

6-4 Aluminum

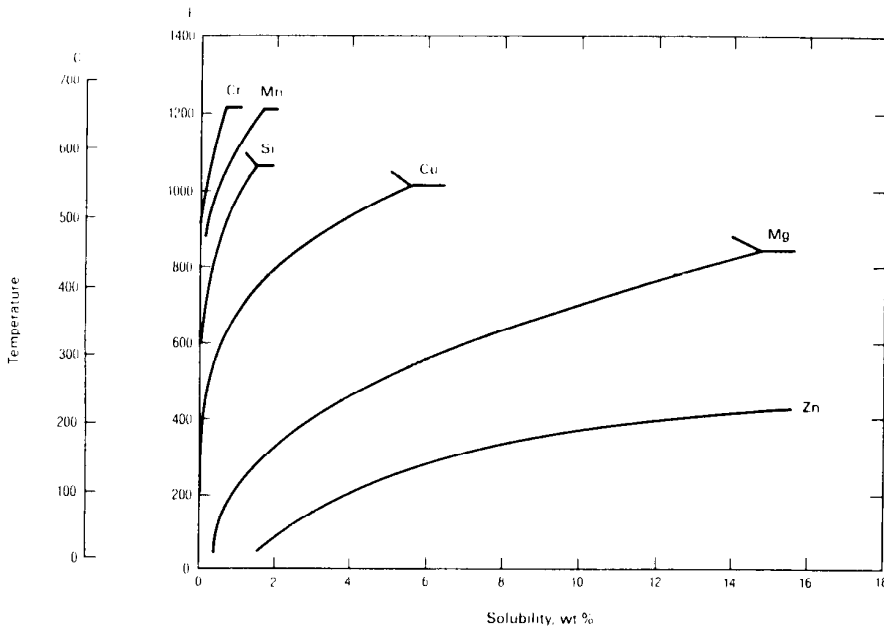


Fig. 1. Equilibrium binary solid solubility as a function of temperature for alloying elements most frequently added to aluminum

containing one or the other of these elements that is appropriate to the alloy composition. This precipitation does not cause appreciable hardening, nor is it intended that it should. Its purpose is to produce finely divided and dispersed particles that retard or inhibit recrystallization and grain growth in the alloy during subsequent heatings. The precipitate particles of $Al_2(Fe, Mn)$, Si , Al_2Cu_3Mn , or Al_2Mg_2Cr are incoherent with the matrix, and concurrent with their precipitation the original solid solution becomes less concentrated. These conditions do not provide appreciable precipitation hardening. Changes in electrical conductivity constitute an effective measure of the completeness of these precipitation reactions that occur in preheating.

The newer "in-line" or integrated processes that shorten the path from molten metal to wrought product, avoiding ingot preheating and reducing the over-all time-temperature history, are changing this conventional or traditional picture. It seems very probable that in order to obtain the best results from such processes, traditional alloy compositions should be adjusted taking into account the fact that larger proportions of these elements would be expected to remain in solid solution through such abbreviated and truncated thermomechanical operations. New capabilities may be obtained with currently standard alloys

in some instances, but it would not be expected that a particular alloy would exhibit the same properties when produced by the two types of processes.

For alloys that are composed of both solid-solution and second-phase constituents and/or dispersoid precipitates, all of these components of microstructure contribute to strength, in a roughly additive manner. This is shown in Fig. 2 for Al-Mg-Mn alloys in the annealed condition.

Non-Heat-Treatable Alloys. By definition, the group of commercial alloys that are classed as non-heat-treatable are those that are not appreciably strengthened by heat treatment—that is, show no effective precipitation hardening. The strengthening mechanisms discussed so far (solid-solution formation, second-phase microstructural constituents and dispersoid precipitates) are those that provide the basis for the non-heat-treatable alloys. Wrought alloys of this type are mainly those of the 3xxx and 5xxx groups containing magnesium, manganese and/or chromium as well as the 1xxx aluminums and some alloys of the 4xxx group that contain only silicon. Non-heat-treatable casting alloys are of the 4xxx or 5xxx groups, containing silicon or magnesium, respectively, and the 1xxx aluminums.

Strain Hardening. Strain hardening by cold rolling, drawing or stretching is a highly effective

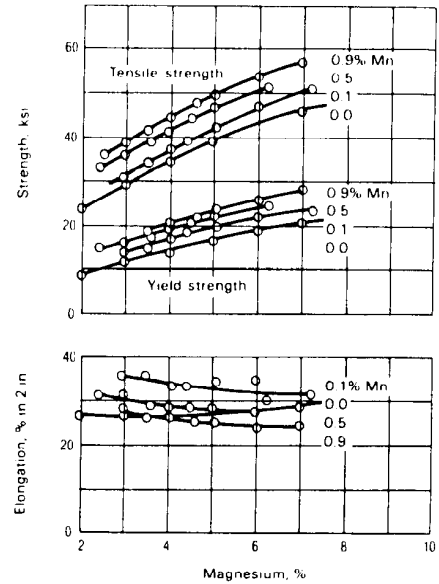


Fig. 2. Tensile properties in Al-Mg-Mn alloys in the form of annealed (O temper) plate 0.5 in. thick

means of increasing the strength of non-heat-treatable alloys. Work- or strain-hardening curves for several typical non-heat-treatable commercial alloys (Fig. 3) illustrate the increases in strength that accompany increasing reduction by cold rolling of initially annealed temper sheet. This increase is obtained at the expense of ductility as measured by percent elongation in a tensile test and by reduced formability in operations such as bending and drawing. It is frequently advantageous to employ material in a partially annealed (H2x) or stabilized (H3x) temper when bending, forming or drawing is required, since material in

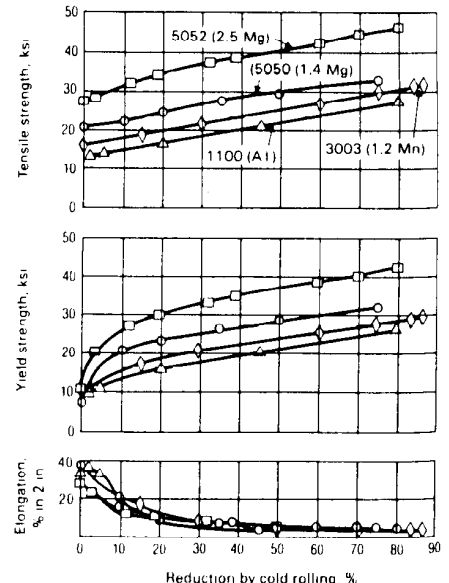


Fig. 3. Strain-hardening curves for aluminum (1100) and for Al-Mn (3003) and Al-Mg (5050 and 5052) alloys

Table 2. Solid-solution effects on strength of principal solute elements in super-purity aluminum(a)

| Element | Difference in atomic radii, $r_x - r_{Al}$, % (b) | MPa/at. % | Yield strength/% addition(c) | | Tensile strength/% addition(d) | | | | |
|---------|--|-----------|------------------------------|----------|--------------------------------|-----------|----------|----------|------|
| | | | ksi/at. % | MPa/wt % | MPa/at. % | ksi/at. % | MPa/wt % | ksi/wt % | |
| Si | -3.8 | 9.3 | 1.35 | 9.2 | 1.33 | 40.0 | 5.8 | 39.6 | 5.75 |
| Zn | -6.0 | 6.6 | 0.95 | 2.9 | 0.42 | 20.7 | 3.0 | 15.2 | 2.2 |
| Cu | -10.7 | 16.2 | 2.35 | 13.8 | 2.0 | 88.3 | 12.8 | 43.1 | 6.25 |
| Mn | -11.3 | (e) | (e) | 30.3 | 4.4 | (e) | (e) | 53.8 | 7.8 |
| Mg | +11.8 | 17.2 | 2.5 | 18.6 | 2.7 | 51.0 | 7.4 | 50.3 | 7.3 |

(a) Some property-percent addition relationships are nonlinear. Generally, the unit effects of smaller additions are greater. (b) Listed in order of increasing percent difference in atomic radii. (c) Increase in yield strength (0.2% offset) for 1% (atomic or weight basis) alloy addition. (d) Increase in ultimate tensile strength for 1% (atomic or weight basis) alloy addition. (e) 1 at. % of manganese is not soluble.

Tension and Compression Testing

Tension Testing

By George E. Dieter, University of Maryland

THE TENSION TEST is the test most commonly used to evaluate the mechanical properties of materials (Ref 1). Its chief use is the determination of properties related to the elastic design of machines and structures. In addition, the tension test gives information on a material's plasticity and fracture. The chief advantages of the tension test are that the stress state is well established, that the test has been carefully standardized (Ref 2) and that the test is relatively easy and inexpensive to carry out. This article will not detail this test technique, because it is well covered by standard methods (Ref 2); instead, the interpretation and limitations of the test results will be discussed.

ENGINEERING STRESS-STRAIN CURVE

In the conventional engineering tension test, stress is defined by the applied load divided by the original cross-sectional area of the specimen. Engineering strain, e , is the change in length divided by the initial length:

$$e = \frac{L - L_0}{L_0} = \frac{\Delta L}{L_0} \quad (\text{Eq 1})$$

In the elastic region of the stress-strain curve (Fig. 1), stress is linearly related to strain, $\sigma = Ee$, where E is the elastic modulus. As long as the specimen is loaded within the elastic region, the strain is totally recoverable and the specimen will return to its original length when the load is relaxed to zero. However, when the load exceeds a value corresponding to the yield stress, the specimen undergoes gross plastic deformation and is permanently deformed when the load returns to zero. The stress to produce continued plastic deformation increases with increasing strain—the metal strain hardens. To a good engineering approximation, the volume remains constant during plastic deformation, $AL = A_0L_0$, and, as the specimen elongates, it decreases uniformly in

cross-sectional area along its gage length. Initially the strain hardening more than compensates for this decrease in area, and the engineering stress continues to rise with increasing strain. Eventually a point is reached where the decrease in area is greater than the increase in deformation load arising from strain hardening. This condition will be reached first at some point in the specimen that is slightly weaker than the rest. All further plastic deformation is concentrated in this region, and the specimen begins to neck or thin down locally. Because the cross-sectional area is now decreasing far more rapidly than the deformation load is being increased by strain hardening, the engineering stress continues to decrease until fracture occurs.

The maximum in the engineering stress-strain curve corresponds to the ultimate tensile strength, σ_u . The strain at maximum load, up to which point the cross-sectional area decreases uniformly along the gage length as the specimen elongates, is the uniform elongation, e_u . For stretching-type forming operations, such as stretch forming of aircraft components or forming of automobile fenders, local necking determines the formability limit, and in such applications uniform elongation can be an important measure of ductility. In many metals, the engineering stress-strain curve is relatively flat in the vicinity of necking, and it may not be possible to establish the maximum load without ambiguity. In these cases, the method suggested by Nelson and Winlock (Ref 3) is useful.

DUCTILITY

The conventional measures of ductility that are obtained from the tension test are the engineering

Table 1. Comparison of several measures of ductility for two aluminum alloys (Ref 4)

| Alloy | e_0 | $e_{2.0}$ | e_u |
|----------------|-------|-----------|-------|
| 24S-O (2024-O) | 1.22 | 0.18 | 0.16 |
| 24S-T (2024-T) | 0.64 | 0.18 | 0.15 |
| 75S-O (7075-O) | 1.55 | 0.16 | 0.11 |
| 75S-T (7075-T) | 0.44 | 0.11 | 0.09 |

strain at fracture, e_f (usually expressed as a percentage elongation), and the reduction of area at fracture, RA (also usually expressed as a percentage):

$$e_f = \frac{L_f - L_0}{L_0} \quad (\text{Eq 2})$$

$$RA = \frac{A_0 - A_f}{A_0} \quad (\text{Eq 3})$$

Because an appreciable fraction of the deformation will be concentrated in the necked region of the specimen, the value of e_f will depend on the gage length L_0 over which the measurement was taken. The smaller the gage length, the greater the contribution from the neck and the higher the value of e_f .

To eliminate this difficulty and to provide a measure of ductility that correlates with forming operations in which the gage length is very short, it is possible to determine the zero-gage-length elongation, e_0 . From the constancy-of-volume relationship for plastic deformation, $AL = A_0L_0$:

$$\begin{aligned} \frac{L}{L_0} &= \frac{A_0}{A} = \frac{1}{1 - RA} \\ e_0 &= \frac{L - L_0}{L_0} = \frac{A_0}{A} - 1 \\ &= \frac{1}{1 - RA} - 1 = \frac{RA}{1 - RA} \end{aligned} \quad (\text{Eq 4})$$

Thus, the zero-gage-length elongation may be determined directly from the reduction of area at fracture or from the change in length of grid marks near the actual fracture. The data presented in Table 1 show how basing a comparison of the formability of aluminum alloys on the elongation in a 2-in. (50.8-mm) gage length would lead to erroneous conclusions for forming operations where local ductility determines the forming limit (Ref 4).

TRUE-STRESS/TRUE-STRAIN CURVE

The necking phenomenon which occurs in the tension test clouds the usefulness of the engineering stress-strain curve beyond the maximum load. The falloff in stress which occurs beyond P_{max} is artificial and occurs only because the stress continues to be calculated on the basis of the original cross-sectional area, A_0 , when in fact the area at the necked region is now much smaller than A_0 . If the true stress, based on the actual cross-sectional area of the specimen, is used, the stress-strain curve increases continuously up to fracture. Then, if strain is expressed as true strain, we have a plot called the true-stress/true-strain curve (Fig. 2).

Note that this curve continues to rise beyond necking all the way to fracture. However, once necking occurs, the constraints produced by the nondeforming region outside the neck produce a state of triaxial stress in the neck. Thus, the average stress required to cause flow from maximum load to fracture is higher than would be required if only a uniaxial stress were present. Bridgman (Ref 5) has made a mathematical analysis of the stresses in the neck that permits correction of the true-stress/true-strain curve for the

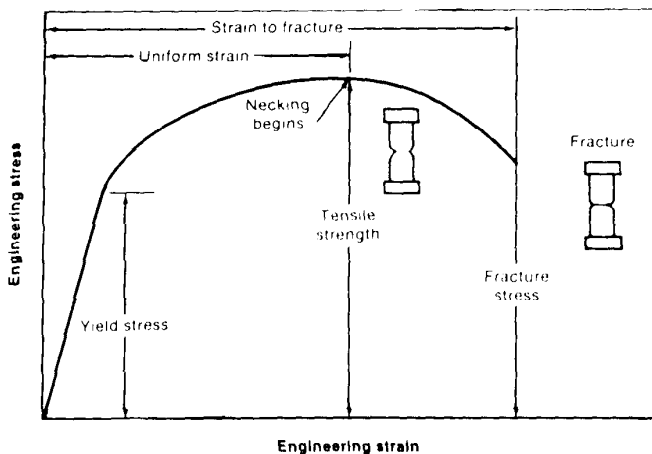


Fig. 1. Engineering stress-strain curve

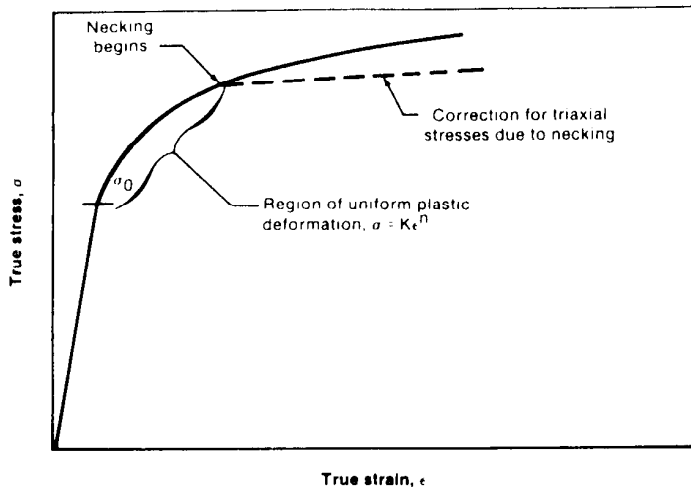


Fig. 2. True-stress/true-strain curve (flow curve)

existence of triaxial stresses. More recent studies have utilized finite-element analysis (Ref 6).

The region from yielding to necking is one of uniform plastic deformation in which the specimen gage length increases and the diameter decreases uniformly along the gage length. In this region the true-stress/true-strain curves for many ductile metals can be expressed by a power-law relation:

$$\sigma = K\epsilon^n \quad (\text{Eq 5})$$

where K is the strength coefficient, defined by the value of true stress at $\epsilon = 1.0$, and n is the strain-hardening exponent. Equation 5 is valid if a plot of σ versus ϵ is a straight line on log-log coordinates. The strain-hardening exponent is the slope of this line. Thus

$$n = \frac{d(\log \sigma)}{d(\log \epsilon)} = \frac{d(\ln \sigma)}{d(\ln \epsilon)} = \frac{\epsilon}{\sigma} \frac{d\sigma}{d\epsilon} \quad (\text{Eq 6})$$

The strain-hardening exponent may have values from $n = 0$ (a perfectly plastic solid) to $n = 1$ (an elastic solid). Values of n for most metals are between 0.05 and 0.50.

An increase in strain rate increases the stress to produce plastic deformation (the flow stress). The effect is modest for cold working, but is quite significant for hot working. The dependence of flow stress on strain rate, at constant strain and temperature, is given by

$$\sigma = C(\dot{\epsilon})^m |_{\epsilon, T} \quad (\text{Eq 7})$$

where m is the strain-rate sensitivity. The exponent m can be evaluated from the slope of a plot of $\log \sigma$ versus $\log \dot{\epsilon}$, or it can be obtained from rate-change tests in which the change in flow stress caused by a step change in $\dot{\epsilon}$ is measured:

$$m = \left(\frac{d \log \sigma}{d \log \dot{\epsilon}} \right)_{\epsilon, T} = \left(\frac{\Delta \log \sigma}{\Delta \log \dot{\epsilon}} \right)_{\epsilon, T}$$

$$= \frac{\log \sigma_2 - \log \sigma_1}{\log \dot{\epsilon}_2 - \log \dot{\epsilon}_1} = \frac{\log \frac{\sigma_2}{\sigma_1}}{\log \frac{\dot{\epsilon}_2}{\dot{\epsilon}_1}} \quad (\text{Eq 8})$$

ANALYSIS OF TENSILE INSTABILITY

The development of a necked region in a specimen loaded in uniaxial tension represents a plas-

tic instability. Because this disturbs the simple analysis of the tension test and limits the engineering usefulness of the test, it has become the subject of considerable study (Ref 7 to 9). A practical application of the ideas presented below is given in the work of Ghosh (Ref 10), who developed a numerical analysis for predicting the shape of the engineering strain-stress curve beyond maximum load as a function of strain hardening, strain-rate hardening, and plastic anisotropy properties of the metal.

Consider a tensile specimen loaded to a value P . At any point a distance L along the specimen, the cross-sectional area is A and $P = \sigma A$. Since P does not vary along the length of the specimen, and $\sigma = f(\epsilon, \dot{\epsilon})$,

$$\frac{dP}{dL} = 0 = A \left\{ \left(\frac{d\sigma}{d\epsilon} \right) \frac{d\epsilon}{dL} + \sigma \frac{dA}{dL} \right\} + \sigma \frac{dA}{dL} \quad (\text{Eq 9})$$

Because the volume of the specimen remains constant, the true strain can be written as

$$d\epsilon = \frac{dL}{L} = -\frac{dA}{A} \quad (\text{Eq 10})$$

and

$$\frac{d\epsilon}{dL} = -\frac{1}{A} \frac{dA}{dL} \quad (\text{Eq 11})$$

Also, from Eq 10, we can express the strain-rate $\dot{\epsilon}$ by

$$d\dot{\epsilon} = \frac{d\epsilon}{dt} = -\frac{1}{A} \frac{dA}{dt} = -\frac{\dot{A}}{A} \quad (\text{Eq 12})$$

so that

$$\frac{d\dot{\epsilon}}{dL} = -\frac{1}{A} \frac{d\dot{A}}{dL} + \frac{\dot{A}}{A^2} \frac{dA}{dL} \quad (\text{Eq 13})$$

The material parameters which are important to the necking process are the dimensionless work-hardening coefficient:

$$\gamma = \frac{1}{\sigma} \frac{d\sigma}{d\epsilon} \quad (\text{Eq 14})$$

and the strain-rate sensitivity:

$$m = \left(\frac{d \ln \sigma}{d \ln \dot{\epsilon}} \right)_{\epsilon} = \frac{\dot{\epsilon}}{\sigma} \left(\frac{d\sigma}{d\dot{\epsilon}} \right) \quad (\text{Eq 15})$$

When Eq 12 and 13 are substituted into Eq 9, and the definitions for γ and m are added through Eq 14 and 15, the result is

$$\frac{dA}{dL} (\sigma - m\sigma - \gamma\sigma) = -\frac{dA}{dL} \frac{m\sigma A}{A} \quad (\text{Eq 16})$$

A final rearrangement gives

$$\frac{1}{A} \frac{dA}{dL} = \frac{d}{dL} (\ln A) = \frac{m + \gamma - 1}{m} \quad (\text{Eq 17})$$

This equation describes the rate of change of area with length, and gives the criterion for the onset of necking.

Any real tension specimen will have variations in cross-sectional area along its length. These can arise from an intentional taper, from machining errors or from heterogeneities of structure which lead to weaker cross sections. Deformation becomes unstable when the smallest cross section of the specimen shrinks faster than the rest. This occurs when $dA/dA > 0$. Deformation will be uniform and stable when $dA/dA < 0$. Since A/A is negative in tension, stable deformation in tension occurs when $dA/dA \geq 0$. Therefore, from Eq 17, the condition for stable, uniform tensile deformation is

$$\gamma + m \geq 1 \quad (\text{Eq 18})$$

Necking is involved with the interplay between the applied stress and the flow resistance of the material. As the specimen elongates under a given load the area decreases and the stress increases. If necking is not to occur, the material's strength must increase through strain hardening (γ) and strain-rate hardening (m).

For room-temperature deformation, $m \rightarrow 0$ and the instability criterion reduces to $\gamma \geq 1$. Thus, stable tensile deformation occurs for

$$\frac{d\sigma}{d\epsilon} \geq \sigma \quad (\text{Eq 19})$$

If the true-stress/true-strain curve is given by $\sigma = K\epsilon^n$, then

$$\frac{d\sigma}{d\epsilon} = nK\epsilon^{n-1} = \sigma = K\epsilon^n$$

and necking occurs when

$$\epsilon = n \quad (\text{Eq 20})$$

Because n in tension rarely exceeds 0.5, we can see that the available uniform strain in the tension test is limited.

ELONGATION MEASUREMENTS IN TENSION TESTING

The measured elongation depends on the gage length or the dimensions of the cross section of the specimen. This is because the total extension consists of two components, the uniform extension up to the point of necking and the localized extension after necking (Fig. 3). The extent of uniform extension will depend on the metallurgical condition of the material (through n) and the

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