetallic alloys
0.2	— Class II solid solutions — Many pure metals
≤ 0.1	— Alloys containing dispersions of hard phases

DISCUSSION

In the following sections, recent advances are described in groups of materials exhibiting values of $m \cong 1$, $m \cong 0.5$, and $m \cong 0.33$.

1. Materials Exhibiting Values of $m \cong 1$

As mentioned in the previous section, there is one type of superplasticity known as internal stress superplasticity. In materials in which large internal stresses can be developed, considerable tensile plasticity can take place under the application of a low, externally applied stress, and the materials can have a strain rate sensitivity exponent of as high as unity, i.e. they can exhibit ideal Newtonian-viscous behavior. Such superplastic materials deform, in fact, by a slip creep mechanism.

There are many ways in which internal stresses can be generated. These include thermal cycling of composite materials in which the constituents have different thermal expansivity coefficients, thermal cycling of polycrystalline pure metals or single phase alloys that have anisotropic thermal expansion coefficients, and thermal cycling through a phase change.

It has been shown that internal stress superplasticity can be utilized to enhance the ductility of metal-matrix composites that are normally brittle. Thus, several whisker-reinforced metal-matrix composites were made ideally superplastic, i.e. Newtonian-viscous in nature, during deformation under thermal cycling conditions. An example of the exceptional tensile ductility that can be achieved in this manner is shown in Fig. 3 for a whisker-reinforced 6061 aluminum alloy. Whereas the metal matrix composite exhibits only 12% elongation under isothermal creep deformation at 450°C, the same composite exhibits 1400% elongation when deformed under thermal cycling conditions (100↔450°C at 100 seconds per cycle). An example of similar behavior was demonstrated with a zinc-30 vol.% alumina particulate composite. Whereas this material exhibited essentially nil ductility when tested in tension at 300°C, it exhibited elongations exceeding 150% when deformed under thermal cycling conditions.

The basis of understanding the effect of internal stress on enhancing the ductility of metal matrix composites is that during thermal cycling internal stresses are developed at the interfaces between the metal matrix and the hard ceramic second phase.(2) This is because the thermal expansion coefficient of the metal matrix is several times larger than that of the ceramic phase. These internal stresses will relax by plastic deformation in the metal matrix to the value of the local interfacial yield stress of the matrix. It is this remaining local yield stress, which we define as the internal stress, σ_i, which contributes to the low applied external stress, and results in macroscopic deformation along the direction of the applied force. The creep behavior of two 2024 Al-SiC_w composites under both thermal cycling and isothermal conditions are shown in Fig. 4. The graph shows a plot of the diffusion-compensated creep rate as a function of the modulus-compensated stress. Three trends can be noted. First, the thermally cycled composites are much weaker than the isothermally tested composites at low applied stresses. Second, the thermally cycled composites have strain-rate-sensitivity exponents of unity at low stresses. Third, the thermally cycled samples and the isothermally tested samples yield data that converge at high stresses; this is expected since the internal stress generated by thermal cycling (a constant) will have a diminishing contribution to creep as the applied stress is increased.

Fig 3. The top sample is an untested 6061-Al-SiC $_w$ reinforced composite. The center sample exhibits 12% elongation under isothermal testing at 450°C, and the bottom sample exhibits 1400% elongation under thermal cycling conditions (100↔450°C) at $\sigma = 10$ MPa.

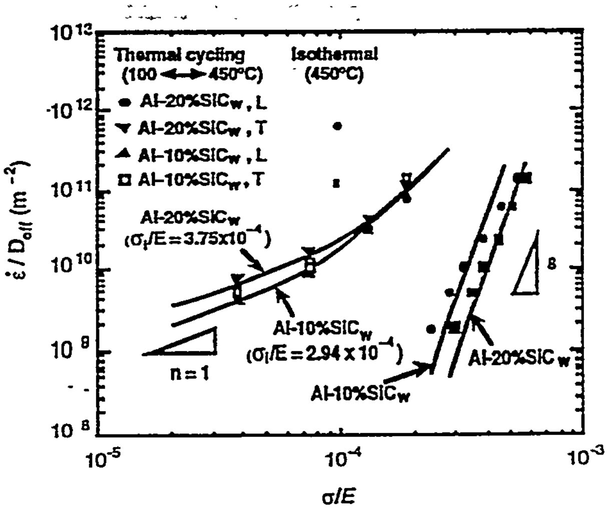

Fig. 4 Diffusion-compensated strain rate as a function of modulus-compensated stress for thermal cycled and isothermal creep of 2024 Al alloys containing 10%SiC_w and 20%SiC_w.

Other examples of internal stress superplasticity are in polycrystalline zinc and alpha uranium. In the case of these two metals, the internal stress arises from the anisotropy of expansion coefficients present in these materials with non-cubic crystal structures. During thermal cycling, internal stresses will be induced at the boundaries between adjacent grains.

Materials that undergo polymorphic changes can exhibit Newtonian-viscous behavior when tested during thermal cycling across the phase boundary. Among the early investigations in this subject were those on iron-base alloys. The internal stress arising in this case is from the difference in volume between the two phases during phase transformation. Such behavior is sometimes known as transformation plasticity. Interest in phase transformation plasticity appears to have been revived recently on commercially pure titanium.

It should also be noted that pressure induced phase transformations have been cited as a source of superplastic flow in geological materials. For example, there is a transformation in the earth's upper mantle, because of pressure, from orthorhombic olivine to a spinel phase at a depth of about 400 km below the earth's surface. It is believed that internal stress superplasticity arising from transformation stresses through pressure cycling (analogous to temperature cycling) leads to a mixed phase region of low effective viscosity.

2. Materials Exhibiting Values of $m \equiv 0.5$

Most classical superplastic metallic alloy materials have values of $m = 0.5$. In recent years, a major breakthrough was discovered in that similar behavior was found in some fine-grained, polycrystalline ceramics.

An example of a superplastic ceramic is yttria-stabilized tetragonal zirconia polycrystal (Y-TZP) consisting of 90% tetragonal phase zirconia and 10% cubic-phase zirconia. Wakai *et al.* showed that this material, with a grain size of about 0.3μm, was superplastic at 1450°C with up to 200% elongation and a strain-rate-sensitivity exponent of 0.5. More recently, studies on this material have shown elongations of up to 800% in Y-TZP and elongations of up to 500% on ceramic composites based on fine-grained zirconia, e.g. 20% Al_2O_3/Y-TZP. Examples of these results are shown in Fig. 5. Transmission electron microscopy of the microstructure of as-received Y-TZP reveals equilibrated, hexagonal-shaped grains with sharp apexes. The mean linear intercept grain size was about 0.3μm.

The presence of glassy (or liquid) phases can affect the deformation characteristics of structural materials, particularly at high temperatures, through enhancement of grain boundary mobility and grain boundary sliding. From a mechanistic standpoint, therefore, it is important to examine the grain boundary structure and chemical composition of superplastic Y-TZP and to determine if there exists a glassy phase at grain boundaries. Special efforts have been made to characterize the structure and chemical compositions at grain boundaries and in particular at grain boundary triple junctions.

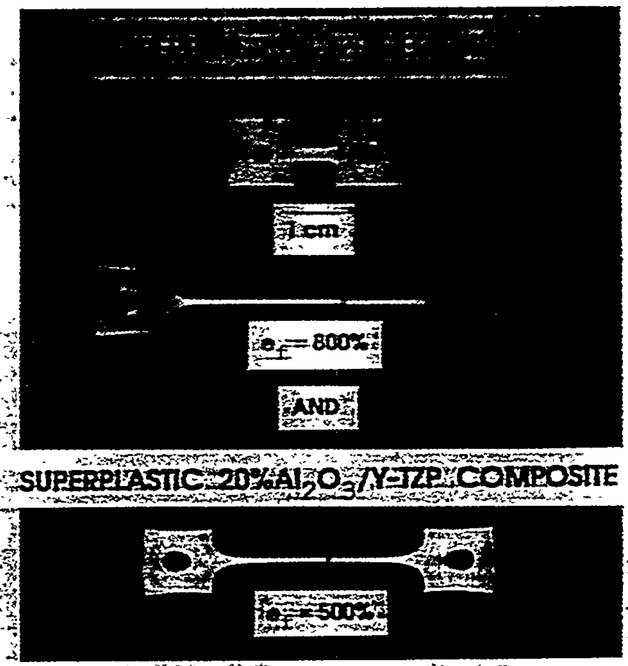

Fig 5. Superplastic behavior in an yttria-stabilized, fine-grained zirconia, and a 20% Al_2O_3/Y-TZP composite ceramic.

In the study of these grain boundary structures, high resolution electron microscopy was used. A representative high resolution lattice image of a grain boundary triple junction in Y-TZP is shown in Fig. 6. Lattice fringes from the three grains shown in Fig. 6 can be followed to their intersections at both the grain boundary interfaces and the triple junction, demonstrating the absence of any second phase. Nieh *et al.* also used Auger Electron Spectroscopy (AES) and X-ray Photoelectron Spectroscopy (XPS) to examine the chemical composition on intergranular fracture surfaces after superplastic deformation to determine if a low melting point amorphous phase containing either Si, Na, or Fe was present, but none was found. It was concluded that there is no requirement for the presence of a liquid phase in superplastic ceramics. However, the presence of a liquid phase at grain boundaries can affect the plastic flow in a superplastic ceramic.

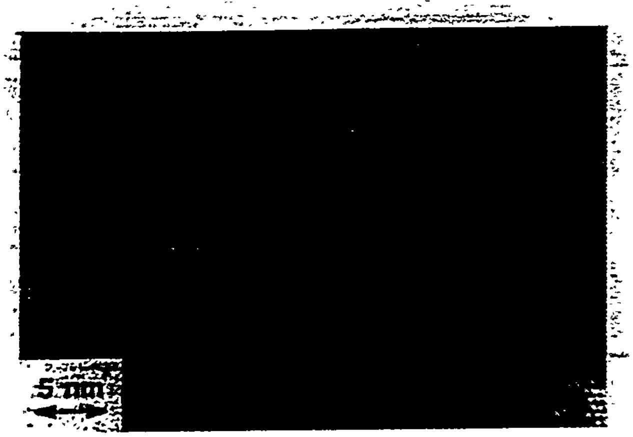

Fig 6. High-resolution lattice image of a grain boundary triple junction indicating the absence of any second phase.

In other recent studies on ceramic materials, Kim *et al.* have demonstrated superplasticity in a fine-grained iron carbide. The material was prepared from rapidly solidified powders of a 5.25%C-1.5%Cr hypereutectic iron. After compaction and extrusion at 1000°C, the microstructure consists of 80% of fine equiaxed grains of Fe_3C and 20% of iron-based material. A sample was deformed to 610% elongation at 1025°C, and at a true strain rate of $1\times10^{-4}s^{-1}$. The grain size after deformation remains equiaxed and fine at about 4μm. The continuous phase is the iron carbide material, and the second phase during superplastic flow is face-centered-cubic iron containing carbon in solution (austenite). In another recent study by Wakai, a silicon nitride material has been shown to exhibit superplastic behavior (100%). This silicon nitride-based particulate composite having an equiaxed grain structure was manufactured from ultrafine powders and consisted of a substantial amount of intergranular liquid phase.

Recently, the subject of superplasticity in intermetallics has started to receive attention. For example, Nieh and Oliver have achieved an elongation of 710% at 1080°C and a strain rate of $10^{-3}s^{-1}$ in a Ni_3Si intermetallic. Other examples of superplasticity in intermetallics are also emerging. For example, superplastic behavior has been reported in Ni_3Al-based alloys (640%), Fe_3Al (620%), Ti_3Al (>1000%), and TiAl (580%). It is noted that almost all superplastic intermetallics are two phase. Some single-phase materials were also claimed to be superplastic, but their elongation values (<200%) are much lower than those of two-phase materials. It is also important to point out that the grain sizes of some superplastic intermetallics are relatively coarse (>10μm), and their m values are close to 0.33. This will be addressed in Section 3.

The ease of solid-state bonding in fine-grained ultrahigh carbon (UHC) steels makes it possible to prepare ferrous laminated composites with sharp interfaces between layers. Such laminated composites have been shown to exhibit unique impact and superplastic properties. Another useful characteristic of the UHC steel laminated composites are their intermediate temperature ductility properties. Thus, it is possible to make non-superplastic mild steel behave in a superplastic-like manner at intermediate temperatures by lamination to superplastic UHC steel. Strain rate sensitivity exponents over 0.30 and elongations to fracture of over 400% were obtained. The curves of strain rate vs. stress show good agreement with constitutive equations for creep based on an isostrain deformation model. This model was used to predict the conditions of strain rate, temperature, and the percentage of non-superplastic component required to achieve nearly ideal superplasticity in a ferritic stainless steel clad to a UHC steel. It has been shown that the predicted conditions are achieved at 825°C and at $\dot{\varepsilon} = 10^{-3}s^{-1}$ for a UHC steel clad with a ferritic stainless steel (12% by volume). This combination of components and test conditions leads to the unexpected result that coarse-grained stainless steels can be made superplastic, as shown in Fig. 7 using gas pressure blow forming experiments with stainless steel and with stainless steel-clad ultrahigh carbon steel to assess die-filling capabilities.

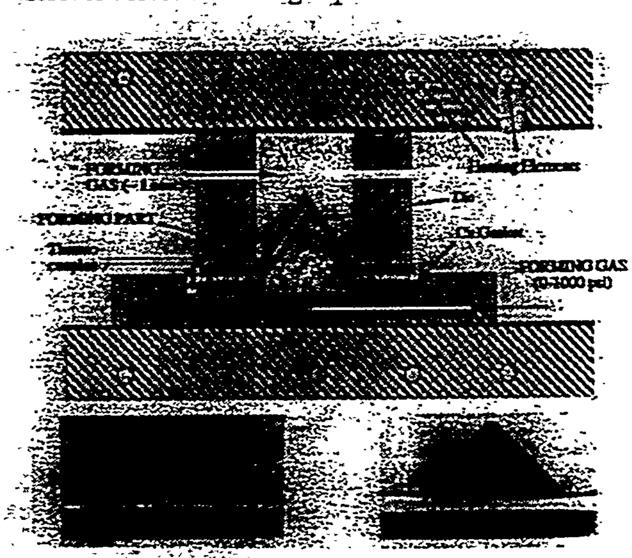

Fig 7. Superplastic forming of cones and examples for the cases of stainless steel (lower left) and superplastic laminated stainless steel/UHCS (lower right).

3. Materials Exhibiting Values of $m = 0.33$

Considerable progress has been made in achieving high elongations under very high strain rate conditions.[2] This success has been achieved by grain size refinement in whisker reinforced aluminum alloys and in mechanically alloyed aluminum alloys. In one example, a composite of Al-2124 containing 20 vol% SiC whiskers behaved in a superplastic-like manner (up to 300% elongation) at the high strain rate of $3 \times 10^{-1}s^{-1}$. In another example, an ultrafine grained, mechanically alloyed IN-9021 (4.2%Cu-20%Mg-1.1%C-0.8%O) exhibited a maximum elongation at strain rates as high as $20s^{-1}$. In another investigation on the same alloy, an optimum elongation of 1250% was found at a strain rate of $50s^{-1}$.

An overview of the superplastic behavior of the family of alloys described above is given in Fig. 8, in which elongation-to-failure is plotted as a function of strain rate. The grain sizes for each class of alloy groups is indicated on the figure. Also, included in the figure are data for mechanically alloyed nickel-base alloys (MA 754 and IN 6000) by Gregory *et al.* who noted high ductility at high strain rates in these fine grained materials. The subject of high ductility at high strain rates is important because conventional superplasticity is only found at relatively low strain rates, typically about 10^{-4} to $10^{-3}s^{-1}$. In contrast, superplasticity in these fine-grained alloys occurs at a much faster rate of about $10^{-1}s^{-1}$ which is similar to the rates for conventional forging. This rate will result in a great reduction of forming time, and is expected to have a significant technological impact on the commercial applications of superplastic materials. Wadsworth and Nieh are currently investigating the precise, operative deformation mechanisms in these alloys. The observation of superplastic-like behavior (m is slightly below the typical value of 0.5) at such high strain rates, however, is believed to be consistent with the extremely fine (submicron) grain size contained in these complex aluminum and nickel composites and alloys. An issue attracting interest is whether or not there is a contribution to deformation from the presence of thin liquid films formed by solute segregation at interfaces.

Class I solid solutions are a group of dilute alloys in which the glide segment of the glide/climb dislocation creep process is rate controlling because solute atoms impede dislocation motion. This group of alloys is also of interest in coarse-grained conditions for superplastic studies because, as a result of the glide-controlled creep mechanism, they have an intrinsically high strain rate sensitivity of about $m = 0.33$ (over certain temperature and strain rate ranges) and therefore can exhibit high elongations of over 200%. The intrinsic nature of the high strain rate sensitivity is important because it suggests that complex thermomechanical processing, such as that needed for fine-grained superplastic alloys, is unnecessary in Class I solid solutions.

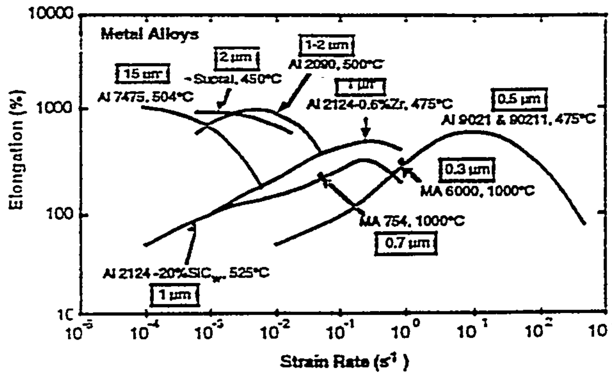

Fig 8. Overview of superplastic behavior as a function of strain rate for a range of Al-based materials of grain sizes from 15µm to submicron.

Because the strain rate sensitivity in Class I solid solutions is not as high as that found in classical superplastic materials (i.e. $m = 0.33$ versus $m = 0.5$ to 1.0), the elongations to failure are typically more modest, i.e. about 200 to 400%. It is probable that some early reports of superplasticity in coarse-grained alloys (e.g. Al-Mg and W-Re) are, in fact, the result of Class I solid solution behavior.[2]

A range of alloys (Table II) have been found that exhibit a strain rate sensitivity of $m \sim 0.33$ and that are also believed to fulfill the criteria for Class I solid solutions.[2, 3] These criteria include atomic size and modulus mismatch between the solute and solvent, as well as chemical diffusion considerations. It should also be noted that the addition of solute atoms is not the only way to cause dislocations to move in a viscous manner. Lattice friction effects may also reduce the glide mobility. This may be important in covalently-bonded solids (such as Ge and Si) as well as in ordered intermetallic compounds. In this latter group, strong repulsive forces exist between atoms of like character. Thus, the deformation behavior of B2 compounds (e.g. CoAl, NiBe, NiAl), in which lattice friction effects limit glide mobility, may be quite similar to that of Class I solid solutions. As mentioned previously, superplasticity occurs in some intermetallic alloys in a coarse-grained condition, for example, Fe_3Al and FeAl with grain sizes of about 100 and 200µm, respectively. The coarse-grained Fe_3Al and FeAl both have a strain rate sensitivity value near 0.33. They also show all the important microstructural features for Class 1 solid solutions, e.g., high dislocation density, absence of evidence for grain boundary sliding, no grain-boundary cavity formation, and samples that neck to a point. It was argued that, when disorder is introduced into an ordered intermetallic solid by the glide of a dislocation, the steady-state velocity is limited by the rate at which chemical diffusion can reinstate order behind the glide dislocation. In such cases, the deformation is essentially controlled by a viscous glide process.

TABLE II
ALLOYS EXHIBITING CLASS I
SOLID SOLUTION BEHAVIOR

Alloys	m	d (μm)	Max. el. (%)
W–33 at. % Re	0.2–0.3	50–400	260
Nb–10Hf–1Ti	0.33	75	>125
Nb–5V–1.25Zr	0.26	ASTM #8	170
Nb–29Ta–8W–0.65Zr–0.32C	0.28	NA[a]	>80
Al–5456 (Al–5Mg-based)	0.33	20	153
Al–4Mg–0.5Sc	0.3	0.5[b]	>1020
A1–2 to 4% Ge	0.3–0.6	100–200	200–260
Al–2.2Li–0.5Sc	0.32	3	500
Al–Ca–Zn	0.25	2	NA*
Mg–5.5Zn–0.5Zr	0.33	NA[a]	100
β-U (Ag, Al, Zn impurities)	0.33	100	680
Fe_3Al-Ti	0.3-0.4	100	625

[a] NA = Not available.
[b] Subgrain size.

The effective forming of near-net-shape parts from aluminum alloys is of significant interest for automotive and aerospace applications. It has traditionally been thought that the very high tensile ductilities necessary for many near-net-shape forming operations, i.e. in excess of 100%, were only available in fine-grain superplastic materials. Tensile ductilities in excess of 300%, however, have been attained in coarse-grain, non-superplastic, binary Al-Mg alloys as a result of a solute-drag-controlled dislocation glide process, in which the rate of diffusion of Mg solute-atom clusters around dislocations controls the creep rate. This process is distinctly different from the grain-boundary sliding process that leads to the extremely large (e.g. 500-1000%) elongations in superplastic materials. The high tensile ductility associated with solute-drag creep in coarse-grain materials is referred to as *enhanced ductility*, a term useful in distinguishing the behavior from classical superplasticity. Such enhanced ductilities from non-superplastic, Class I Al-Mg alloys might offer an inexpensive alternative to superplastic alloys that often require elaborate processing to develop fine grain sizes.

Recently, an experimental study has been made into the mechanical properties of two coarse-grain, binary aluminum-magnesium alloys, i.e. Al-2.8 wt.% Mg and Al-5.5 wt.% Mg.(4) Binary alloys were chosen in order to minimize the presence of cavity-inducing precipitates, thereby assuring that failure occurs by necking to a point. Tensile tests were conducted over a range of temperatures and strain rates in order to determine the effects of magnesium content, strain rate sensitivity, strain-hardening rate, and strength on tensile ductility. Data from both the binary single-phase alloys showed steady-state strain-rate sensitivities of $m = 0.33$ at temperatures of T ≥ 400°C over a wide range of strain rates. As stated earlier, a value of $m = 0.33$ is indicative of the solute-drag creep mechanism. Because little or no strain hardening is observed in the solute drag regime, the strain rate sensitivity controls tensile ductility. Ductility in these coarse-grain alloys is expected to be highest under conditions promoting solute-drag creep, where $m = 0.33$.

The diffusion-compensated strain rate, $\dot{\varepsilon}/D$ where $\dot{\varepsilon}$ is the true strain rate and D is the diffusivity of Mg in Al, can be used to correlate data from different temperatures and strain rates. It also, therefore, determines the value of m, which is the slope of this single curve for data plotted as the logarithm of flow stress versus the logarithm of strain rate. Because m controls tensile ductility, a plot of tensile elongation, e_f, against $\dot{\varepsilon}/D$ should yield a single curve. Such a plot is given in Fig. 9 with data from both binary alloys at various temperatures and strain rates. In general, the data taken even at different temperatures and strain rates give similar elongations when the values of $\dot{\varepsilon}/D$ are similar. It is demonstrated in Fig. 9 that ductilities of 200% to over 300% are obtained for both binary, coarse-grain Al-Mg alloys when deformed under conditions for which $\dot{\varepsilon}/D < 10^{-12}\ m^{-2}$. This $\dot{\varepsilon}/D$ value corresponds to the transition from power-law breakdown to solute-drag creep.

Also given in Fig. 9 are lines showing the dependence of strain rate sensitivity on $\dot{\varepsilon}/D$, with m values given on the right vertical rule. For both binary alloys, the increase in tensile ductility corresponds directly to the increase in strain rate sensitivity with decreasing $\dot{\varepsilon}/D$. This result is not surprising since the strain rate sensitivity is known to govern the rate of neck growth in such materials. Because failure in these high-purity, single-phase alloys occurs by necking to a point, as opposed to cavitation, tensile elongation is limited by the rate of neck growth. Further, the Al-2.8Mg and Al-5.5Mg materials, with linear-intercept grain sizes of 30 and 250μm, respectively, exhibited no observable difference in neck shapes between tests at similar values of $\dot{\varepsilon}/D$ when $m = 0.33$. The necks developed at failure in both Al-Mg materials were observed to be steeper than those typically observed in fine-grain superplastic materials; this would be expected from the differing m values (0.3 versus 0.5). In summary, the ductility of these alloys is controlled only by the mechanics of neck growth, which is governed by the strain rate sensitivity, m.

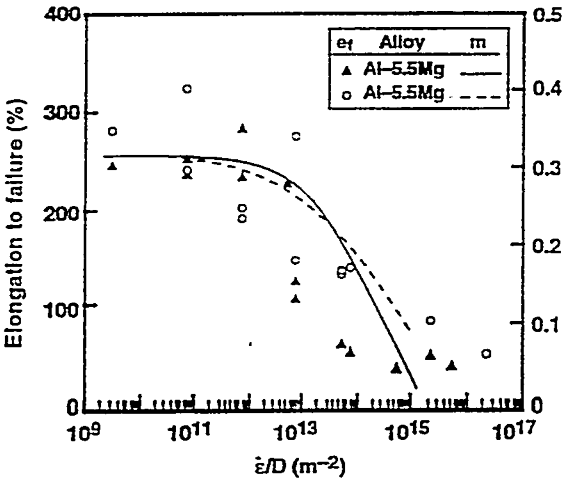

Fig 9. Tensile elongation to failure (left axis) and strain rate sensitivity (right axis) are plotted versus diffusion-compensated strain rate. The diffusion coefficient, D, for Mg diffusion in Al is given by $D_o = 5\times10^{-5}$ m^2/s with a corresponding activation energy of $Q = 136$ kJ/mol.

Numerical Predictions

By making a few simple assumptions, it is possible to construct a model of neck growth in strain-rate-sensitive materials. These assumptions include: (1) a small, slightly varying initial perturbation in cross-sectional area, (2) a slightly varying neck of long wavelength, (3) no strain hardening, and (4) power-law creep where σ, the flow stress, is proportional to $\dot{\varepsilon}^m$. Defining a failure condition will allow the use of this model in predicting tensile elongations in materials where failure is limited by neck growth. The failure condition was chosen to be the point at which the minimum cross-sectional area decreases to less than 90% of the maximum cross-sectional area. The numerical model was used with this failure condition to predict the tensile elongation of a strain-rate-sensitive material as a function of the strain rate sensitivity. Since the numerical model requires an initial perturbation in the cross-sectional area of the sample to initiate necking, a range of reasonable values was chosen: $\delta = 1.0\%$, $\delta = 0.5\%$, and $\delta = 0.1\%$. The results of these calculations are plotted in Fig. 10 as predicted tensile elongation versus strain rate sensitivity for the three choices of initial perturbation. Also plotted in Fig. 10 are the data for the binary Al-Mg alloys. The data are consistent with the predicted values for cases where $0.1\% < \delta < 1.0\%$. The agreement between predictions and test data indicates that the strain rate sensitivity governs tensile elongation in alloys which fail by necking.

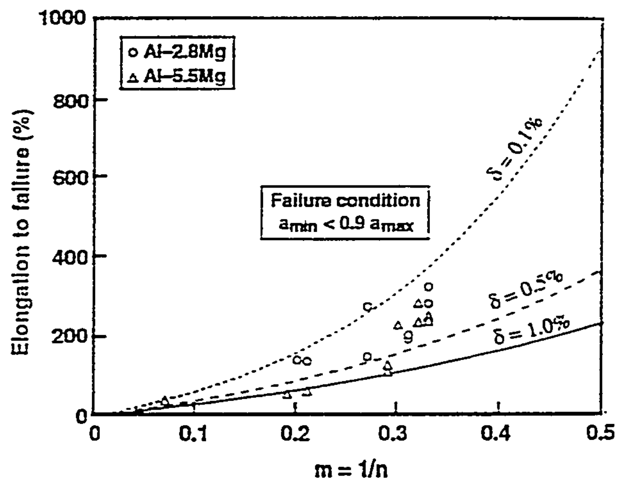

Fig 10. Predicted elongations-to-failure are shown vs. strain rate sensitivity, m, assuming that failure occurs when the minimum sample area is 90% of the maximum sample area. Predictions are made for three different initial perturbations in the sample area, δ.

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