

Reliability and Failure of Electronic Materials and Devices

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The cover image is an artist's rendition of an electrostatic spark discharge event that lasts less than one billionth of a second. This miniature lightning bolt from a charged object can easily damage a semiconductor device. The photo is courtesy of Wayne Tan, Advanced Micro Devices.

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Chapter 7

Environmental Damage to Electronic Products

7.1 INTRODUCTION

The sources of IC damage considered up to this point originated within the materials and substances used in manufacturing. But now we confront degradation and damage to electronic products from outside sources such as humid atmospheres, airborne contaminant particles, and ionizing radiation. Even in the very early years of microelectronics it was already recognized that moisture and atmospheric particulates, often ionic in nature, posed serious reliability problems in devices, components, and packages. Thin-film metallizations and fine wires composed of corrodible metals (e.g., Al, Ag) and insulation (plastic, ceramics) prone to moisture absorption, resulted in degradation products that were uncomfortably large in size compared to the small feature dimensions involved. Today, humidity-induced corrosion of exposed or unprotected metals and water pickup, as well as release from polymers, are among the major failure mechanisms in electronic products.

In this chapter we explore a number of issues related to environmental damage starting with the sources of atmospheric contamination. Then the kinetics of permeation as well as leakage of harmful environmental gases (e.g., water vapor) and contaminants into electronic packages will be discussed. Later we shall see how particulates and humid environments act in synergy to promote corrosion and metal migration damage.

Quite different from the environmental damage to electronic products from atomic, molecular, and particulate matter is that wrought by exposure to ionizing radiation. The damage from this source normally causes instabilities in MOS transistor function through charging effects in the gate oxide. Errors in stored information due to impingement by cosmic rays and alpha particles emitted by radioactive elements is another common form of radiation damage that will be discussed at the end of the chapter.

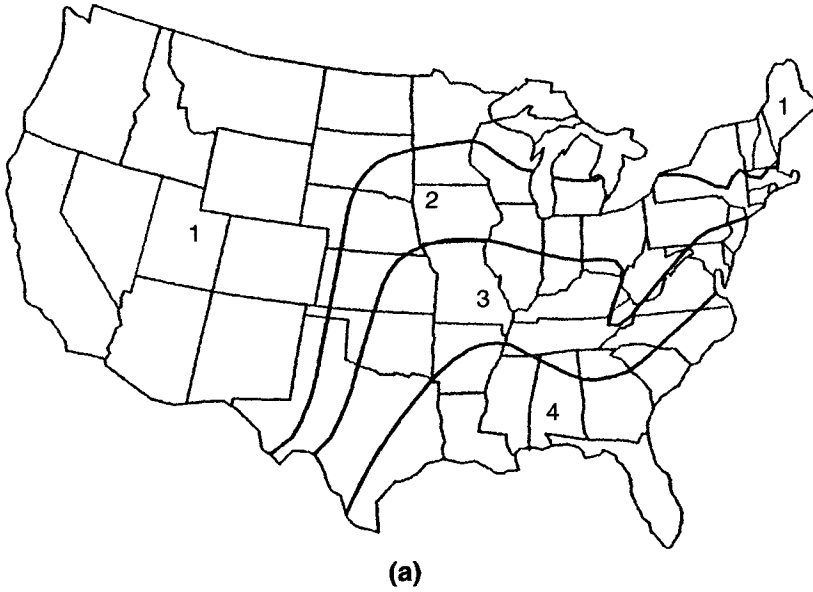
7.2 ATMOSPHERIC CONTAMINATION AND MOISTURE

7.2.1 Sources of Airborne Contamination

Important constituents of the outdoor environment include corrosive gases and particles (1). The compositions and concentrations of these harmful species vary with geographic location in the United States due to the distributions of humidity and air pollution, shown in Fig. 7-1. Included under air pollution is acid rain composed of raindrops (200–5000 μm in size) as well as fog droplets (0.5–50 μm), which in turn contain NH_4^+ , SO_4^{2-} , and NO_3^- ions. These and other ionic species are found in airborne particulates; a breakdown of their concentrations within Newark, New Jersey, indoor and outdoor ambients is given in Table 7-1.

The size distribution of atmospheric particles typically ranges from 0.1 to 15 μm . There are very many particles smaller than 0.1 μm , but their total mass is small. Within the 0.1 to 15 μm size range are coarse particles ($> 2.5 \mu\text{m}$) of mineralogical or biological origin that are usually formed by wind-induced abrasion. Their residence time in the atmosphere is typically hours to days, and they deposit by gravitational settling or wash out by the action of rain. As Table 7-1 notes, their concentration indoors is but a fraction of that outdoors, a fact explained by the use of modest filtration systems. On the other hand, particles smaller than 2.5 μm stem from natural sources such as volcanic ash and fossil fuels. In urban areas where fossil fuels are the dominant pollution source, small carbonaceous/oxide particles 0.05 μm in size are the seed nuclei for SO_2 and N_2O , which then oxidize to H_2SO_4 and HNO_3 . Ammonia released from fertilization and biological decay then neutralizes the acidified solid, all the while enlarging the particles to the stable 0.1 to 2.5 μm range. Such particles can remain airborne for months and even years through diffusional and convective transport. Importantly, the concentration ratio of fine particles within buildings to that outdoors is 0.2 to 0.5, depending on the filtration efficiency, the rate of air exchange, and the indoor sources of particles, e.g., due to smoking.

**Relative Humidity Risk
during Summer**



Air Pollution Risk

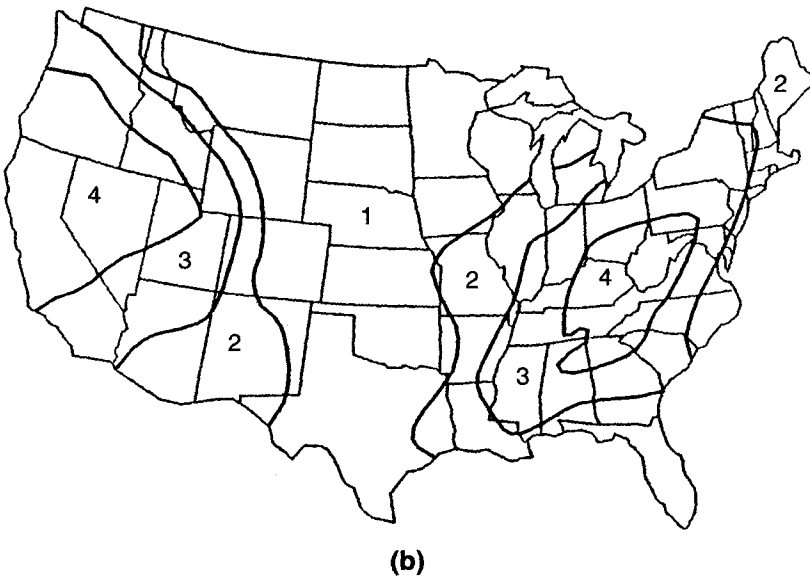


Fig. 7-1 Relative summer humidity (a) and air pollution (b) risks in different U. S. geographic locations rated from 1 (least severe) to 4 (most severe). After D. J. Sinclair, Lucent Technologies, Bell Laboratories Innovations.

Table 7-1 Average Concentrations of Ionic Species and Corrosive Gases in the Air Around Newark, NJ (ng/m³)

Particles	Indoor		Outdoor	
	Fine	Coarse	Fine	Coarse
Cl ⁻	4	4	66	502
SO ₄ ²⁻	721	25	5,213	725
Na ⁺	14	3	103	446
NH ₄ ⁺	168	0	1,631	1
K ⁺	19	2	53	30
Mg ²⁺	2	1	14	90
Ca ²⁺	10	20	25	297
TOTAL MASS	2880	6100	18,090	12,340

Gases	
Sulfur dioxide	5,400
Nitrogen oxides	24,000
Chlorine gases	510
Reduced sulfur	800
Ammonia	40,000

From R. B. Comizzoli, R. P. Frankenthal, R. E. Lobnig, G. A. Peins, L. A. Psota-Kelty, D. J. Siconolfi, and D. J. Sinclair, *The Electrochem. Soc. Interface* 26, Fall (1993).

The particulate and gaseous contaminants that manage to bypass filtration barriers in buildings land on the vertical and horizontal surfaces of electronic components and systems, where they have the potential to cause corrosion. In wafer fabrication facilities some finer particles even penetrate clean-room defenses and become the defects that lower yields.

It is instructive to predict the surface accumulation rate (A) of a given pollutant particle assuming the known outdoor concentration (C_o) and filtration efficiency (F). Filters are rated in terms of the ratio of contaminant level indoors (C_i) after passing through the filter, to the outdoor level; i.e., ($F = C_i/C_o$). Typically $C_i/C_o = 0.1$ for high-efficiency filters. An additional quantity that must be known is the particle terminal settling, or deposition, velocity, v , (Eq. 3-13). We now simply note that the accumulation rate is the product of the indoor concentration and deposition velocity, and in typical units

$$A (\mu\text{g}/\text{cm}^2\text{-yr}) = 0.526 C_i (\mu\text{g}/\text{m}^3) v (\text{cm}/\text{s}). \quad (7-1)$$

As an example consider the accumulation of SO_4^{2-} for which $C_o = 20 \mu\text{g}/\text{m}^3$. In passing through a filter for which $C_i/C_o = 0.12$, and assuming $v = 0.005 \text{ cm}/\text{s}$, substitution yields a surface accumulation of $6.31 \times 10^{-3} \mu\text{g}/\text{cm}^2\text{-yr}$. Measure-

ments in many cities across the U.S. typically reveal that annual sulfate and chloride accumulations on outdoor zinc and aluminum surfaces range between 0.1 to 0.8 $\mu\text{g}/\text{cm}^2$.

7.2.2 Moisture Damage

Moisture-induced damage (e.g., swelling of IC packages, electrical shorting) and corrosion of metal components (e.g., conductors, contacts) and auxiliary hardware (e.g., closed-loop laser cooling system (2)) associated with electrical and electronic equipment have traditionally been major reliability concerns. An annual monetary toll of a few billion dollars is not an exaggeration of the magnitude of moisture/corrosion problems in electrical equipment. Although our focus later will be corrosion in microelectronics, larger electrical and electronic systems are also quite vulnerable (3). Similarly, a review of corrosion damage in disk-recorder heads, aircraft avionic equipment, stepping motors, klystron tubes and other microwave equipment, printed wiring boards, antennas, contacts, and so on, can be found in reference 4. Since a necessary condition for corrosion is the presence of water, we start by exploring aspects of moisture accumulation on surfaces and permeation of electronic packages (5, 6).

7.2.2.1 Moisture Accumulation on Surfaces According to the kinetic theory of gases, a surface immersed in an environment containing a gas at pressure P will be impacted by a flux (Φ) of its molecules given by

$$\Phi = \frac{P}{(2\pi MRT)^{1/2}}$$

or

$$\Phi = \frac{3.5 \times 10^{22} P}{(MT)^{1/2}} \text{ (molecules/cm}^2\text{-s).} \quad (7-2)$$

In these expressions M is the molecular weight, the pressure is expressed in mm Hg or torr (1 torr = 1mm Hg = 133.3 Pa), and RT has the usual meaning. Before this formula is used, let us consider relative humidity (RH), which is defined as the ratio of the actual pressure of water vapor to the equilibrium vapor pressure at the given temperature. For water at 25°C the equilibrium vapor pressure is 24 torr; if, for example, RH = 42%, the value of P is 0.42×24 , or 10 torr. Substitution in Eq. 7-2 yields a flux of 4.8×10^{21} H_2O molecules/ $\text{cm}^2\text{-s}$ at this temperature. Since a solid surface contains some 10^{15} atomic sites per cm^2 , a monolayer of condensed water will form in approximately 0.2 μs , assuming that all impinging vapor molecules stick.

Although a monolayer film forms almost instantaneously on a metal surface, it does not grow much thicker than a few monolayers with time. Furthermore, the pure water film does not generally destroy the thin surface oxide layer that passivates and protects the metal. However, when airborne pollutants of reactive chlorides, sulfides, sulfates, and oxygen and nitrogen compounds simultaneously impinge on the surface, the trouble begins. As the latter dissolve in the thin moisture films, high concentrations of corrosive ions arise to initiate and perpetuate damage by the mechanisms listed in the next section.

Moisture poses different reliability problems in nonmetallic materials. Continuous water films containing ions (electrolytes) that coat ceramics and glasses are paths for surface leakage currents and may cause insulators to break down in applied electric fields.

7.2.2.2 Moisture Permeation of Bodies In porous electronic products permeation of moisture often creates reliability problems. Semiconductor chips are encased within packages that occasionally leak. Thus moisture and contaminants are able to permeate plastic, ceramic, and metal packages through microscopic interconnected pores, capillaries, channels, flaws, and even cracks in these materials and the glues, sealants, and solders used to seal them. In addition, moisture in packages arises from entrapped water due to poor bakeout procedures and is generated internally by decomposition of polymers. For example, moisture ingress in plastic packages has been identified to be the cause of “popcorn” cracking during soldering, a subject discussed later, in Sect. 8.2.3.2.

How large is package permeation or leakage, and how is it quantitatively characterized? Fortunately, package leaks typically do not pose a significant reliability hazard. A quantitative measure of the size of a leak is the leak flow rate (Q_s), defined to be the quantity of dry air that passes through the leak when there is atmospheric pressure on one side and vacuum on the other and expressed in units of atm-cm³/s or torr-liter /s. Leak rates can actually be calculated for certain leakage path geometries. Such calculations are done for vacuum system components and can be located in the vacuum science literature (7). They are based on either kinetic theory of gases or compressible fluid-flow, depending on whether the molecular or viscous gas-flow regime, respectively, is involved. Molecular flow occurs when the mean free path of gas molecules (λ) is much larger than the minimum dimensions ($d_{\min}$) of the leakage channel, i.e., $\lambda > d_{\min}$; molecules then collide with the channel walls more often than with each other. The leak rate for a gas of molecular weight M , at temperature T (K), in a cylindrical channel of length L (cm) and radius r (cm), is given by

$$Q_s \text{ (molecular)} = \frac{30.48 r^3}{L} \left(\frac{T}{M} \right)^{1/2} (P_1 - P_2). \quad (7-3)$$

In this equation, the units of Q_s are torr-liter/s, while P_1 and P_2 are the upstream and downstream pressures in torr, respectively.

The more complex case of viscous flow occurs when $d_{\min} > \lambda$. Now molecules collide with each other more frequently than with the walls of the channel. For gases of viscosity η (in Poise), the flow rate in torr-liters/s through a long circular channel of length L and radius is

$$Q_s \text{ (viscous)} = \left(\frac{\pi r^4}{16 \eta L} \right) (P_1^2 - P_2^2). \quad (7-4)$$

Transitional flow occurs when $\lambda = d_{\min}$ under conditions intermediate between molecular and viscous flow.

Now we consider the rate of pressure change caused by transport of a gaseous species into a package that does not contain gas initially. Assuming that ingress occurs by molecular flow, then

$$\text{Rate of ingress} = \frac{P_e Q_s}{P_o} \left(\frac{M_{\text{air}}}{M} \right)^{1/2}, \quad (7-5)$$

where P_o is atmospheric pressure, P_e is the external pressure, and M and M_{air} are the molecular weight of the species and air, respectively. Gas that entered through the leak can also escape via the leak. Therefore, if the partial pressure has reached P_i ,

$$\text{Rate of loss} = \frac{P_i Q_s}{P_o} \left[\frac{M_{\text{air}}}{M} \right]^{1/2}. \quad (7-6)$$

The net time (t) rate of gas pressure buildup or throughput is given by the difference, or

$$V \frac{dP_i}{dt} = (P_e - P_i) \frac{Q_s}{P_o} \left[\frac{M_{\text{air}}}{M} \right]^{1/2}. \quad (7-7)$$

Integration of this simple differential equation for a package of volume V yields

$$P_i(t) = P_e \left\{ 1 - \exp \left(- \frac{Q_s t}{P_o V} \left[\frac{M_{\text{air}}}{M} \right]^{1/2} \right) \right\}. \quad (7-8)$$

Similarly, it is easily shown that if the initial gas pressure within a package surrounded by a vacuum is $P_i(0)$, leakage will cause the pressure to decay according to

$$P_i(t) = P_i(0) \exp - \left[\frac{Q_s}{P_o V} \left(\frac{M_{air}}{M} \right)^{1/2} t \right]. \quad (7-9)$$

Diffusion of water through packages is a different way to view moisture ingress. Rather than molecular or viscous flow of water vapor through leakage channels, the diffusion equation (Eq. 5-2) models transport of H₂O through the package matrix with an "effective" diffusivity D . As an example, consider an epoxy-encapsulated package with a continuous concentration of moisture C_c at its surface $x = 0$. The moisture concentration buildup $C(x, t)$ at the chip surface located $x = h$ deep is given by

$$\frac{C(x, t) - C_i}{C_c - C_i} = \frac{4}{\pi} \sum_{n=0}^{\infty} \frac{1}{n} \sin \left(\frac{n \pi x}{2h} \right) \exp \left[- \left(\frac{n \pi}{2h} \right)^2 D t \right], \quad (7-10)$$

where C_i is the initial moisture content. Note that this approach incorporates the geometry of the package, which is not true of Eq. 7-8. In this case, the diffusivity depends on the geometric array of pores, capillaries, channels, and flaws in the epoxy.

7.3 CORROSION OF METALS

7.3.1 Introduction

Chief among the mechanisms of degradation in metals are the electrochemical reactions that lead to corrosion or loss of metal. The potential for metallic corrosion exists whenever two metals, one acting as a cathode, the other as an anode, are surrounded by an aqueous electrolyte and connected electronically so as to complete an electrical short circuit. These requirements can be physically manifested in many varied ways. Galvanic corrosion can often be easily diagnosed because two different metals are involved. But corrosion can also occur when two regions of a single metal are somehow different (e.g., due to differential deformation, defects, local compositional inhomogeneities). This makes it difficult to recognize all of the corrosion disguises whose ominous faces are responsible for the large toll of damage to electrical and electronic equipment in humid environments.

Extensive treatments of the science and engineering of corrosion can be found in a number of excellent texts (8, 9). As an introduction to the subject let us consider a metal M immersed in an acidic electrolyte. As the metal dissolves or corrodes, two reaction zones can be identified. The anodic reaction can be written as



where the loss of electrons signifies chemical oxidation. The electrons are then consumed in cathodic regions where ions are reduced or discharged. In acidic environments the cathodic reduction reaction



occurs. However, in basic electrolytes the production of hydroxyl ions via the reaction



is favored.

In the dissolution process above both anode and cathode reactions occurred in the same metal. Each of these so-called half-cell electrode reactions can be isolated, however, under ideal conditions. Associated with each is a potential E_{emf} that depends on the metal, nature of the surrounding electrolyte, ions present, and their concentration and temperature. For the anode reaction of Eq. 7-11, the Nernst equation yields

$$E_{emf} = E^o + \frac{RT}{nF_a} \ln (M^{+n}) = E^o + \frac{0.059}{n} \log (M^{+n}) \text{ at } 25^\circ\text{C}, \quad (7-14)$$

where E^o is the standard potential and F_a is the Faraday constant. A similar expression defines the electric potential at the other electrode. If the two half-cells are electrically connected, the individual emfs are added to yield the overall cell or reaction potential. Standard potentials (E^o) and reactions for a number of metals used in electrical applications are entered in Table 7-2. Relative positioning of the involved metals in this table determines the tendency to react chemically and corrode. The driving force for galvanic corrosion directly depends on the magnitude of the algebraic difference between the two half-cell potentials.

7.3.2 Pourbaix Diagrams

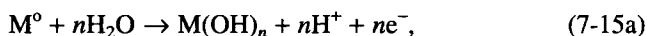
A useful way to graphically depict thermodynamic aspects of aqueous metallic corrosion is through Pourbaix diagrams (8). These diagrams, so named for their developer, map the oxidizing power, or potential, (E) of electrolyte solutions versus their acidity or alkalinity (pH). The result is a kind of metal–water phase diagram with boundaries defining regions of stable phases (e.g., ions, oxides, compounds) that either promote the tendency of metals to corrode or remain passive. Construction of Pourbaix diagrams is straightforward in principle, but involved in practice. The reactions of metals with water are pertinent, and there are essentially three kinds, defined by Eq. 7-11, by

Table 7-2 Standard Electrode Potentials at 25°C

Electrode Reaction	Electrode Potential E° (Volts) (relative to standard hydrogen)
$\text{Au} \rightarrow \text{Au}^{3+} + 3\text{e}^-$	+1.498
$\text{H}_2\text{O} + \text{Cl}^- \rightarrow \text{H}^+ + \text{HOCl} + 2\text{e}^-$	+1.49
$2\text{Cl}^- \rightarrow \text{Cl}_2 (\text{g}) + 2\text{e}^-$	+1.36
$2\text{H}_2\text{O} \rightarrow \text{O}_2 + 4\text{H}^+ + 4\text{e}^-$	+1.229
$\text{Pt} \rightarrow \text{Pt}^{2+} + 2\text{e}^-$	+1.200
$\text{Ag} \rightarrow \text{Ag}^+ + \text{e}^-$	+0.799
$\text{Fe}^{2+} \rightarrow \text{Fe}^{3+} + \text{e}^-$	+0.771
$4(\text{OH})^- \rightarrow \text{O}_2 + 2\text{H}_2\text{O} + 4\text{e}^-$	+0.40
$\text{Cu} \rightarrow \text{Cu}^{2+} + 2\text{e}^-$	+0.337
$\text{Sn}^{2+} \rightarrow \text{Sn}^{4+} + 2\text{e}^-$	+0.150
$\text{H}_2 \rightarrow 2\text{H}^+ + 2\text{e}^-$	0.000
$\text{Pb} \rightarrow \text{Pb}^{2+} + 2\text{e}^-$	-0.126
$\text{Sn} \rightarrow \text{Sn}^{2+} + 2\text{e}^-$	-0.136
$\text{Ni} \rightarrow \text{Ni}^{2+} + 2\text{e}^-$	-0.250
$\text{Co} \rightarrow \text{Co}^{2+} + 2\text{e}^-$	-0.277
$\text{Cd} \rightarrow \text{Cd}^{2+} + 2\text{e}^-$	-0.403
$\text{Fe} \rightarrow \text{Fe}^{2+} + 2\text{e}^-$	-0.440
$\text{Cr} \rightarrow \text{Cr}^{3+} + 3\text{e}^-$	-0.744
$\text{Zn} \rightarrow \text{Zn}^{2+} + 2\text{e}^-$	-0.763
$\text{Al} \rightarrow \text{Al}^{3+} + 3\text{e}^-$	-1.662
$\text{Na} \rightarrow \text{Na}^+ + \text{e}^-$	-2.714

Cathodic behavior or reduction occurs more readily the *higher* the electrode reaction is in the table.

Anodic behavior or corrosion occurs more readily the *lower* the electrode reaction is in the table.



and by



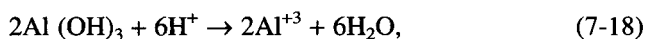
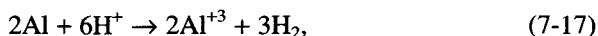
Let us consider the case of a metal undergoing the reaction defined by Eq. 7-15b.

By the Nernst equation and the definition of pH as $-\log (\text{H}^+)$,

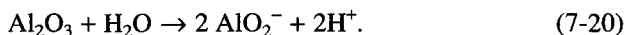
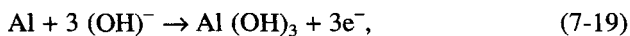
$$\begin{aligned} E_{\text{emf}} &= E^\circ + \frac{0.059}{n} \log \{(\text{MO}_n^{n-})(\text{H}^+)^{2n}\} \\ &= E^\circ + \frac{0.059}{n} \log (\text{MO}_n^{n-}) - 0.118 \text{ pH}. \end{aligned} \quad (7-16)$$

A plot of E_{emf} versus pH yields a line with a slope of -0.118 .

As a further example, we consider aluminum, arguably the most vulnerable metal in microelectronics from a corrosion standpoint. Reactions in aqueous solutions include



and



The first two reactions occur in acidic solutions. On the other hand, the latter two reactions are favored when the solutions are basic and when Na^+ and K^+ impurities are present to raise the local pH. Importantly, the potential in each of these reactions can be cast into the form of Eq. 7-16 and plotted as E (relative to the standard hydrogen electrode) versus pH lines once it is recalled that $\text{pH} + \text{pOH} = 14.00$. Proceeding in this way, the resultant Pourbaix diagram for aluminum is shown in Fig. 7-2, where the following items are noteworthy:

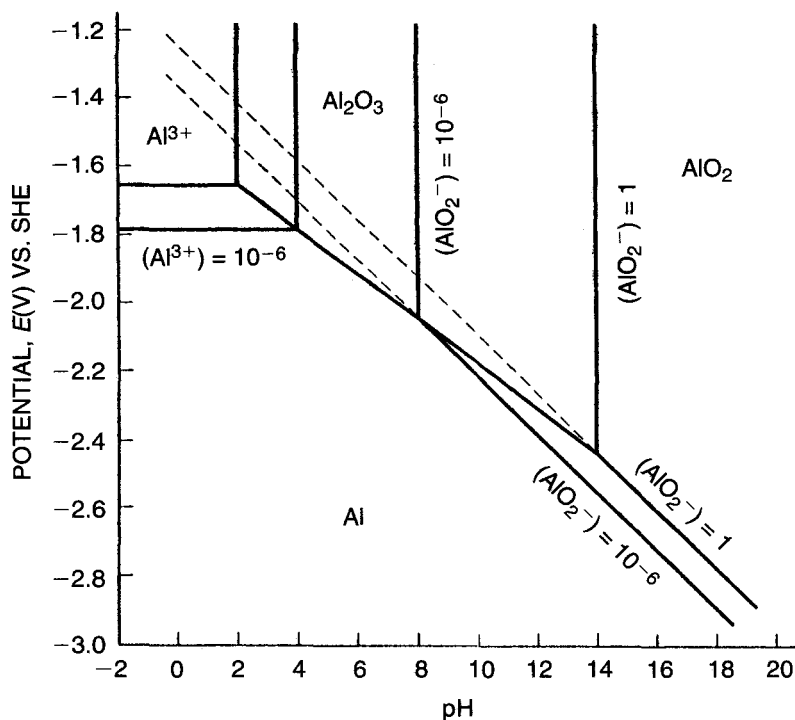


Fig. 7-2 Pourbaix diagram for aluminum. Lines are shown for arbitrarily, but frequently chosen, ionic concentrations of 1 and 10^{-6} . From reference 8.

1. Aluminum is amphoteric, and the protective oxide film dissolves at both low and high pH values. In acids, aluminum dissolves, forming Al^{3+} ions, while in basic solutions corrosion occurs with the generation of AlO_2^- ions.
2. There is a region of passivity extending between pH 4 to 8, where aluminum is protected by Al_2O_3 . Passivation also requires a potential above $\sim -1.8\text{V}$.
3. Lower concentrations of ions containing Al compress the field of oxide stability.
4. Aluminum metal is immune to corrosion at potentials that lie below those indicated by the lowest lines on the Pourbaix diagram.

7.3.3 Corrosion Rates

In assessing reliability it is the rates of corrosion that are of interest. Corrosion rates are not predictable from thermodynamic considerations and cannot be determined from Pourbaix diagrams. Rather complex processes at the metal–electrolyte interface generally govern the speed of corrosion reactions, and these proceed at finite rates. If the discharge of hydrogen is not rapid enough to accommodate all of the electrons provided, the cathode surface charges negatively. The negative potential that develops is known as *cathodic polarization*. Similarly, *anodic polarization*, or a positive potential, develops at the anode because of a depletion of electrons. And with greater electron deficiencies, higher anodic polarizations are produced. This, in turn, favors more dissolution of metal or corrosion. Thus we may view anodic polarization as the driving force for corrosion. What happens is that the surface reaches a steady-state potential that depends on the rate at which electrons are exchanged between anode and cathode reactions. Without polarization the smallest driving force would result in infinitely large electrochemical reaction rates.

Irrespective of the specific metals or aqueous environment, Faraday's law determines the extent of damage done. Thus, the passage of 96,500 Coulombs (C) of charge will remove a gram equivalent weight of anode metal and discharge a like weight of ions at the cathode. Because corrosion occurs heterogeneously, corrosion currents are usually unknown. This makes it difficult to predict the extent of corrosion damage and the life of involved components.

7.3.4 Damage Manifestations

Corrosion damage in microelectronics is similar to that which occurs in materials used in mechanically functional applications. Although the morphology of corrosive attack can vary appreciably, there are several basic mechanisms, and these are presented in turn.

7.3.4.1 Uniform Corrosion In this case corrosion damage occurs uniformly over large areas of metal or alloy surfaces. Low-carbon steels and copper alloys are subject to such attack. In the *passive* state the rate of corrosion is essentially negligible. Protection is typically provided by a thin oxide film. However, if passivity is destroyed by puncturing the oxide, the metal becomes *active*, with effective anodic and cathodic sites constantly shifting, so that metal loss is uniform. Although it is relatively easy to protect against uniform corrosion, examples of failures by this mechanism are generally rare.

7.3.4.2 Galvanic Corrosion Two different metals (alloys) are required for galvanic action, perhaps the easiest form of corrosion to recognize. When dissimilar metals in electrical contact are immersed in an electrolyte, an electrochemical cell (i.e., a battery) is produced. The more active metal becomes the negatively charged electrode, or anode, while the more noble metal becomes the positively charged electrode, or cathode.

Galvanic corrosion is magnified when the anode is small and the cathode large. Examples of such a size effect occur commonly in electronics when electrodeposited noble metal coatings are used to protect less noble substrates, e.g., Au on Ni. If the Au coating is porous or punctured at a point, Ni may rapidly corrode. The reason has to do with conservation of current i (A) in both electrode reactions. Thus, if j is the current density (A/cm²) and A_o is the electrode area, the condition that $i(\text{Au}) = i(\text{Ni})$ means that $j(\text{Au})A_o(\text{Au}) = j(\text{Ni})A_o(\text{Ni})$. Because $A_o(\text{Au}) > A_o(\text{Ni})$, $j(\text{Ni}) > j(\text{Au})$; the high current density causes the small unprotected Ni region to rapidly dissolve, while the cathode reaction is collectively borne by the large coating area.

7.3.4.3 Pitting Corrosion As the description implies, pit or hole formation is the form of attack during pitting corrosion. Localized pitting attack generally occurs while the bulk of surface is passive. Chloride, and other halogen ions to a lesser extent, are notorious for initiating pitting attack in otherwise passive metals and alloys. A suggested mechanism for the pitting of stainless steel in the presence of Cl^- ions is shown in Fig. 7-3. Corrosion begins with the loss of passivity caused by the local destruction of the thin chromium oxide passive film. A copious production of Fe^{2+} ions attracts Cl^- , where hydrolysis occurs by the reaction



In the process, $\text{Fe}(\text{OH})_2$ forms, and the pH is lowered at the pit initiation site. The acid chloride solution further accelerates anodic dissolution, which in turn concentrates more chloride at the pit, etc., in this so-called self-propagating

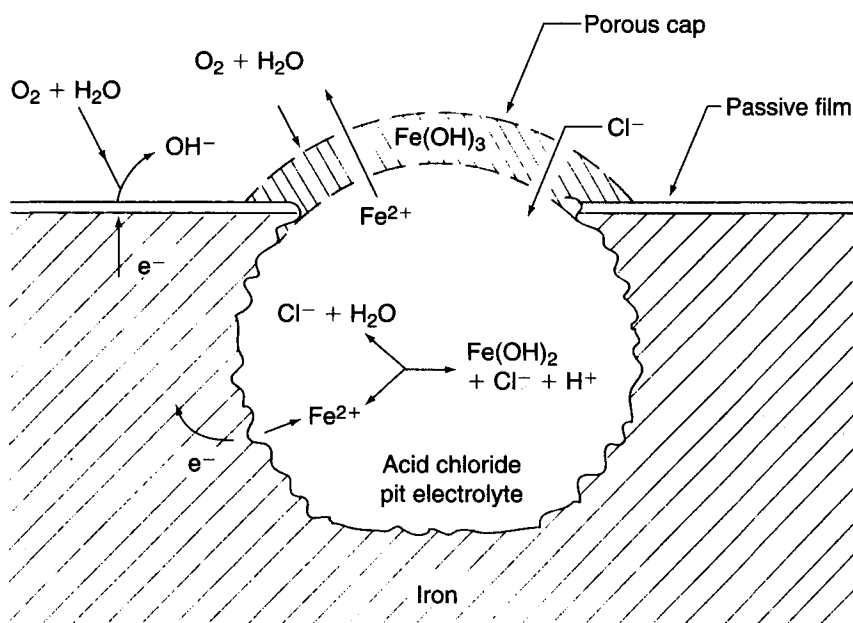


Fig. 7-3 Model for the pitting of stainless steel in the presence of chloride ions.

autocatalytic reaction. In addition, Fe^{2+} oxidizes to Fe^{3+} , so that a porous cap of rust ($\text{Fe}(\text{OH})_3$) covers the pit hole. This encourages Cl^- ions to migrate in order to balance the excess positive charge generated by metal oxidation. All the conditions are now fulfilled for continued pitting corrosion. Similar pitting reactions occur for the nickel and chromium atoms in stainless steel.

Crevice corrosion also causes local attack of metal that is similar to pitting damage. Metal surface regions hidden under washers, gaskets, sealing rings, surface scale, switch contacts, rivets, etc., serve as crevices. Restricted oxygen supply to the crevice diminishes the cathode reaction of Eq. 7-13.

7.3.4.4 Fretting Corrosion Tin, plated solder, and other electrical contact materials often suffer corrosion-accelerated wear when the contacting metals undergo small-amplitude, cyclic rubbing or fretting. Particles chemically dissolve and mechanically detach, leaving a surface geometry behind that generally increases the contact resistance. Fretting effects at electrical contacts are addressed again in Chapter 9.

7.3.4.5 Stress Corrosion Cracking (SCC) Corrosion in the presence of tensile stress accelerates the damage to the metal surface. The mechanisms are

not well understood, but it is thought that applied or internal stress causes atoms to leave the anodic tips of incipient cracks, thereby sharpening them. Usually, particular environment-metal combinations are associated with SCC, e.g., chlorides for stainless steel, ammonia for copper, and H_2S for low-carbon steel. Residual thermal stresses, or those induced by bending electrical lead wires, in combination with moisture serve to promote such cracking. When the loading is cyclic rather than static, the stress-corrosion damage is known as *corrosion fatigue*.

7.3.4.6 Atmospheric Corrosion This form of corrosion attack stems from the environment, but without the presence of a bulk electrolyte. Instead, the electrolyte is usually a thin film of moisture that condenses from the atmosphere when the relative humidity is sufficiently high. The electrolytic process of metal migration on nominally insulating substrates occurs in this manner (Sect. 7.4).

7.4 CORROSION IN ELECTRONICS

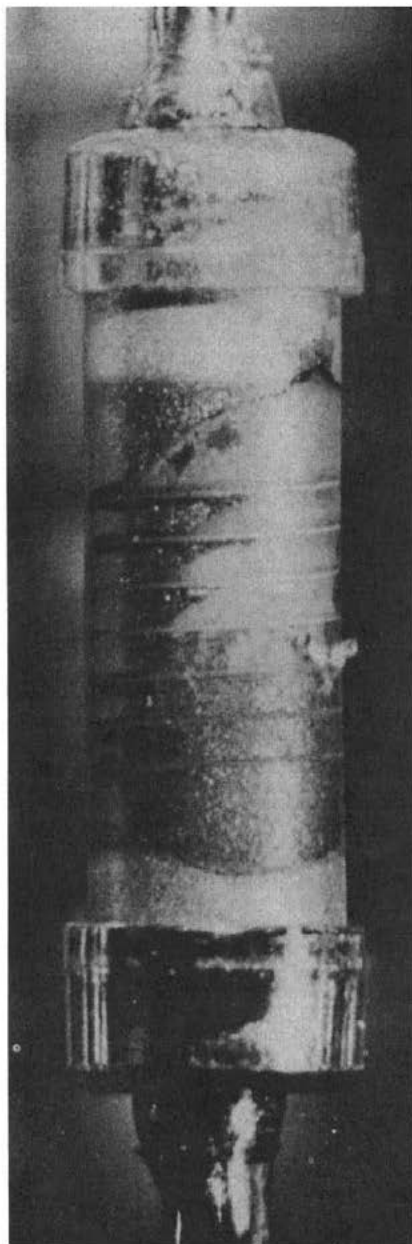
7.4.1 Overview

After having just introduced some of the classic manifestations of corrosion, we now focus on corrosion failure in electronic components. A couple of corrosion case histories are reproduced in Fig. 7-4 and discussed in the figure captions. In microelectronic devices the scale of damage is far smaller but no less destructive. Vulnerable metal structures are thin film-contacts, interconnect lines, and wire or solder bonding pads (10, 11). These structures are often coated with glassy inorganic and/or organic polymers, which not only provide electrical insulation but mechanically protect and chemically passivate devices as well. However, defective coatings and packages render unprotected metals vulnerable to chemical attack from moisture as well as atmospheric and encapsulant contaminants.

An old but still instructive review by Schnable *et al.*(12) of corrosion in IC chips encased within hermetically sealed and plastic-encapsulated packages has usefully classified corrosion as

1. chemical (where no bias voltage is present),
2. electrochemical (where a dc bias voltage is applied and currents flow), and
3. galvanic (involving dissimilar metals).

The role of the passivating layers is of crucial significance in the corrosion of underlying metals. Pinholes and cracks in these brittle dielectric films (SiO_2 ,



(a)

Fig. 7-4 Examples of corrosion-related problems in electronics. (a) Core of failed nichrome-film resistor showing missing metal. The resistor was damaged by excessive heat applied during the lead retinning operation. This opened the resistor epoxy end-cap seal and allowed entry of chlorides from flux and cleaning solvents. Because chlorine was detected in the region of nichrome loss, the role of this element in fostering corrosion attack was confirmed. From J. D. Guttenplan, *Materials Performance*, p. 76, April (1990).



(b)

Fig. 7-4 (b) Corrosion of copper on a printed-circuit board plated through-hole. Prior to soldering, the copper was mechanically brushed, generating metal flow and surface crevices. Flux, entrapped in the crevices, remained after the Sn-Pb solder covered the Cu (see Sect. 8.3.3.2). The combination of different metals in contact, salts in the flux, and a humid environment resulted in galvanic action with the formation of blue (CuSO_4) corrosion products surrounding the lower half of the hole. Courtesy of Manko Associates.

silicon nitride) allow the ingress of moisture and contaminants, while their residual internal stresses may promote stress corrosion cracking of the metallizations. Corrosion on the chip level is largely a processing problem because metallizations are exposed then. Once metal structures are passivated, corrosion largely ceases to be a reliability issue. On the packaging level, ionic contaminants (e.g., Cl^-) provided a substantial corrosion hazard for thin-line metallizations in early molded plastic packages. However, today's epoxy molding materials have a lower concentration of corrosive ions, so that failures are infrequent.

7.4.2 Accelerated Corrosion Testing

The observation that aluminum metallization on integrated circuits is prone to corrosion has led to accelerated testing at elevated temperatures and relative humidity (RH) levels. Results distilled from such testing reveal that mean failure times can be described by

$$\text{MTTF} = AV^{-m} (\text{RH})^{-n} \exp \left[\frac{E_c}{kT} \right], \quad (7-22)$$

where V is the applied voltage, E_c is an effective activation energy, A , n , and m are constants, and kT has the usual meaning. If the test voltages, circuit designs, and specimen geometry and spacings are constant, Peck (13) has suggested that

$$\text{MTTF} \approx (\text{RH})^{-3} \exp \left[\frac{0.8 \text{ eV}}{kT} \right]. \quad (7-23)$$

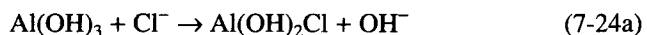
Like similar formulas that describe electromigration (Eq. 5-31) and dielectric breakdown (Eq. 6-10), Eq. 7-22 is quite useful in design and for making life projections. Unfortunately, however, little insight into the kinetics of the degradation process can be gleaned from such phenomenological formulas.

While early test conditions varied widely, much testing is presently carried out at 85°C and 85% RH. Highly accelerated stress tests (HAST) are popular in order to accentuate the adverse roles humidity and temperature have on corrosion; otherwise, testing times are impracticably long. Results of a wide survey of available corrosion-failure data for aluminum in epoxy packages are shown in Fig. 7-5. Relative to the standard 85°C/85% RH test conditions, higher temperatures shorten failure times according to Eq. 7-23.

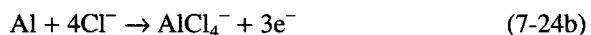
7.4.3 Corrosion of Specific Metals

7.4.3.1 Aluminum Corrosion effects have been widely studied in aluminum and Al base alloy metallizations. As we have already seen, Al suffers corrosion damage in the presence of water. Even the passive, amorphous native-oxide layer can undergo hydrogen blistering damage by the reaction sequence depicted in Fig. 7-6 (14). Most commonly, however, Al corrodes at both anodic and cathodic sites in the presence of contaminant ions. For example, chloride ions from the atmosphere, solder fluxes, chlorinated hydrocarbons (Freon, trichloroethylene), solvents, etc., accelerate a pitting-type anodic corrosion reaction that is autocatalytic via the following cycle:

1. Dissolution of protective hydrated oxide upon adsorption of Cl^- ions



2. Chloride attack of Al



3. Regeneration of Cl^-

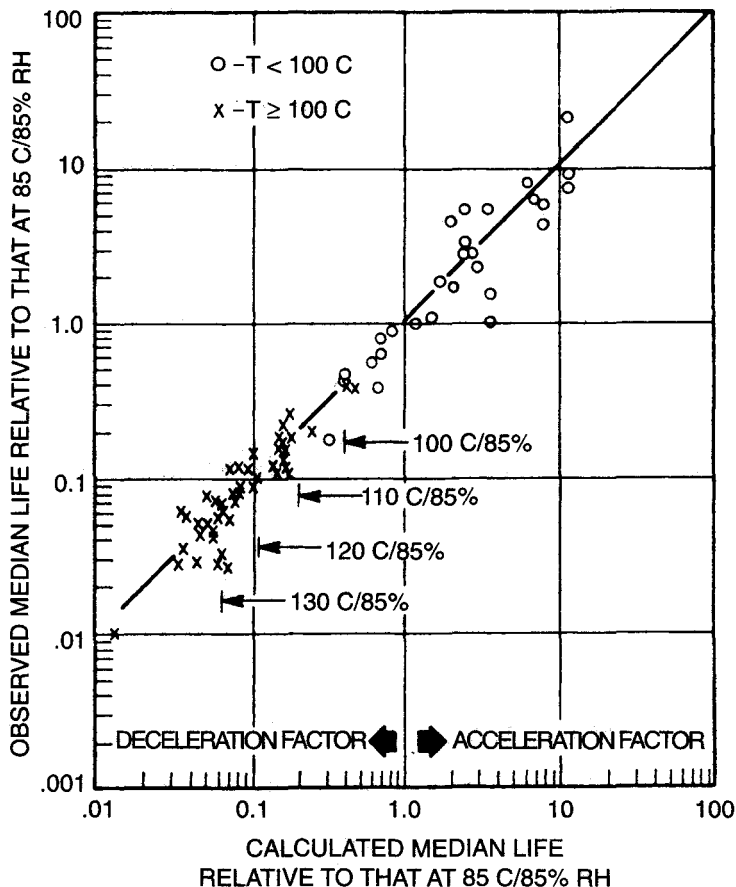
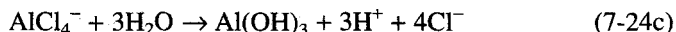


Fig. 7-5 Ratio of the life of aluminum metallization in epoxy packages obtained in humidity tests relative to 85°C/85% RH conditions. From reference 13.



Corrosion problems in IC chips can start during processing. For example, during patterning of thin-film interconnects by reactive plasma etching, gases containing chlorine are used, which can leave halide residues behind to contaminate photoresists and metallizations. Exposure to humidity then sets the above chemical reactions in motion. In addition, cathodic corrosion occurs when Al interconnects contact phosphosilicate glass films that contain excessive phosphorus levels. For reasons explained in Chapter 5, aluminum metallizations contain copper, and sometimes silicon. This enhances the prospects for galvanic corrosion, which takes the form of intergranular corrosive attack and loss of Al (15) in Al-Cu

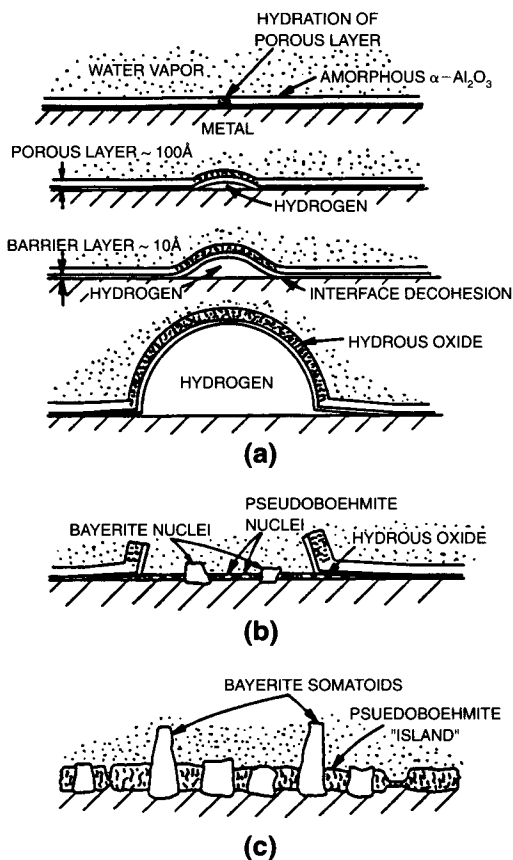
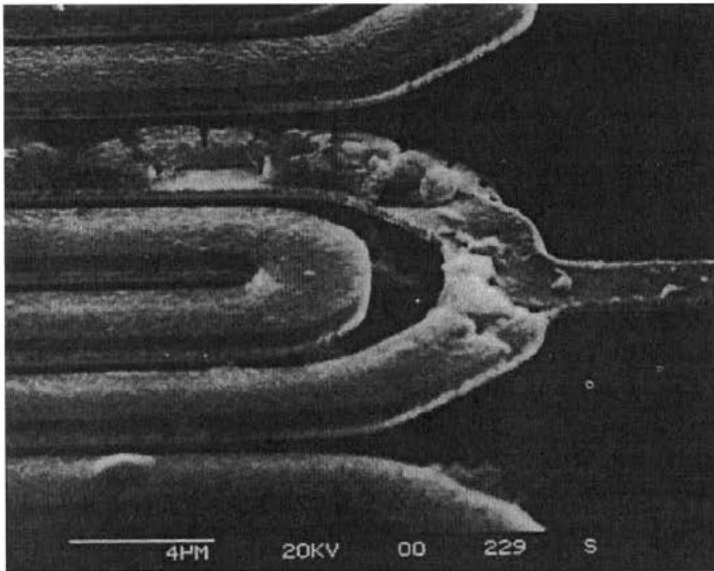


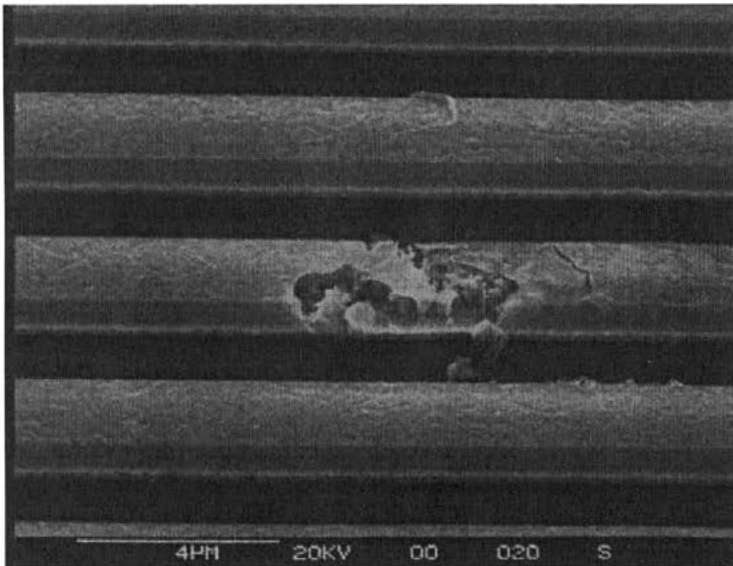
Fig. 7-6 Schematic model of reaction occurring at aluminum metal/oxide-water vapor interface. (a) Initiation of hydrogen attack, (b) Blister fracture and nucleation of aluminum-oxygen-hydrogen phases, (c) Hydroxide growth. From G. M. Scamans and A. S. Rehal, *J. Materials Science* **14**, 2459 (1979).

metallizations during wet etching and use of chlorinated solvents. Similarly, microcorrosion of Al-Cu bonding pads has been found to be a barrier to high-quality wire-bonding (16). Examples of accelerated corrosion in aluminum metallizations exposed to chlorides and humid atmospheres are shown in Fig. 7-7.

The specter of even more catastrophic corrosive attack exists in different regions of the chip because the Al metallization is additionally in contact with other metals and alloys. These serve as semiconductor contacts (e.g., PtSi),



(a)



(b)

Fig. 7-7 Corrosion of aluminum metallizations exposed to 1000 ppm solutions of NH_4Cl . (a) Metal corrosion extending from bonding pad into an array of interconnections. Once corrosion of the aluminum is complete, current flows in the more resistive Ti:W layer. (b) Localized corrosion where the native oxide has been penetrated. From J. D. Lawrence and J. W. McPherson, *29th Annual Proceedings of the IEEE Reliability Physics Symposium, IEEE*, p. 102 (1991).

diffusion barriers (Ti-W, Cr), bonding pads (Au wire), vias (W, TiN), and flip-chip solder pads (Au, Cu-Cr). To expose the extent of potential corrosion between assorted metal pairs, Griffen *et al.* (17) have recently experimentally determined the galvanic series for thin-film metallization and diffusion barriers displayed in Table 7-3. Corrosion tests revealed that the loss of Al from Al-0.5 wt.%Cu-1wt.%Si on CVD tungsten occurred some 25 times more rapidly than from Al-0.5 wt.%Cu-1wt.%Si on sputtered W-10 wt.%Ti, despite the ordering of metals in the table; passivation of W-Ti is the probable cause.

Exposure of these metal couples to an aqueous environment would unleash a galvanic corrosion nightmare, but for the most part metallizations are not susceptible to such attack because they are passivated with dielectric overcoats. Lastly, it should be appreciated that Al, as well as other metal films, is normally stressed in tension following oxide passivation. Since bulk Al alloys exhibit stress-corrosion cracking, the role of stress in accelerating corrosion failures of Al metallizations is always a possibility.

7.4.3.2 Gold Even the most noble of metals, gold, corrodes. For the most part, damage requires application of a dc bias voltage to the metal. In the presence of moisture and chloride ions, gold migration occurs by an anode dissolution-cathode plating process known as *metal migration*, which is described more fully in Sect. 7.5. The anode reaction is thought to be

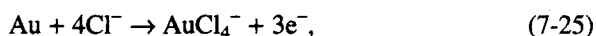


Table 7-3 Galvanic Series for Thin-Film Metallizations and Diffusion Barriers

Metal	EMF (mV)*
Cu (1μm)	+70
Al-0.5 wt.%Cu-1wt.%Si (0.75μm) on CVD W (0.55μm)	-6
CVD W (0.75μm)	-23
W-10 wt.%Ti (0.75μm)	-92
Al-2 wt.%Cu(1μm)	-368
Al-2 wt.%Cu-1wt.%Si (1μm)	-686
Al-1wt.%Si (0.75μm) on CVD W (0.55μm)	-783
Al (0.75μm) on CVD W (0.55μm)	-800
Al (0.75μm) on sputtered W-10 wt.%Ti (0.55μm)	-948
Al (1μm)	-1030

*Potentials relative to standard hydrogen electrode

Electrolyte: Stirred solution of 2000 ppm NH₄Cl at 25°.

Reference Electrode: Saturated Calomel (+0.242 mV).

From A. J. Griffen, S. E. Hernandez, F. R. Brotzen, J. D. Lawrence, J. W. McPherson, and C. F. Dunn, *31st Annual Proceedings of the IEEE Reliability Physics Symposium*, p. 327 (1993).

while the reverse reaction at the cathode results in dendritic Au deposits that often cause leakage currents or shorting of components. This type of electrolytic corrosion has also been observed in gold-containing metallizations, e.g., Ti-Pd-Au, Ni-Au-Cr, Ti-W-Au, and Ti-Pt-Au.

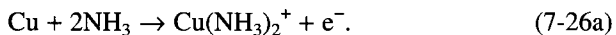
Polymer-encapsulated gold-wire bonds to aluminum pads have been reported to corrode when there is sufficient moisture permeation of the package (18). Not only is there the AuAl_2 (purple plague—see Sect. 5.3.4) compound to contend with, but the epoxy flame-retardants used contain halogens, which provide the Br^- necessary to autocatalyze local pitting corrosion.

Gold is commonly employed in GaAs MESFET contact and metallization schemes. Accelerated temperature-humidity-ionic contamination testing not surprisingly revealed that Al gate electrodes were more susceptible to corrosion damage than Au/refractory metal gates (19). Corrosion of Al, migration of Au between source and gate, shorting metallic bridges, and formation of Al-Au compounds correlated with degradation of device characteristics.

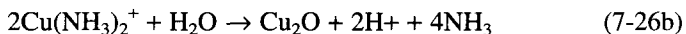
7.4.3.3 Other Metals A number of other metals are used in electronic applications, and a brief survey of corrosion damage in these follows.

a. Nickel Nickel-plated components and hardware are widely used in electronics. An example of nickel corrosion in solid-state microwave hybrid-blocking and bypass capacitors has been reported by Torkington and Lamoreaux (20). Failure in Ni, a normally corrosion-resistant metal, was attributed to the sealed-in moisture derived from the cover gas used during package solder sealing. With condensed moisture, the galvanic voltage (-1.75 V between Au and Ni) added to the larger applied voltage (5 V). In the presence of chlorine and CO_2 , NiCl_2 and NiCO_3 corrosion products were observed to short the capacitor electrodes. Interestingly, the 50:1 Au:Ni ratio is an example of the corrosion-prone, small anode-large cathode configuration (Sect. 7.3.4.2). In this case it greatly increased the current density and attack of nickel.

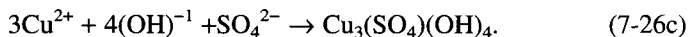
b. Copper For fear of contaminating and poisoning silicon, copper is sparingly used in microelectronics. Nevertheless, there has been recent research and development work on Cu-based, thin-film metallizations (21). Despite its relatively minor present role on the chip level, copper and its alloys play a very major role in electrical components such as wire, lead frames, printed-circuit board circuitry, heat sinks, vacuum-tube electrodes and RF gaskets, as well as in systems that cool electronics. Much of the corrosion damage in these items is attributable to attack by atmospheric pollutants, in particular $(\text{NH}_4)_2\text{SO}_4$ (1). In the presence of sufficient moisture, copper dissolves in ammonium sulfate solution to form $\text{Cu}(\text{NH}_3)_2^+$; i.e.,



The corrosion products Cu_2O and hydrated copper sulfates then form via the following typical reactions:



and



Copper is also susceptible to corrosion in humid sulfide atmospheres (22). Dendrites of corrosion products were observed to grow laterally from interconnections on printed-circuit boards in the absence of applied electric potentials. However, the processes of corrosion, dendrite growth, and filament shorting are accelerated by several orders of magnitude when electric fields are present.

7.4.3.4 Corrosion in Magnetic Disks Among the metal-containing electronic products sensitive to corrosion are computer hard disks (23, 24). Magnetic recording media typically consist of a layer of either Fe_2O_3 particles or a thin metal-alloy (Co–Ni) film sandwiched between an aluminum-substrate pad on the bottom, and metal oxide plus lubricant film on top. Common corrodents were found to be chlorine and sulfur. In the form of SO_4^{2-} , the latter attacked the Co, which was found on the disk surface.

7.4.4 Modeling Corrosion Damage

Over the years there have been numerous attempts to model corrosion damage with the objective of predicting failure times. This is an extraordinarily difficult task because:

1. the chemical reactions are often complex;
2. ion movements occur in an electrolyte fluid under the influence of spatially complex electric fields, concentration gradients, and convective flows; and
3. the metal surface structure, chemistry, and heterogeneous nature of damage are not well known, particularly under conditions of pitting or stress corrosion.

For these reasons it has been common to use familiar, phenomenological formulas like Eq. 7-22 to predict failure times. However, these have obvious shortcomings, as noted earlier. The approach taken by Pecht (24) and reproduced here is more satisfying, particularly because it connects uniform corrosion rates to reliability functions. To see how, we consider degradation of microelectronic packages that is caused by the two-step process of moisture permeation of packages followed by uniform corrosion of the metallization within.

An equation describing permeation of moisture was given earlier (Eq. 7-8):

$$\frac{P_i(t)}{P_e} = 1 - \exp[-\beta t], \quad (7-27)$$

where $\beta = Q_S/(P_o V) [M_{\text{air}}/M]^{1/2}$. It is further assumed that uniform corrosion is essentially described by a first-order chemical reaction of the form

$$\frac{dC}{dt} = -\alpha (C - C_o). \quad (7-28)$$

In this equation, C may be thought of as a corrosion density per unit volume or concentration, with C_o being the maximum, or self-limiting, value of C ; i.e., $0 \leq C \leq C_o$. Thus, the rate of corrosion is greatest initially when $C = 0$ and declines thereafter. Rate constant α combines a product of terms including the relative humidity (RH) in the package and the ubiquitous Boltzmann factor, $\exp[-E_c/kT]$, where E_c is the activation energy for corrosion and kT has the usual meaning. Explicitly, $\alpha = \alpha_o(\text{RH}) \exp[-E_c/kT]$, where factor α_o is dependent on the nature of the materials and voltage bias level.

The effect of temperature is often not what one might expect. When the internal package temperature is higher than the ambient temperature, moisture cannot condense. Thus devices that dissipate a lot of heat may actually suffer less corrosion damage than devices that consume and dissipate little power, e.g., CMOS devices. Moisture condenses more readily in the latter, and there is greater corrosion risk for metallizations, wirebonds, leadframes, etc.

Since a measure of RH is $P_i(t)/P_e$, for humidity conditions that allow for condensed moisture on metallic components, Eq. 7-28 yields

$$\frac{dC(t)}{dt} = -B(1 - \exp[-\beta t]) (C - C_o), \quad (7-29)$$

where constant $B = \alpha_o \exp[-E_c/kT]$. Upon separation of variables and integration,

$$\frac{C(t) - C_o}{C_i - C_o} = \exp \left[-\frac{B}{\beta} \{ \beta t - (1 - \exp[-\beta t]) \} \right]. \quad (7-30)$$

The constant of integration C_i satisfies the initial condition that at $t = 0$, $C = C_i$. If we now expand the term $\{ \beta t - (1 - \exp[-\beta t]) \}$ as $\beta t - 1 + 1 - \beta t + \frac{1}{2}(\beta t)^2 + \dots$, or $\frac{1}{2}(\beta t)^2$ for times of order t^2 , then

$$\frac{C(t) - C_o}{C_i - C_o} = \exp \left[-\frac{1}{2} B \beta t^2 \right]. \quad (7-31)$$

If it is assumed that $C_i = 0$, then $(C(t) - C_o)/(C_i - C_o)$ is a measure of the reliability R , so that

$$R(t) = \exp \left[-\frac{1}{2} B \beta t^2 \right]. \quad (7-32)$$

This association is consistent with full reliability initially (no corrosion) and total loss of reliability (fully corroded) at very long time. A test of experimental corrosion failure data plotted as R versus t is shown in Fig. 7-8 and compared with a fit of Eq. 7-32. Evidently, there is satisfactory agreement between the data and the predicted form of this equation.

From Eq. 4-4 we recall that $F(t) = 1 - R(t)$, and therefore,

$$F(t) = 1 - \exp \left[-\frac{1}{2} B \beta t^2 \right]. \quad (7-33)$$

This cumulative distribution function is of the Weibull form, and it suggests the use of such distributions for physically modeling corrosion failures. Finally, the hazard function, $\lambda(t)$, defined by $F'(t)/R(t)$, is simply calculated to be

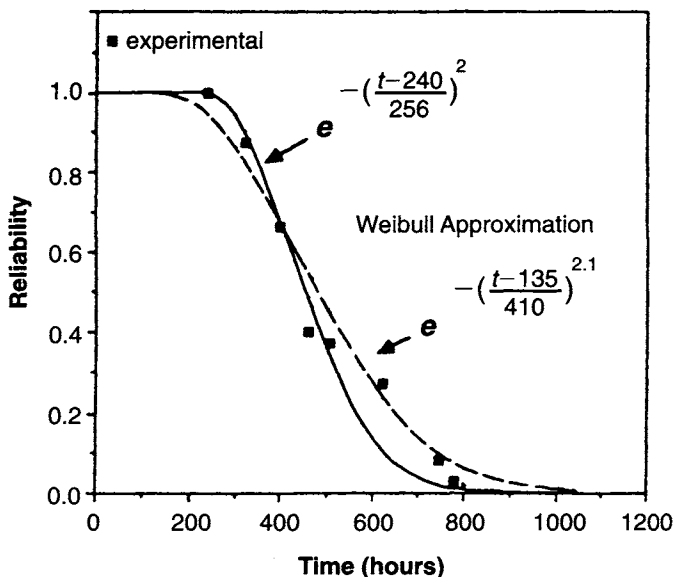


Fig. 7-8 Reliability of an encapsulated package in a corrosive environment. The experimental failure data are fit to a Weibull distribution. From reference 24.

$$\lambda(t) = \frac{B\beta t \exp\left[-\frac{1}{2}B\beta t^2\right]}{\exp\left[-\frac{1}{2}B\beta t^2\right]} = B\beta t. \quad (7-34)$$

This result corresponds directly to the Rayleigh failure rate (see Sect. 4.3.3).

7.5 METAL MIGRATION

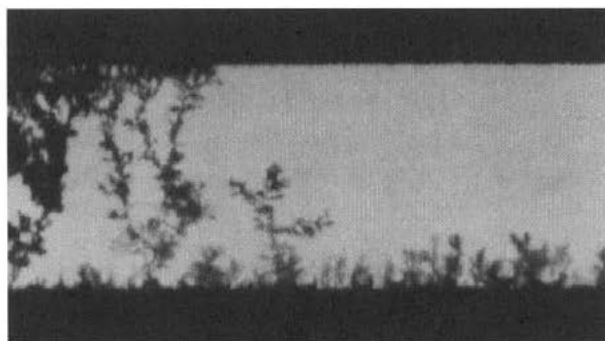
7.5.1 Introduction

The phenomenon of metal migration, also referred to as “metallic,” “wet,” and “electrolytic” electromigration, has caused many catastrophic microcircuit failures (22, 25). Unlike solid-state electromigration (Sect. 5.5), which generally requires elevated temperatures and high current densities ($\sim 10^5$ A/cm²), metal migration can occur at room temperature with passage of less than 1×10^{-3} A/cm². Like corrosion, it is an electrochemical phenomenon that can occur both within and outside of IC packages. The effect was first observed in polymer (phenol fiber) circuit boards containing silver lead wires and contacts. In essence, Ag metal was ionically removed from its initial location, then Ag ions migrated in the electric field, and finally Ag metal redeposited elsewhere (26). Water adsorbed on the insulator surface served as the electrolyte. At the anode, Ag formed a colloidal oxide, while at the cathode Ag metal assumed the dendritic morphology. Examples of dendritic growth in silver and copper conductors are shown in Figs. 7-9a and b.

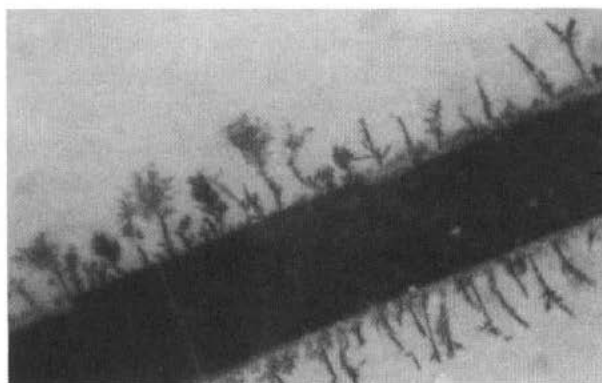
An extensive review of metal migration–accelerated testing (27) suggests that the effects are largest in silver. In addition to Ag, metals that migrate in a distilled water electrolyte include Bi, Cd, Cu, Pb, Sn, and Zn. When distilled water is contaminated with sodium or potassium chloride (0.001 to 0.01 M), Au, Pd, and Pt migrate. Other metals, e.g., Ni, Cr, and Ag–Au–Pd alloys, also migrate under certain conditions. When liquid water is present, electromigration effects increase greatly in speed and severity.

For metal migration to occur, the following conditions are generally required:

1. The anode metal must easily form ions.
2. Nearby there must be a metal that serves as the cathode.
3. A dc voltage must exist between the anode and cathode. Depending on the inter-electrode spacing, as little as 1 to 2 V up to 100 V or more may be required.
4. Moisture or a water film must be present on the insulator substrate surface to provide the electrolyte needed for ionic conduction. Porous and fibrous substrates, e.g., epoxy resin/glass fiber, polyesters, laminates, and fabrics, serve to retain the electrolyte.



(a)



(b)

Fig. 7-9 Examples of wet electromigration. (a) Silver dendrites growing in deionized water on glass-epoxy printed circuit board under applied voltage (24 $\times$). (b) Copper sulfide dendritic filaments growing from copper conductor between polyester sheets without application of voltage, but after exposure to a humid H_2S atmosphere (24 $\times$). S. J. Krumbein, *33rd Meeting of the IEEE Holm Conference on Electrical Contacts* (1987). Courtesy of S. J. Krumbein, AMP Incorporated.

Metal ions also migrate on oxide surfaces, and these effects are discussed in Sections 7.5.3 and 7.5.4.

7.5.2 Metal Migration on Porous Substrates

The need to accumulate moisture within the porosity of the insulator was emphasized above in order for metal migration to become a reliability problem. In this regard the model developed by DiGiacomo (28) is instructive. The analysis

assumes that the metal ion current density that flows is directly proportional to the fraction of substrate pores containing condensed water. For a lognormal distribution of pore radii, the fractional area (f) on which condensation occurs is given by

$$f = \frac{1}{2} \operatorname{Erfc} \left[\frac{1}{2} \sigma \ln \left(\frac{r_{\text{avg}}}{r} \right)^2 \right], \quad (7-35)$$

where σ (see Sect. 4.3.5) is $\ln(r_{50}/r_{16})$, and r_{avg} is the average pore radius. Water condensation in pores from a saturated vapor pressure (p) is basically driven by a chemical reaction whose free energy change per unit volume ΔG_V is given by $-kT \ln \frac{p}{p_{\infty}}$ (29). The ratio of the vapor pressure above the pore radius, or meniscus, r , relative to that over a flat pore surface p_{∞} , may be thought of as the relative humidity (i.e., RH). In this context, elementary nucleation or capillarity theory establishes the connection between r and ΔG_V as $r = \frac{2\gamma\Omega}{kT \ln(\text{RH})}$, where γ and Ω are the water surface tension and atomic volume, respectively, and kT has the usual meaning.

The ionic mass flux (J_i) is the product of the ion concentration (C_i in g/cm^3) and its velocity, where the latter is given by the Nernst–Einstein equation (Eq. 1-6); i.e., $v_i = DF/RT$. As in electromigration (Sect. 5.5.2), the force F is equal to $Zq\mathcal{E}$, where Z is the ion valence, q is the electronic charge, and $\mathcal{E}$ is the electric field. Combining all of the above terms, the equation

$$J_i = \left(\frac{ZqC_i D \mathcal{E}}{RT} \right) f = ZqC_i D \mathcal{E} \operatorname{Erfc} \left[\frac{\sigma}{2} \ln \left(\frac{r_{\text{avg}} kT \ln(\text{RH})}{2\gamma\Omega} \right)^2 \right] \quad (7-36)$$

results. The goal is to obtain the time to failure (MTTF), which is defined by how long it takes a dendritic filament to span conductors separated by a length L when a voltage V is applied. To produce and extend a dendrite, the current density at its tip must be orders of magnitude higher than its average value. Dendritic growth occurs by a focusing of the ion current at its tip of radius r_d , an effect that magnifies the local field to $\mathcal{E} = V/r_d$. The critical mass of metal that is transported to create the shorting dendrite filament is assumed to be $Lr_d \delta d$, where δ is the electrolyte thickness and d is the metal density. Failure occurs in time $t = \text{MTTF}$, and by definition the corresponding mass flux (mass/area-time) is $J_i = Lr_d \delta d/r_d \delta \text{MTTF} = L d/\text{MTTF}$. Solving for MTTF yields the desired result

$$\text{MTTF} = \frac{r_d RT L d}{ZqC_i DV \operatorname{Erfc} \left[\frac{\sigma}{2} \ln \left(\frac{r_{\text{avg}} kT \ln \text{RH}}{2\gamma\Omega} \right)^2 \right]}. \quad (7-37)$$

In this closed-form equation, the effects of voltage, humidity, nature of the substrate, dendrite geometry, and temperature (through D and RH) are all linked.

7.5.3 Metal Migration on Ceramic Substrates

In this section a simple model (30) for metal migration based on Fig. 7-10a will be developed. A parallel pair of metal stripes with indicated dimensions has a potential V applied across it. Due to moisture condensation, a thin electrolyte layer coats the ceramic substrate and spans the metal electrodes. The electrolyte contains ionic contaminants that make it sufficiently conductive to cause metal migration. Corrosion will proceed until a length of electrode, approximately equal to its width, is corroded to an open condition (Fig. 7-10b). Dendrite

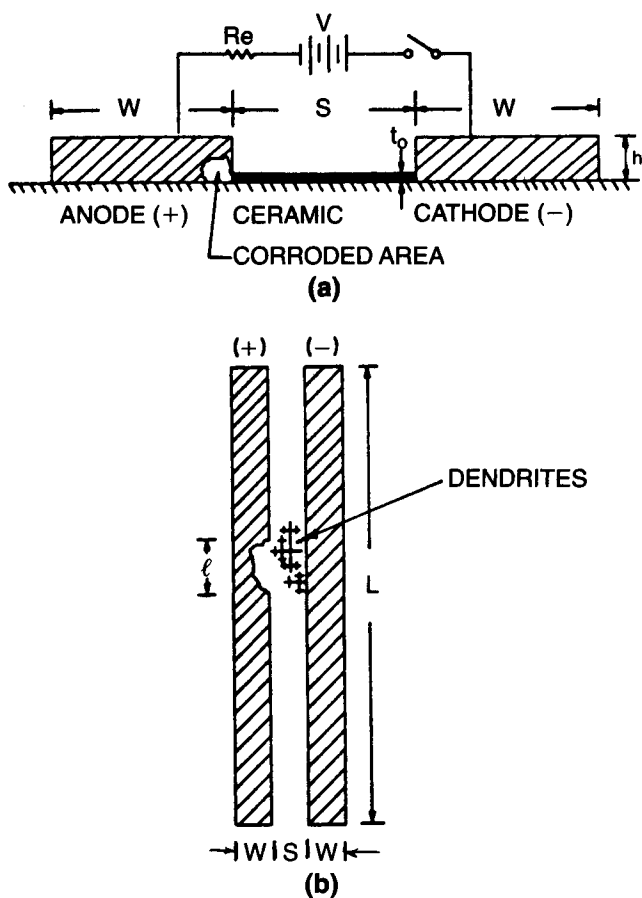


Fig. 7-10 (a) Cross-section of corrosion process involving metal migration between conducting stripes on a ceramic substrate covered with a water film. (b) Plan view of corrosion process. From reference 30.

growth, which can short conductors, is often a secondary effect of corrosion. Using Ohm's law ($i = V/R$) and the definition of resistance $R(\Omega) = \rho S/(L t_o)$, we obtain

$$i = \frac{VL}{(Sp/t_o)}, \quad (7-38)$$

where ρ ($\Omega\text{-cm}$) is the resistivity of the electrolyte. The volume of corroded metal, v (cm^3), is proportional to the total charge Q (C) passed, or $i \times \tau$, where τ is the time. Faraday's law essentially states that $Q = n d v F_a/M$, where F_a is the Faraday constant and n , d , and M are the metal valence, density, and atomic weight, respectively. Therefore, combining terms, the effective time to failure ($\tau = \text{MTTF}$) when $v = lwh$, is given by

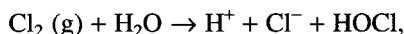
$$\text{MTTF} = \frac{n d v F_a S}{MVL} \left(\frac{\rho}{t_o} \right) = \text{constant} \times \rho/t_o, \quad (7-39)$$

where the constant depends on the metal in question and the amount lost, the corrosion geometry, and the applied voltage.

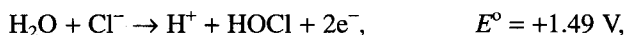
This model predicts that corrosion failure times are directly proportional to the resistivity of the electrolyte divided by its thickness. The quantity ρ/t_o is known as the sheet resistance and has the peculiar units of ohms per square. This definition stems from the fact that the electrical resistance of a thin square of material is independent of the size of the square. In thin films or layers, the sheet resistance is measured by probing the wide surface area with a four-point resistivity probe. In aqueous solutions the equivalent conductances (Λ) are given by $\Lambda = 1000/(\rho C)$, and are tabulated for concentrations (C) of different ionic species (31). This makes it possible to estimate the sheet resistance.

Example 7-1 After introduction of 1.7 ppm of Cl_2 into an $85^\circ\text{C}/80\%$ RH ambient, the corrosion leakage current across metal stripes on a ceramic substrate increased. If the moisture layer is 100 nm thick, what is the electrolyte sheet resistance?

Answer Chlorine dissolves in water through the reaction



which is the difference of the half reactions



and



Applying Eq. 7-14,

$$0 = (1.49 - 1.36) + 8.314 (358) / 2 (96,500) \ln (H^+) (Cl^-) / (Cl_2).$$

Substituting for the chlorine concentration, $(Cl_2) = 1.7 \times 10^{-6}$ yields $(H^+) (Cl^-) = 3.36 \times 10^{-10}$. Thus there are 1.83×10^{-5} equivalents per liter for each ion. The equivalent conductance of H^+ and Cl^- is 775 at $85^\circ C$ (31). Since $\rho = 1000/(CA)$, the sheet resistance, $\rho/t_o = 1000/(CA t_o)$. Substitution yields $\rho/t_o = 1000/1.83 \times 10^{-5} (775)(10^{-5}) = 7.1 \times 10^9 \Omega/sq$.

7.5.4 Mobile Ion Contamination in CMOS Circuits

Related to the problem of ion migration in liquid electrolytes is ion transport through a solid ionic medium like SiO_2 . Though not a corrosion problem, the fact that ionic motion is involved makes it of interest here. One of the most serious reliability problems exhibited by early MOS transistors was instability of the threshold voltage (V_T) causing operation at more negative voltages than desired. With positive bias, V_T drifted with time to more negative values. At elevated temperatures device behavior was very erratic. The cause of these problems was drift of positively charged mobile ions in the gate oxide. A particular culprit was identified to be sodium, hence the appellation "the sodium problem." Sodium entered processing lines from the quartzware used as well as from the salt introduced by winter footwear. Despite the progress made in tracking down sources of impurities and reducing instabilities due to positive mobile ion contamination, failures due to this cause still occur (32).

Ion drift velocities (v) can be calculated from the measured mobilities (μ in units of $cm^2/V\cdot s$) through the formula

$$v = \mu \mathcal{E}, \quad (7-40)$$

where $\mathcal{E}$ is the electric field. For Na^+ and K^+ in SiO_2 the respective values of mobility are

$$\mu (Na^+) = 1.0 \exp \left(- \frac{0.66 \text{ eV}}{kT} \right) cm^2/V\cdot s \quad (7-41a)$$

and

$$\mu (K^+) = 0.03 \exp \left(- \frac{1.09 \text{ eV}}{kT} \right) cm^2/V\cdot s. \quad (7-41b)$$

This means that Na^+ ions will traverse a 15 nm thick oxide in 10^{-5} s at $200^\circ C$ in fields of 1 MV/cm. On the other hand, the larger K^+ ions require about 10 s. Maximum tolerable Na^+ and K^+ concentrations are $1 \times 10^{10} cm^{-2}$. When positive

ions drift through the oxide of an MOS structure they also shift the effective flat-band voltage. Thus a positive gate bias voltage applied to NMOS transistors drives ions toward the Si interface and displaces the flat-band voltage by $-Q/C_o$ (Sect. 2.5.3), where Q is the ion charge and C_o is the oxide capacitance. Equivalently, transistor threshold-voltage shifts occur as a result of charge transport (Eq. 2-6). Such instabilities are not as evident in PMOS transistors, which is the reason these devices were the first to be commercially developed.

7.6 RADIATION DAMAGE TO ELECTRONIC MATERIALS AND DEVICES

7.6.1 A Historical Footnote

It was a “big bang” on July 9, 1962, that catapulted the reliability of radiation-damaged solid-state electronics to the center stage of world politics (33). On that day the U.S. high-altitude nuclear device Starfish was detonated. This was followed by several similar Russian tests later that year. The resulting nuclear contamination of the exoatmosphere generated sufficient electronic perturbation to the Van Allen belt to cause failure of the Telstar 1 communications satellite. Not long after this, Nikita Khrushchev, premier of the USSR, announced that this radiation vulnerability would be considered a possible strategy to destroy future U.S. military space systems. This threat provided an incentive and challenge to develop survivable space electronics and acquire an understanding of the nature of radiation damage to semiconductor devices. Fortunately, the damage to Telstar was reversible. In time, enough was learned so that damaged p-n junctions could be repaired from the ground by altering their bias voltages.

Even though the cold war is now history, interest persists in the subject of radiation damage to semiconductor devices and electronic equipment used in the peaceful exploitation of nuclear energy. A number of recent books on this and related topics (34, 35, 36) amply demonstrate this interest and importance.

7.6.2 Some Definitions

The radiation we are primarily concerned with is composed of energetic particles and photons possessing energies ranging from tens to millions of electron volts. While the former include neutrons, electrons, and alpha particles, X-rays, gamma rays, and ultraviolet light are examples of photons. The radiation energy density that materials are exposed to is known as the dose (D). If dE is the mean energy absorbed per unit mass dm of irradiated material, the dose is defined by

$$D = \frac{dE}{dm}. \quad (7-42)$$

Although the absorbed dose in a small region, i.e., a gate oxide, can be rigorously defined, it is impossible to know the actual energy imparted. The common unit of dose is the rad, defined as 1 rad = 100 ergs/g. In the SI system the gray (Gy) is the unit of dose, where 1 Gy = 1 J/kg, so that 1 Gy = 100 rads. Since the absorbed dose depends on the material in question, the latter should be referenced directly following the dose unit, e.g., rads (Si), rads (SiO₂), Gy (GaAs).

In many applications the *dose rate* dD/dt , expressed in rad/s, is important. For example, during exposure to long-lived radioactive isotopes, D is assumed to be constant, and the dose rate is zero. If radiation is pulsed, e.g., flash X-ray machines or a nuclear blast, then the dose rate also becomes a significant variable. Examples will be given of cases where D and dD/dt are reported.

7.6.3 Radiation Environments

The primary radiation environments that degrade or damage electronic materials and devices include outer space, nuclear reactors, radiation processing activities, nuclear weapon and controlled fusion facilities (35). Electronic equipment is exposed to these radiation environments during interplanetary space flight, in power generating plants, in factories, and in defense applications.

In outer space particle and photon radiation have typical energies ranging from keV to GeV and consist of

- a. a broad spectrum of energetic particles trapped in the earth's magnetic field, forming radiation belts;
- b. galactic cosmic rays, which are the fluxes emanating from energetic heavy ions. Energies can extend even beyond TeV;
- c. solar winds and flares, which are eruptions of energetic protons and smaller amounts of alpha particles, heavy ions, and electrons with energies ranging to hundreds of MeV.

The nuclear-reactor environment depends on reactor type. Within the core of pressurized-water, boiling-water, and fast-breeder reactors, neutron fluxes range from 10^{12} to 10^{14} cm⁻² s⁻¹, while gamma-radiation levels between 10^5 to 10^{10} rad/h are common. This contrasts with neutron fluxes of 3×10^5 cm⁻² s⁻¹ and gamma ray doses of 10^{-3} to 10^2 rad/h in the containment area where electronic equipment and components operate.

Radiation processing is employed in a number of industrial activities, which are noted together with the doses involved; retardation of food spoilage (30–500 krad), sterilization of medical supplies (2.5 Mrad), sewage processing (~1

Mrad), and modification of polymers (10–15 Mrad). The processing of spent nuclear fuels generates maximum dose rates of 10^5 rad (Si)/h for gamma radiation and 0.1 rad (Si)/h for neutrons.

In contrast, a 1 megaton nuclear weapon produces prompt radiation dosages of 2100 rad neutrons and 62,000 rad gamma at a distance of 1 mile from the detonation point. If and when controlled nuclear fusion reactors are a reality, radiation levels of 10^{12} – 10^{16} equivalent neutrons-cm⁻² and ionizing doses of 10^3 – 10^7 rad (Si) s⁻¹ will have to be dealt with. The primary source of this radiation is 14 MeV neutrons from the deuterium–tritium reaction.

In each of the above environments electronic equipment is used to sense and monitor radiation as well as to control robots and other machinery remotely. To appreciate the damage to solid-state and electro-optical devices that the above radiation magnitudes can wreak, we must first understand how energetic particles and ionizing radiation interact with materials.

7.6.4 Interaction of Radiation with Matter

We are primarily concerned with the consequences of atomic displacements and ionization in materials caused by radiation. This broad subject, which has been researched for a half century, does not easily lend itself to a brief review. Nevertheless, in flow-chart style, Fig. 7-11 neatly summarizes radiation damage into particle- and photon-induced categories. Of particular interest is the long-lived, as well as transient, damage produced by the particular radiation.

a. Atomic Displacement Starting with particles, we note that physical displacement of atoms from their original positions is the major form of damage created. The particle trajectory, and the damage produced, depends on its energy, mass, and charge, as well as on the nature of the matrix. What happens during ion implantation doping of silicon illustrates these effects. Dopant ions with kinetic energies of ~50–100 keV, or thousands of times larger than the binding energies of lattice atoms, knock atoms off lattice sites in collision cascades, creating vacancy and interstitial defects in the process. It is usually difficult to estimate just how many vacancy–interstitial pairs are created. For example, if the displacement energy is 10 eV, a 12 MeV proton is predicted to generate only 17 vacancy–interstitial pairs (35). Each incoming ion executes a different trajectory and comes to rest at a different depth from the surface. The ion range is statistically distributed, and lighter particles penetrate further than heavier ones of the same energy. Heating the radiation-damaged matrix heals it because atoms again resume their equilibrium positions by diffusional processes.

Materials exposed to an energetic-particle environment may also become radioactive. Protons, neutrons, and nuclei are all capable of transforming stable

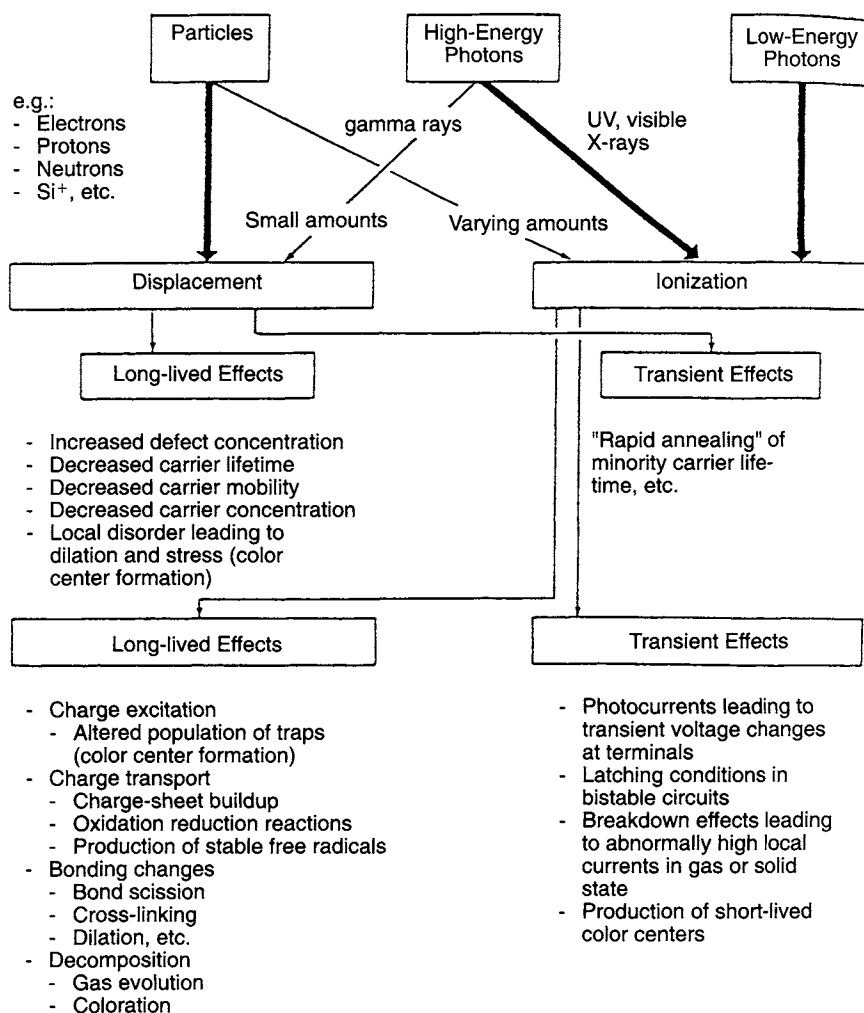


Fig. 7-11 Summary of radiation-induced degradation effects in solid-state materials and devices. From reference 35.

nuclei into radioactive ones by removal of nucleons, or in the case of low-energy neutrons by neutron capture. Depending on the particular atom, such induced radioactivity decays with a characteristic half-life through emission of gamma rays, beta rays (electrons), and alpha particles.

It should be realized that different classes of solids suffer different levels of particle-induced damage. Most vulnerable are semiconductors and insulators

because significant changes in electrical properties occur. Metals, on the other hand, suffer some mechanical hardening and embrittlement but no appreciable change in electrical behavior. Furthermore, damage to metals from gamma and X-rays is negligible, unlike their effect in nonmetals.

b. Electronic Charge Displacement (Ionization) The interactions between energetic radiation and the electronic structure of atoms is every bit as varied as the particle momentum transfer to atoms described above. Radiation energies are usually large enough to cause complex core-electron transitions in individual atoms. These involve a train of events that begins when electrons are ejected from low orbital levels by incoming energetic photons (X-rays, gamma rays) or particles (electrons, ions). Electrons from outer levels fill these now vacant levels, and X-ray emission, characteristic of the excited atom, results. More complex internal electron excitations and transitions ensue in the ejection of Auger electrons. Such processes cause no damage; they are capitalized upon to fingerprint atoms. Applications of X-ray fluorescence and Auger electron spectroscopy in failure mode analysis will be discussed in Chapter 11.

In addition to localized atomic excitation, energy loss in semiconductors and inorganic insulators is ultimately converted into electron-hole pair production. Both energetic particles and photons are capable of causing such ionization damage. The excitation of valence-band electrons to the conduction band creates charge carriers that are mobile in electric fields. Simultaneously, slower-moving holes remain behind in the valence band. As we have already seen, trapped holes in thin dielectrics enhance the probability of degrading dielectric behavior.

Depending on whether particles or photons of the same energy are involved, there is a difference with respect to the amount of ionization produced. Because some of the particle energy is partitioned to momentum transfer to atoms, less is available for ionization than is the case with photons. Thus ionization effects produced by megavolt particles are equivalent to those induced by much less energetic photons, electrons, or UV radiation. Unlike alpha particles and protons, which can be effectively shielded by thin metal foils, cosmic, gamma, and X-rays easily penetrate housings of electronic equipment.

7.6.5 Device Degradation Due to Ionizing Radiation

The subject of primary and secondary damage mechanisms in various devices and components has been reviewed by several authors (34–39) and summarized in Table 7-4. In this section a number of these mechanisms are described; the important subject of single-event upset, however, merits a discussion all its own and is treated in Sect. 7.6.6.

Table 7-4 Primary and Secondary Damage Mechanisms in Various Devices, Circuits, and Components in the Space Radiation Environment

Devices, Circuits, and Components	Permanent					Temporary		
	Lifetime Degradation	Carrier Removal	Trapping	Mobility Degra- dation	Charge Buildup	Latchup	Absorption	Single Event Upset
Si bipolar transistors and integrated circuits	P	S			S	P		P
Si MOS transistors and integrated circuits				S	P	P		P
JFETs		P	P	S	S			
p-n junction diodes	P	P	P		S			
LEDs, laser diodes	P							
Charge-coupled devices	P		P		P			
Photodetectors	P				P			
Microwave devices and circuits	P	P		S				
GaAs transistors and integrated circuits		P		S	S			
Capacitors					P			P
Diffused resistors		P		S				
Optical components (fibers, windows, mirrors)								P

P = Primary; S = Secondary.
Charge buildup can be a primary failure mechanism in some types of bipolar integrated circuits
Adapted from J. R. Srour and J. M. McGarrity, *Proc. IEEE* **76**, 1443 (1988).

7.6.5.1 MOS Devices The basic effect of ionizing radiation in SiO₂ films can be understood in terms of dielectric charging effects previously introduced in Chapter 6. Damage begins in the gate oxide in the form of electron-hole pairs. Because it requires 17 eV to produce electron-hole pairs in SiO₂, such charge can be generated with radiation of modest energy. In comparison, carrier pair production is energetically easier in semiconductors, requiring only 3.6 eV in Si and 4.8 eV in GaAs. For an exposure of a rad, approximately 10¹³ pairs/cm² are generated in SiO₂. While the more mobile electrons are removed, some holes tend to be trapped at the Si/SiO₂ interface. This additional radiation-induced oxide charge (Q_r) reduces the threshold voltage (V_T) by an amount Q_r/C_o , where C_o is the oxide capacitance (see Sect. 2.5.3). Thus the MOSFET may be turned on even without any applied gate voltage.

Threshold-voltage shifts are both dose and dose rate dependent. This is illustrated in Fig. 7-12 for n -channel MOS devices. At first, V_T is unchanged up to doses of approximately 10^5 rad (SiO_2), then it goes negative, where it reaches a minimum and finally swings up toward more positive values as the dose rises. This behavior, which is accentuated by increasing dose rates, is a superposition of two effects. Gate oxide-trapped holes cause a monotonic decrease in V_T with dose, while negative interface states induce a rise in V_T . Other deleterious effects of radiation in MOSFETs include channel-mobility degradation and increased leakage currents at high dose levels.

7.6.5.2 GaAs Devices In some environments GaAs devices are far more radiation tolerant than Si devices, but in other radiation environments the damage is comparable. For example, GaAs FETs have no gate insulators to experience charge buildup and, therefore, are relatively immune to radiation. In bulk GaAs, like Si, ionizing radiation generates electron-hole pairs, introduces defect levels and scattering centers in the energy gap, removes carriers, and reduces mobility. The latter two effects are responsible for transconductance or gain degradation in transistors. For this to occur in GaAs, dose levels of $\sim 10^8$ rad are typically required when channel doping levels are relatively high.

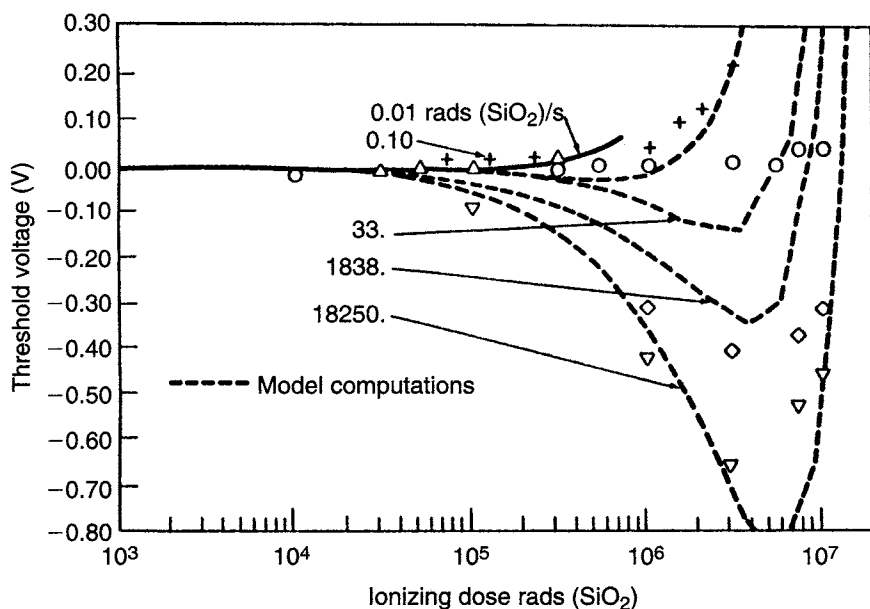


Fig. 7-12 Threshold voltage for n -channel CMOS transistors as a function of dose for various dose rates. From D. B. Brown, W. C. Jenkins, and A. H. Johnston, *IEEE Trans. Nuclear Science* **36**, 1954 (1989).

7.6.5.3 Electro-optical Devices Laser diodes, light emitting diodes, and photodetectors all suffer degradation of operating characteristics due to radiation damage (37). In lasers, permanent increases in threshold currents result from ionizing radiation, which generates nonradiative recombination centers. The latter are also primarily responsible for degradation of LEDs. Silicon-based photodetectors and CCDs are prone to the same degradation mechanisms, i.e., charge buildup, surface leakage currents, reduction in carrier mobility and lifetime, that plague MOS devices.

Optical fiber, often used with these devices, offers advantages relative to conventional transmission lines because it is insensitive to electromagnetic noise. However, optical fiber is more sensitive to radiation damage in the form of luminescence and increased absorption. Spurious light signals and attenuation of light pulses can then be expected.

In general, the damage and degradation discussed to this point occurs only at high radiation levels that are created artificially. We now consider the often more serious radiation damage that can occur in circuits powered in what would be considered natural, everyday environments.

7.6.6 Soft Errors

7.6.6.1 Introduction Computer circuits function by identifying small packets of charge as elemental bits of information. Any noise that modifies this charge also may change the information stored. This can happen, for example, when radiation, a source of this noise, impinges upon and alters the electrical state of an internal node, causing a false logic transition, e.g., a bit flop. Such errors may mean data corruption, execution of wrong commands, or even a “locking up” that requires reinitialization. In digital electronics, errors that are not caused by permanent damage to circuits are referred to as soft errors, soft fails, or single event upsets (SEU). Soft fails, happily, do not necessarily affect the computer user, because the system may be turned off or the incorrect memory bits may be overwritten before they are used. In microelectronic circuits employed in satellites, digital avionics, high-flying aircraft, as well as in terrestrial electronics and computers, soft errors are a major concern. Over the years from 1970 to 1982 there have been over 40 recorded upsets in space satellites (e.g., Voyager, Pioneer Venus, Galileo) that affected NMOS, CMOS, and bipolar memories, as well as assorted flip-flop circuits (40). Occasionally, however, the more serious effect of single event latchup (SEL) has been triggered, resulting in circuit destruction.

Soft errors and their reliability implications are sufficiently important to be the subject of an entire recent issue of the *IBM Journal of Research and Development* (41). In it there is an interesting history of the IBM experience from 1978

to 1994 with terrestrial cosmic rays and soft errors, including experimental measurements to detect them, results of accelerated testing, and models used to analyze the response of integrated circuits. Much of what follows is derived from this excellent survey of the subject.

7.6.6.2 Sources of Radiation That Cause Soft Errors In 1978 the first evidence of sea-level soft fails from energetic particle impact was published by May and Woods (42). Radioactive alpha particles emanating from ceramic packages were determined to be the cause of damage. The source was the water used in the factory, built downstream from the tailings of an old uranium mine that contained high levels of radioactive elements. A year later Ziegler and Lanford (43) suggested that sea-level cosmic rays could also cause upsets in electronics. Thus the nuclear particles that create electronic noise and soft fails stem from two sources, namely, the decay of radioactive atoms and extraterrestrial cosmic ray particles.

a. Nuclear particles from radioactive atoms A number of materials employed in integrated-circuit manufacture contain trace levels of radioactive atoms that emit either gamma rays (high-energy photons), beta rays (electrons) and alpha particles (helium ions), or some combination of these. From the standpoint of producing soft errors, alpha particles are the most “upsetting.” These alpha particles, with energies ranging from 3.8 to 8.8 MeV, stem from the decay of naturally occurring radioactive heavy actinides like uranium (U^{238}) and thorium (Th^{232}) and their daughter nuclei. Because these elements are part of the earth’s crust, trace amounts are present in metallization alloys, gold, alumina, and lead frame alloys used in packaging; they can also be introduced during processing, as we shall see.

Alpha particle (α P) activity is often quoted in terms of a flux, or rate of alpha emission from the surface per unit area (cm^2) per unit time (h). As a rule of thumb, an alpha flux of 1 alpha/ cm^2 -h corresponds to one part per million (ppm) of U^{238} nuclei. The following calculation shows us how to estimate the flux more exactly (44). First we note that U^{238} has an activity of 6.85×10^{-7} Ci/g, where 1 curie (Ci) = 2.2×10^{12} disintegrations/min (dpm). Furthermore, the decay of U^{238} to Pb^{206} releases 8 alpha particles.

Example 7-2 Estimate the alpha particle flux from an Al_2O_3 package containing 1 ppm U^{238} per gram of ceramic. Assume that only 25% of the alphas generated escape the package from a depth of 14 μm or less, and that the density of Al_2O_3 is 3.9 g/ cm^3 .

Answer In 1 cm^3 of Al_2O_3 there are $1 \times 10^{-6} \times 1 \text{ cm}^3 \times 3.9 \text{ g/cm}^3 = 3.9 \times 10^{-6}$ g of U^{238} . This corresponds to $6.85 \times 10^{-7} \text{ Ci/g} \times 3.9 \times 10^{-6} \text{ g/cm}^3 = 2.67 \times 10^{-12}$ Ci/ cm^3 , or an equivalent radiation flux of $2.67 \times 10^{-12} \text{ Ci/cm}^3 \times 14 \times 10^{-4} \text{ cm} =$

$3.74 \times 10^{-15} \text{ Ci/cm}^2$. Collecting terms, $8 \alpha\text{P/dis} \times 0.25 \times (3.74 \times 10^{-15} \text{ Ci/cm}^2) \times (2.2 \times 10^{12} \text{ dpm/Ci}) \times 60 \text{ min/h} = 0.98 \alpha\text{P/cm}^2\text{-h}$.

Typical αP flux levels in IC chips without tungsten silicides rarely exceed $0.005 \alpha\text{P/cm}^2\text{-h}$. Ceramic packaging components (lids, leadframes) may exhibit flux levels ten to as much as a thousand times larger. In both cases there is considerable scatter, depending on processing conditions and involved vendors. Nevertheless, such αP fluxes are sufficient to cause troublesome SEUs. Details of the αP damage mechanism and energy required in silicon will be treated in Section 7.6.6.4.

An interesting example of chip contamination by such radioactive elements involves Po^{210} , an isotope of polonium that occurs naturally as a daughter of radon gas. Hera RAM memory chips manufactured by IBM in 1987 were found to be contaminated by Po^{210} , an alpha emitter, such that hot chips suffered failure rates a thousand times higher than clean chips. Following a frantic search for the source of contamination, the culprit was found to be radioactive nitric-acid bottles. In producing these, the manufacturer had to meet stringent semiconductor standards of cleanliness. Therefore, a bottle-cleaning machine employing radioactive Po^{210} was used (as in smoke alarms) to ionize an air jet that dislodged electrostatic dust left behind after washing. The jets leaked radioactivity only intermittently, confounding the problem. This example is illustrative of the painstaking efforts that periodically surface in the industry to track down and eliminate sources of SEUs.

b. Nuclear particles from cosmic rays The second, and in many ways more troubling, source of nuclear particles stems from cosmic rays that bombard Earth from the recesses of outer space. Primary sources of cosmic rays are energetic galactic particles and solar-wind particles. The former are far more energetic than the latter, which typically have energies of $\sim 1 \text{ GeV}$. Secondary cosmic rays are generated when the primary cosmic rays hit atmospheric atoms and create a shower, or cascade, of secondary particles. Finally, there are the terrestrial cosmic rays that hit Earth. These are composed of assorted nuclear particles, e.g., neutrons, pions, protons, and muons, that have energies ranging from 1 MeV to 1 GeV . About 97% of the nucleon flux at sea level is due to neutrons, and fluxes are typically $10^5/\text{cm}^2\text{-yr}$. It is interesting to note that cosmic-ray fluxes vary strongly with solar cycle activity and altitude (45). In going from New York (0 ft) to Denver (5280 ft) the neutron flux increases 3.4 fold, and at Leadville, Colorado (10,200 ft), the flux is 12.8 times higher than at sea level. This trend continues and accounts for the fact that the soft fail rate of electronics at airplane altitudes is 100 times the magnitude recorded at terrestrial altitudes.

A relative comparison of the radioactive and cosmic-ray contributions to soft error rates (SER) depends on various factors, including levels of nuclear conta-

mination, location, and altitude of devices. Data for Po^{210} contaminated 288 Kb DRAM chips in Fig. 7-13 depicts measured soft error rates as a function of time and altitude. The results can be described by the equation

$$\text{SER} = A \exp\left(-\frac{t}{\tau_{1/2}}\right) + B + C, \tag{7-43}$$

where A is the initial SER due to Po^{210} contamination, $\tau_{1/2} = 138$ d half-life, B is the contribution due to longer-lived alpha emitters, and C is the cosmic ray contribution.

The roles of these radiation sources may also be coupled. For example, cosmic-ray activation of boron in borophosphosilicate glass has recently been suggested as a dominant source of alpha particles that cause soft errors in Si (46).

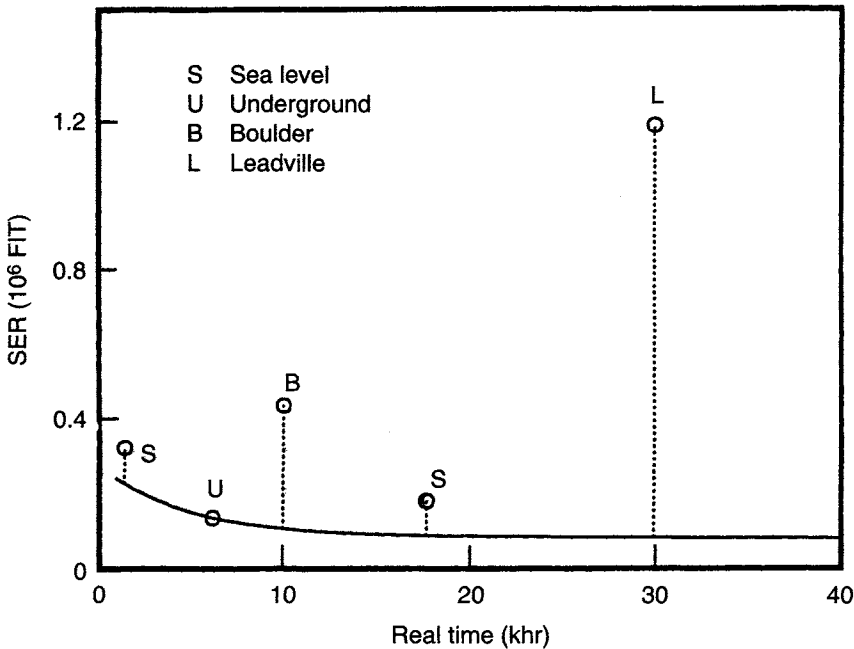


Fig. 7-13 Soft error rates for 288 Kb DRAMs contaminated with Po^{210} measured at various altitudes. The solid line is the alpha-particle component of SER, while the vertical distance above (dotted lines) corresponds to the cosmic ray contribution to the SER. From T. J. O’Gorman, J. M. Ross, A. H. Taber, J. F. Ziegler, H. P. Muhlfeld, C. J. Montrose, H. W. Curtis, and J. L. Walsh, *IBM J. Res. Develop.* **40** (1), 41 (1996).

7.6.6.3 Sample Calculations of Neutron–Silicon Interactions Since energetic neutrons are the primary particles involved in cosmic ray SEUs, it is instructive to reproduce Ziegler’s (45) thumbnail calculations of their interactions with Si circuits. Silicon contains 28 nucleons (protons and neutrons), each of which has a strong interaction radius of 1.3×10^{-13} cm; therefore, for the Si nucleus the radius scales to about 4×10^{-13} cm, corresponding to a cross-sectional area of 5×10^{-25} cm². This circular area for Si is the cross-section for interaction with neutrons. If the nuclei are uniformly distributed, they present a black wall, or total absorption cross-section, to a *neutron flux* given by

$$1/(\text{area/atom}) = 1/(5 \times 10^{-25} \text{ cm}^2) = 2 \times 10^{24} \text{ atoms/cm}^2. \quad (7-44)$$

For Si with an atomic density of 5×10^{22} atoms/cm³, the absorption length is $(2 \times 10^{24} \text{ atoms/cm}^2)/(5 \times 10^{22} \text{ atoms/cm}^3) = 40$ cm. Assuming that Si–neutron interactions must occur within 10 μm of the surface to be important, the probability of a nuclear “hit” is

$$(\text{target depth})/(\text{absorption length}) = 10^{-3} \text{ cm}/40 \text{ cm} = 0.25 \times 10^{-4}. \quad (7-45)$$

Thus one out of 40,000 incident neutrons will interact within 10 μm of the surface.

The key question we now address is how many of these hits will cause a soft error. Assuming bipolar transistor ICs whose typical active device area is 0.04 cm², the number of *active atoms/chip* is (active area)(electrical depth)(atom density), or $(0.04 \text{ cm}^2)(10^{-3} \text{ cm})(5 \times 10^{22} \text{ atoms/cm}^3) = 2 \times 10^{18}$ atoms. Further, the *active cross-section* = (active atoms)/(absorption cross section) = $(2 \times 10^{18} \text{ atoms})/(2 \times 10^{24} \text{ atoms/cm}^2) = 10^{-6}$ cm². Accelerated testing on a 4096 bit bipolar memory chip revealed a soft-fail cross-section of 40×10^{-12} cm²/bit. Thus the *SER cross section* = $(4096 \text{ bits})(40 \times 10^{-12} \text{ cm}^2/\text{bit}) = 1.6 \times 10^{-7}$ cm². The ratio of the active cross-section to the SER cross-section = $(1 \times 10^{-6} \text{ cm}^2)/(1.6 \times 10^{-7} \text{ cm}^2) = 6$. This means that one in six neutron-silicon “hits” will result in a soft fail.

Consequently, the problem increases with smaller dimensions. If the active circuit thickness is reduced to one micron, a soft fail will result from almost every hit!

7.6.6.4 Energetic Particle Damage in Silicon When a high-energy cosmic ray or alpha particle penetrates a reverse-biased p–n junction, it leaves a track of ions and electrons in its wake, as illustrated in Fig. 7-14. In the DRAM cell shown, the plasma-like track collapses the depletion layer and distorts the equipotentials there into the shape of a cylindrical funnel. This “funneling” produces a very large electric field that propels electrons into the depletion layer while holes flow down into the bulk Si. If the particle track impales or passes

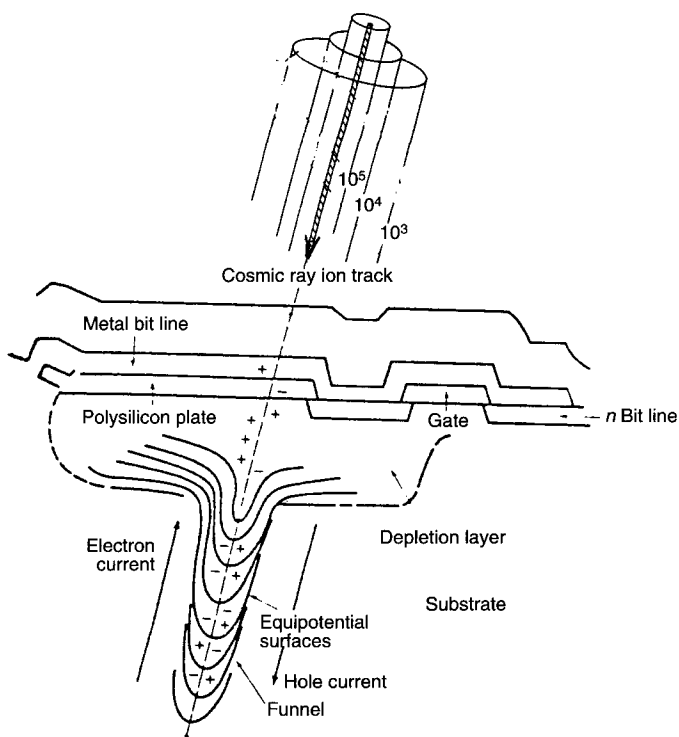


Fig. 7-14 Collapse of DRAM cell depletion layer along cosmic ray track produces a funnel in the device bulk. Representative dose profiles around the single ionizing particle are indicated. After reference 36.

close to a memory charge-storage node, an upset error may occur within nanoseconds. Subsequent diffusion in the next microsecond may distribute charge to neighboring nodes, causing upset there. Finally, the depletion-layer damage relaxes back to its original state without any permanent damage to devices. The funneling phenomenon has been numerically modeled (47) and experimentally verified through charge collection measurements (48).

It has been found that a critical charge (Q_C) is required to produce upsets, and that Q_C depends on the particular device and feature dimensions (l) as shown in Fig. 7-15. The linear relationship between the variables is well described by

$$Q_C = 0.023 l^2, \quad (7-46)$$

with Q_C in pC and l in μm . Interestingly, this equation is independent of whether Si or GaAs devices are involved. In $1 \mu\text{m}$ DRAM cells, the critical

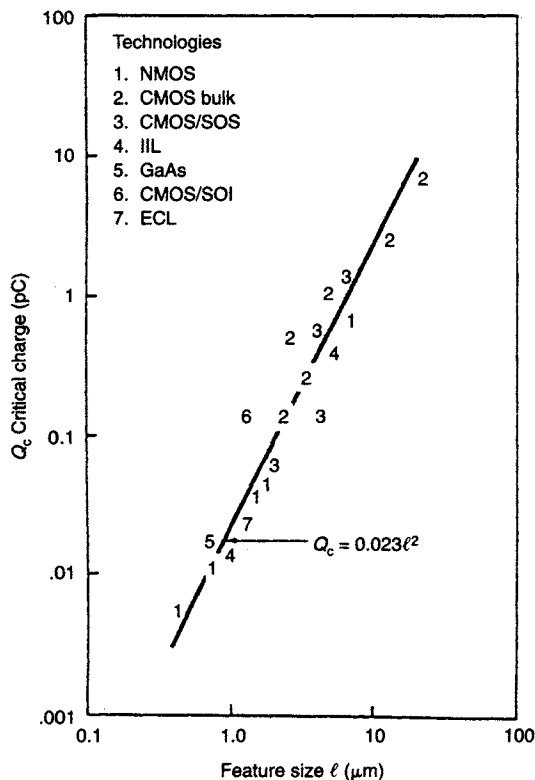


Fig. 7-15 Critical charge for SEU (Q_c) vs feature size in different devices. From E. L. Peterson and P. Marshall, *J. Rad. Eff. Res. and Eng.* (Jan 1989).

charge to produce upset is about 2.3×10^{-14} C, or close to 1.5×10^5 electronic charges. If hit by a 5 MeV alpha particle, it has been estimated that the particle range is a few tens of microns in Si; because the energy to produce electron-hole pairs in Si is 3.6 eV, about 1.4×10^6 electronic charges can be expected. Thus, it is easy to see why alpha particles can trigger soft memory errors. All that is needed is a sufficient flux of them.

7.6.7 Radiation Hardening of Devices

We close the chapter by commenting on strategies employed to protect or harden devices against the effects of radiation, and in particular, against soft errors. The increasing amount of critical spaceborne electronics in satellites, deep space probes, and strategic weapons systems has driven the need for increased performance and reliability of circuits employed in the space radiation environment. Perhaps the primary method to harden both silicon and GaAs devices against

radiation has been through circuit design. This subject is well beyond the scope of this book, but has been reviewed by Kerns and Schafer (49). In the processing area a favorite strategy has been to isolate devices from electrical continuity with the base semiconductor wafer. Thus dielectric isolation, or the imposition of an oxide layer or moat around individual MOSFETs or bipolar transistors, has been extensively practiced in silicon technology. Numerous silicon on insulator (SOI) approaches, including growth of Si on sapphire (SOS), separation by ion implantation of oxygen (SIMOX), zone melt recrystallization (ZMR), porous oxidized Si (FIPOS), and epitaxial lateral overgrowth (ELO), have been attempted (35). In general, dielectric isolation has removed latch-up paths and reduced volumes around junctions and oxides where carrier generation and trapping can occur.

With respect to soft errors, it is instructive to consider the techniques adopted to minimize their influence in 16 Mb DRAMs subjected to α particles (51). In addition to error-correcting codes, the α particle sensitivity was reduced by the following means:

1. The use of a deep-trench capacitor. Charge stored on this capacitor was designed to be greater than that transferred by three α particles traversing the longest diagonal through the memory-cell volume.
2. Minimizing charge collected by p-n junctions of the MOS transfer device source and drain implants. By fabricating the transfer device in an n -well, charge generated in the substrate is collected by the well supply.
3. Reducing charge collection efficiency by storing memory data "inside" rather than "outside" the dielectrically isolated storage trench. In the latter, charge is collected over a larger potential substrate volume and source of radioactive atoms.

EXERCISES

- 7-1 An electronic vacuum tube 20 cm^3 in volume is pumped to 10^{-11} atm prior to glass-metal sealing. For a moisture penetration rate of $1 \times 10^{-8} \text{ atm-cm}^3/\text{s}$, how long will it take for the pressure in the tube to rise an order of magnitude?
- 7-2 The mean time to failure of plastic packages as a function of temperature and relative humidity is given by $\text{MTTF} = B(\text{RH})^{-n} \exp [E_c/kT]$, where B is a constant.
- a. If $n = 2.66$ and $E_c = 0.79 \text{ eV}$, does a 10% increase in relative humidity have a greater effect in reducing mean life than a 10% increase in temperature?

- b. For relative humidity and temperature variables to have an equal influence in degrading plastic packages, as measured by MTTF, what relationship must connect RH, T , n and E_c ?
- 7-3 An aluminum contact film that is $1\text{ }\mu\text{m}$ thick and $0.1\text{ mm} \times 0.1\text{ mm}$ in area carries a current of $1\text{ }\mu\text{A}$. Under a worst-case corrosion scenario how long would it take the metal to be totally converted into ions?
- 7-4 Write possible balanced chemical reactions for the formation of NiCl_2 and NiCO_3 corrosion products when Ni is exposed to the atmosphere and water containing chlorides.
- 7-5 a. Is “purple plague” formation likely to occur during aqueous corrosion of a gold-wire bond to an aluminum contact pad?
b. Suppose a condensed moisture layer containing a 10^{-6} molar concentration of Al^{+3} surrounds the pad, while a 10^{-8} molar concentration of Au^{+3} coats the wire. What maximum potential can develop between these metals? What minimum potential can develop?
- 7-6 It has been claimed that the Au_4Al intermetallic compound corrodes in the presence of bromine ions according the reaction $\text{Au}_4\text{Al} + 3\text{Br}^- \rightarrow 4\text{Au} + \text{AlBr}_3 + 3\text{e}^-$. Suggest a subsequent reaction that regenerates bromine ions. What evidence is there that such a reaction sequence occurs?
- 7-7 The contact metal WSi_2 is subject to electrolytic corrosion in the presence of moisture and voltage bias. Write a chemical reaction that would tend to convert this conducting silicide to insulating oxide byproducts. If the corresponding free energies of formation (ΔG) from elements at 298K are $\Delta G(\text{WSi}_2) = -22\text{ kcal/mol}$, $\Delta G(\text{WO}_3) = -204\text{ kcal/mol}$, and $\Delta G(\text{SiO}_2) = -182\text{ kcal/mol}$, what is the overall free energy change of the corrosion reaction? Note: You may have to consult a table of thermodynamic data.
- 7-8 The use of aluminum electrical wiring in the home has been known to cause fires that apparently initiate at plug outlets; in contrast, copper wiring appears to be immune from this problem. Give a possible reason for these observations.
- 7-9 Suppose corrosion of metal within a plastic package occurs by moisture permeation followed by a chemical reaction described by second-order kinetics such that $dC/dt = -k(C - C_0)^2$ where k is the rate constant. Derive an expression for the reliability function.
- 7-10 What similarities and differences in matrix environment, mass-transport mechanisms, current densities, damage morphology, temperature depen-

- dence, and degradation kinetics exist between wet migration and solid-state electromigration?
- 7-11 During electrolytic refining of copper, the impure-anode sheet metal is plated onto pure Cu cathodes in cells that employ aqueous CuSO_4 electrolytes. In this process, as with wet metal migration, dendrites often span the anode–cathode spacing and short the electrodes. Suggest reasons why the formation of dendrites is so heterogeneous and why they grow with such variable velocity. In the case of copper refining, what practical measures can be taken to minimize cell shorting by dendrites?
- 7-12 The 100 nm thick SiO_2 film of a capacitor was uniformly contaminated with a concentration of 1×10^{15} sodium ions/ cm^3 . Application of 5 volts swept all of the sodium to the negative electrode.
- How long does it take Na^+ ions to electromigrate across the oxide at 300K?
 - The voltage is reduced to zero. Sketch resultant time-dependent concentration profiles of Na^+ as a function of time.
 - How long will it take the Na^+ to diffuse back across the oxide?
- 7-13 As a result of exposure to radiation:
- Metals harden but become less ductile and suffer damage in the form of creep, void swelling, and hydrogen embrittlement.
 - Ceramics undergo color changes and exhibit loss of thermal conductivity but enhancement of electrical conductivity.
 - Polymers generally lose strength and ductility, discolor, and evolve gas (e.g., hydrogen).
- Select one of these material classes and write a report explaining reasons for the radiation-damage effects manifested. (The April 1997 issue of the MRS Bulletin is devoted to materials performance in radiation fields.)
- 7-14 How many electron–hole pairs are created in Si and GaAs by exposing each to a radiation dose of 1 Gy?
- 7-15 Nuclei of Th^{232} have an activity of 1.1×10^{-7} Ci/g and release 6 alpha particles per disintegration when they decay to Pb^{206} . If these alpha particles can penetrate 50 μm of Si, what αP flux can be expected?
- 7-16 A source of neutrons impinges on both silicon and germanium. What ratio of subsurface interaction depths will enable the probability of neutron–Ge atom hits to equal the number of neutron–Si atom hits?
- 7-17 Soft-error fail rates per chip in 288 Kb DRAMS (1986) were measured to be approximately 1 fail per year. More recent data for 4 Mb DRAM (1993)

memory chips from different manufacturers reveal SER values of 0.0026 to 0.00046 fails per year.

- a. Suggest possible reasons for the differences in these values.
- b. How many FITS does 0.00046 fails per year correspond to?

- 7-18 A 64 Mb bipolar cache memory chip containing 5×10^8 bits in the memory is exposed to an energetic nucleon flux equal to $10^5/\text{cm}^2\text{-yr}$. The probability of a soft error is given in terms of the cross-section, which has a value of $4 \times 10^{-12} \text{ cm}^2/\text{bit}$. How many soft errors can be expected if this chip operates for a year?

REFERENCES

1. R. B. Comizzoli, R. P. Frankenthal, R. E. Lobnig, G. A. Peins, L. A. Psota-Kelty, D. J. Siconolfi, and D. J. Sinclair, *The Electrochem. Soc. Interface* 26, Fall (1993).
2. D. J. Schneider and D. C. Williams, *Lasers and Optronics* 25, Oct. (1994).
3. J. D. Guttenplan, *ASM Handbook*, Vol. 13, 1107 (1987).
4. E. White, G. Slenski and B. Dobbs, *ASM Handbook*, Vol. 13, 1113 (1987).
5. M. Pecht, *IEEE Trans. on Components, Hybrids and Manuf. Tech.* 11(3), 383 (1990).
6. R. P. Merrett, *Tutorial Notes, IEEE International Reliability Physics Symposium*, 1-2.1 (1984).
7. A. Roth, *Vacuum Technology*, North-Holland, Amsterdam (1976).
8. D. A. Jones, *Principles and Prevention of Corrosion*, Macmillan, New York (1992).
9. M. G. Fontana, *Corrosion Engineering*, 3rd ed., McGraw Hill, New York (1978).
10. M. Tullmin and P. R. Roberge, *IEEE Trans. on Reliability* 44, 271 (1995).
11. X. Shan and M. Pecht, in *Thermal Stress and Strain in Microelectronics Packaging*, ed. J. H. Lau, van Nostrand Reinhold, New York (1993).
12. G. L. Schnable, R. B. Comizzoli, W. Kern, and L. K. White, *RCA Review* 40, 416 (1979).
13. D. S. Peck, *24th Annual Proceedings of the IEEE Reliability Physics Symposium*, p. 44 (1986).
14. M. Iannuzzi, *IEEE Trans. on Components, Hybrids and Manuf. Tech., CHMT-6*, 181 (1983).

15. P. Totta, *J. Vac. Sci. Tech* **13**, 26 (1976).
16. S. Thomas and H. M. Berg, *23th Annual Proceedings of the IEEE Reliability Physics Symposium*, p. 153 (1985).
17. A. J. Griffen, S. E. Herna'ndez, F. R. Brotzen, J. D. Lawrence, J. W. McPherson, and C. F. Dunn, *31st Annual Proceedings of the IEEE Reliability Physics Symposium*, p. 327 (1993).
18. K. H. Ritz, W. I. Stacy, and E. K. Broadbent, *25th Annual Proceedings of the IEEE Reliability Physics Symposium*, p. 28, (1987).
19. W. T. Anderson and A. Christou, *IEEE Trans. on Reliability*, R-29, 222 (1980).
20. R. S. Torkington and D. Lamoreaux, *ASM International Symposium for Testing and Failure Analysis (ISTFA)*, p. 225 (1990)
21. J. M. E. Harper, E. G. Colgan, C.-K. Hu, J. P. Hummel, L. P. Buchwalter, and C. E. Uzoh. *MRS Bulletin*, **XIX** (8), 23 (1994).
22. S. J. Krumbein, *IEEE Trans. on Reliability* **44** (4), 539 (1995).
23. T. Saigusa, T. Ohtaki, M. Nakajima, and Y. Nakao, *Materials Developments in Microelectronic Packaging Conf. Proc., Quebec, Canada*, p. 345, Aug (1991).
24. M. Pecht, *IEEE Trans. on Components, Hybrids and Manuf. Tech.* **13**, 383 (1990).
25. B. Rudra and D. Jennings, *IEEE Trans. on Reliability* **43**, 354 (1994).
26. G. T. Kohman, H. W. Hermance, and G. H. Downes, *Bell Syst. Tech J.* **XXXIV**, No. 6, 1115 (1955).
27. J. J. Steppan, J. A. Roth, L. C. Hall, D. A. Jeanotte, and S.P. Carbone, *J. Electrochem. Soc.* **134** (1), 175 (1987).
28. G. DiGiacomo, in *Microelectronics Manufacturing Diagnostics Handbook*, ed. A. H. Landzberg, Van Nostrand Reinhold, New York (1993).
29. M. Ohring, *The Materials Science of Thin Films*, Academic Press, Boston (1992).
30. R. T. Howard, *IEEE Trans. on Components, Hybrids and Manuf. Tech.* CHMT-4, 520 (1981).
31. *Handbook of Chem and Physics*, 59th Edition, CRC Press, Cleveland (1978-79).
32. R. L. Hance, J. W. Miller, K. B. Erington, and M. A. Chonko, *Tutorial Notes, IEEE International Reliability Physics Symposium*, 4.1 (1995).
33. H. Hughes in T. P. Ma, and P. V. Dressendorfer, eds., *Ionizing Radiation Effects in MOS Devices and Circuits*, J. Wiley and Sons, New York (1989).
34. T. P. Ma and P. V. Dressendorfer, eds., *Ionizing Radiation Effects in MOS Devices and Circuits*, J. Wiley and Sons, New York (1989).

35. A. Holmes-Siedle and L. Adams, *Handbook of Radiation Effects*, Oxford University Press, Oxford (1993).
36. G. C. Messenger and M. S. Ash, *The Effects of Radiation on Electronic Systems*, 2nd ed., Van Nostrand Reinhold, New York (1992).
37. J. R. Srouf and J. M. McGarrity, *Proc. IEEE* **76**, 1443 (1988).
38. K. F. Galloway and R. D. Schrimpf, *Microelectronics Journal*. **21**, 67 (1990).
39. G. J. Papaioannou, in A. Christou and B. A. Unger, eds., *Semiconductor Device Reliability*, Kluwer, Amsterdam (1990).
40. L. Adams, *Microelectronics Journal* **16**, 17 (1985).
41. J. F. Ziegler and G. R. Srinivasan, eds., *IBM J. Res. and Develop.* **40** (1) (1996).
42. T. C. May and M. H. Woods, *IEEE Trans. on Electron Devices*, ED-26, 2 (1979).
43. J. F. Ziegler and W. A. Lanford, *Science* **206**, 776 (1979).
44. L. Lantz, *IEEE Trans on Reliability* **45** (2), 174 (1996).
45. J. F. Ziegler, *IBM J. Res. and Develop.* **40** (1), 19 (1996).
46. R. Baumann, T. Hossain, S. Murata, and H. Kitagawa, *33rd Annual Proceedings of the IEEE Reliability Physics Symposium*, p. 296 (1995).
47. H. L. Grubin, J. P. Kreskovsky, and B. C. Weinberg, *IEEE Trans on Nuclear Science*, NS-31, 1161(1984).
48. G. C. Messinger, *IEEE Trans on Nuclear Science*, NS-29, 2024 (1982).
49. S. E. Kerns and B. D. Shafer, *Proc. IEEE* **76**, 1470 (1988).
50. C. H. Stapper and W. A. Klassen, *30th Annual Proceedings of the IEEE Reliability Physics Symposium*, p. 3 (1992).

Chapter 9

Degradation of Contacts and Package Interconnections

9.1 INTRODUCTION

The unifying theme of this chapter encompasses problems that arise in maintaining electrical continuity between conductors. We have all instinctively diagnosed poor contacts as the source of malfunction in some piece of electrical equipment or other, and often been correct. In the days before transistors, the simple act of making a local telephone call activated hundreds of electro-mechanical contacts, whose inefficient operation or malfunction meant signal distortion, noise, or even lack of signal transmission. Whether they are “static,” as in permanent soldered interconnections, or are “dynamic,” sliding and rotating, as in switches and motors, electrical contacts pose a variety of interesting and important reliability problems.

In this chapter we will discuss both types of contacts, starting with a classical treatment of the nature of contacting, current-carrying conductors. Dynamic contact problems arising from mechanical wear, corrosion, and stress relaxation are among the first reliability issues we will address. Much of the chapter’s later focus, however, will be devoted to the reliability of solder materials used in interlevel packaging connections. In this sense the chapter continues the discussion that began in Chapter 8 dealing with packaging configurations, joining techniques, and solder defects.

Building on the previously developed thermal-stress analyses, failure modeling of contacts and solder interconnections is possible, at least in principle. Nevertheless, it still remains a formidable task to obtain estimates for contact failure times because

1. stress or strain levels required to cause failure are often unknown;
2. the applied stresses are usually cyclic in nature;
3. component contact and structural geometries are complex;
4. contacting surfaces may wear due to the relative motion between them; and
5. the elevated temperature plastic and viscoelastic behavior of the involved materials often displays a complex temperature–time–environment history.

When applied to solder interconnections, a logical failure-modeling approach first assumes the operative thermal stress levels expected as calculated previously. The cyclic nature of the stressing must then be considered and this information folded together with material properties into the relevant expressions that describe the deformation response. Creep, fatigue, and fracture are the operative mechanisms that cause packaging components and structures to fail. These important subjects are basically the substance of two of the later chapter sections.

Finally, there is a discussion of vibration and dynamic loading effects. In addition to static and cyclic differential thermal-expansion effects, dynamic loading and vibration can also be quite harmful from the standpoint of equipment reliability. Because of potential fatigue damage to soldered leads due to vibratory flexure of circuit boards and components, the subject is treated in this chapter.

9.2 THE NATURE OF CONTACTS

9.2.1 Scope

The purpose of electrical contacts is threefold: to make or complete a circuit, to maintain current flow through the circuit, and to break or open the circuit when required. Contacts must reliably repeat these functions over specified time intervals that may be as short as a fraction of a millisecond. The electrical spectrum of contacting devices and methods includes switches of all kinds, relays, pin and socket connectors, as well as solder contacts. Although there is considerable overlap, applications may be broadly classified according to their power usage as follows:

1. power engineering, where high voltages and currents are involved;
2. medium power, with moderate voltages and currents, as in home appliances; and
3. low voltage–low current applications, as in telecommunications and computers.

In these applications, electrical continuity is often achieved mechanically through impact, sliding, or rotational motion of metal surfaces that are forced together to make and sustain contact. Different mechanical phenomena are involved in each of these domains of usage, and the corresponding contact-failure mechanisms reflect this. In the first category, for example, electrical discharges and arcs may be operative, causing extensive erosion and wear of the contacting metal-surfaces. Contacts also degrade by mechanical wear in medium and low-power applications, where in addition, contact opening and closing causes insulating surface films to form as a result of (dry) oxidation or (wet) corrosion. In such cases, contact unreliability arises through destruction of the initially smooth contacting and conducting surfaces. The legacy is open and noisy or intermittent electrical behavior.

There is considerable literature on the subject of contacts (1, 2). Investigations continue to be carried out on contact problems, where for over 40 years there has been an annual IEEE Holm Conference on Electrical Contacts, so named for R. Holm, a prominent researcher in this field. In these conference proceedings, also published in the *IEEE Transactions on Components, Hybrids and Manufacturing Technology*, many papers are devoted to damage of high-power contacts, a subject we will not discuss further. Instead, the subsequent treatment will be limited to low- and medium-power contacts, the involved materials, and the degradation and reliability issues that surround them.

9.2.2 Constriction Resistance

Consider the physics of what happens when two nominally identical smooth metals contact one another. Although there appears to be contact over a substantial area, actual contact is made at a limited number of microscopically small areas. (If the involved metals were infinitely hard, there would be only three points of contact.) With analytical tools such as the scanning electron microscope and atomic force microscope, we now know that even very finely polished metal surfaces exhibit arrays of protruding surface asperities separated by valleys, with peak-to-peak height differences of perhaps 10 nm.

When a potential difference is applied across touching asperity-free cylindrical conductors, the current flow lines are parallel, as shown in Fig. 9-1a. The contact voltage (V_c) that develops between points *A* and *B* is simply given by Ohm's law; i.e., $V_c = i R_o$, where i is the current and R_o is the metal ohmic resis-

tance. This situation is approximated by a soldered joint because the solidified solder effectively fills the spaces between asperities. However, when contact occurs only at asperities, current flows in the constricted manner of Fig. 9-1b, distorting the flow lines. The contact voltage (V_c) developed is now larger in order to maintain the same current and is given by

$$V_c = i (R_o + R_c), \quad (9-1)$$

where R_c is the so-called constriction resistance. Analogy to fluid flow through a constricted pipe under a fixed pressure head (constant V_c) means a smaller throughput (i), which is caused by the added resistance (R_c). If, in addition, there is a resistance contribution R_f due to a surface film at the contact, it is added to $R_o + R_c$.

Using concepts of electromagnetic theory it has been shown that when two plane surfaces make contact over an asperity area of radius a , the total constriction resistance is

$$R_c = \frac{\rho}{2a}, \quad (9-2)$$

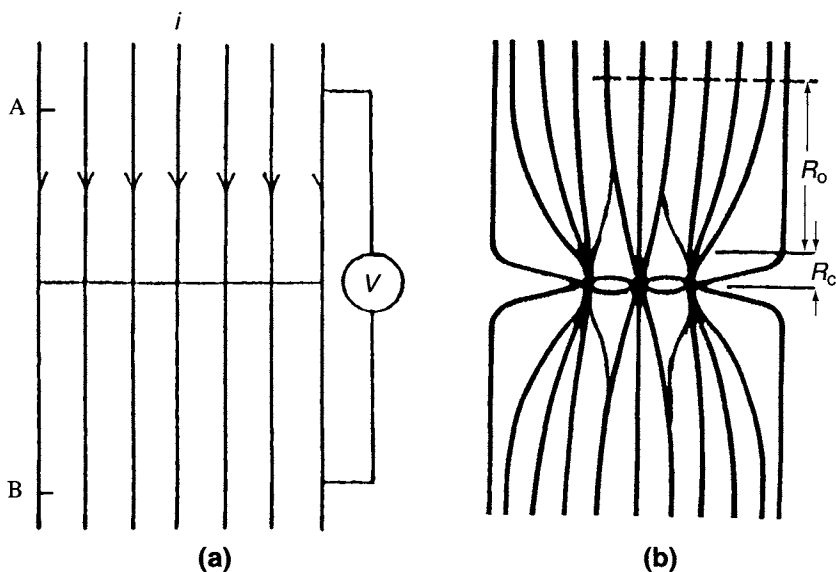


Fig. 9-1 (a) Parallel current flow lines over a large asperity-free area of contact. (b) Constricted current flow lines at contact asperities.

where ρ is the resistivity. For metals we recall that $\rho \approx 10^{-6}$ to $10^{-4} \Omega\text{-cm}$, and, we note that if a is large, R_c is negligible; but if a is sufficiently small, R_c can rise to unacceptable levels and cause circuit malfunction.

9.2.3 Heating Effects

One might imagine currents to cause appreciable local heating effects at the asperity-asperity contact welds, known as **a-spots**, because of the small areas involved. Then the contact resistance would rise from $R_c(1)$ at temperature $T(1)$ to $R_c(2)$ at $T(2)$ according to the relation $R_c(2) = R_c(1) \{1 + \alpha [T(2) - T(1)]\}$, where α is the temperature coefficient of resistivity. Assuming that no heat is lost from the hot contact constriction and that steady-state temperatures are quickly reached, it is not hard to roughly estimate the temperature rise, $T(2) - T(1)$. The power generated from Joule heating (V_c^2/R_c) is converted into an equivalent heat (power) flow ($\kappa[T(2) - T(1)] A_c/\Delta x$), where A_c and Δx are the constriction area and length, respectively, and κ is the thermal conductivity. Noting that $R_c = \rho \Delta x/A_c$, and equating the two power contributions, the approximate expression

$$V_c^2 = \rho \kappa [T(2) - T(1)] \quad (9-3)$$

emerges. In metals the electrical conductivity ($\sigma = 1/\rho$) and thermal conductivity are proportional to each other through the formula, $\rho\kappa = \mathcal{L}T$, where $\mathcal{L}$ is the Lorenz number (Sect. 5.5.3). If T is assumed to be some mean temperature between $T(1)$ and $T(2)$, then $V_c^2 = \mathcal{L}T[T(2) - T(1)]$. A more accurate expression (3) is

$$V_c^2 = 4\mathcal{L}[T^2(2) - T^2(1)]. \quad (9-4)$$

To see the implications of this formula, let us assume $V_c = 1\text{V}$ and $T(1) = 300\text{K}$. Taking $\mathcal{L} = 2.44 \times 10^{-8} \text{ W-}\Omega/\text{K}^2$, $T(2)$ is calculated to be 3200K , a value much higher than the melting point of almost all metals. The mere hint of such high temperatures implies sources of reliability problems. If, for example, the contact area at asperities were 10^{-8} cm^2 , the current density would be sufficiently high for electromigration damage to occur; and if contact tarnish films existed, they would probably not survive the heat generated.

9.2.4 Mass Transport Effects at Contacts

The unusually high temperatures and electric current densities present at contact spots mean that at least three different atomic mass transport mechanisms may be operative to transfer material from one contact metal to the other (4). Diffusion, electromigration, and thermomigration are the mechanisms, and they are explored in this section.

a. Diffusion The diffusional flux of atoms (J_a) in a concentration gradient (dC/dx) driving force, with x normal to the contact surface, is a subject treated in this book on at least two prior occasions. Because a discontinuous concentration gradient exists for atoms at a contact junction between two different metals, the mass flux given by

$$J_m = -D \frac{dc}{dx}, \quad (5-1)$$

is potentially large. For appreciable interdiffusion to occur, the diffusivity given by $D = D_0 \exp [-E_D/RT]$ must be sufficiently large, and this requires a high temperature. Even if both contacting metals are identical, atomic diffusion is possible. In this case the driving force is the surface-energy gradient that stems from differences in surface curvature. Atoms at asperity tips will tend to diffuse into the valleys and lower the overall surface area. Essentially the same mechanism that is operative during the sintering of metal powder compacts is active here. Sintering or welding at hot spots effectively widens the contact area and reduces the contact resistance.

b. Electromigration It was previously shown that the electromigration flux (J_e) is given by

$$J_e = \frac{C D Z^* q \rho j_e}{RT}. \quad (5-20)$$

Just as on the chip level at metal-semiconductor interfaces, electromigration at contacts often represents a severe mass-flux divergence. Electromigration mass-transport effects are also expected to be narrowly confined to the contact interface where the current density (j_e) is large and the temperature high. Further away the conductor widens lowering j_e as well as the temperature. An important term in Eq. 5-20 is the effective valence Z^* whose value may range widely in magnitude, i.e., from ~ -5 to -50 . Measurement has shown that Z^* decreases slowly with increasing temperature.

c. Thermomigration Mass transport driven by temperature gradients is known as thermomigration. The corresponding flux J_{th} is given by

$$J_{th} = - \frac{C Q^* D}{RT^2} \frac{dT}{dx}, \quad (9-5)$$

where Q^* is the heat of transport for the element in question. A positive value for Q^* implies atom migration to colder regions.

The ratio of the electro- to thermo-migration fluxes is given by

$$\frac{J_e}{J_{th}} = \frac{Z^* q \rho j_e}{-(Q^*/T) dT/dx} \quad (9-6)$$

and simply represents a ratio of the two generalized forces.

Example 9-1 Assuming representative values of $Z^* = -10$, $Q^* = 10$ kJ/mol, $j_e = 1 \times 10^{10}$ A/m², $\rho = 1 \times 10^{-7}$ Ω -m, $T = 1000$ K and $dT/dx = 10^7$ K/m, does electromigration or thermomigration dominate?

Answer Substituting in Eq. 9-6 we have

$$\frac{J_e}{J_{th}} = \frac{-10 \times 1.6 \times 10^{-19} \times 1 \times 10^{-7} \times 1 \times 10^{10} \times 6.02 \times 10^{23}}{-(10,000/1000) \times 10^7} = 9.6.$$

(Note that to make units match, Avogadro's number must be included in the numerator.) This calculation suggests that electromigration dominates thermo-transport.

Without knowing the temperature distributions and material constants in a contact it is sometimes easier to make measurements. This was done by Runde *et al.* (5), who deposited minute quantities of radioactive zinc on one of two aluminum contact electrodes. A buildup of Zn observed on the counter electrode, consistent with the current polarity, demonstrated that electromigration had occurred. Transfer of metal at contacts changes the alloy composition and generally raises R_c .

9.2.5 Contact Force

Applied force or pressure is required to insure that low-resistance electrical contact is made and then sustained. We have already seen that contact resistance depends on the contact area. Since contact heating can potentially alter the area through local mass transfer and/or melting, the relationships among contact resistance, force, and temperature are critical to an understanding of contact performance; what is required is a combination of a few simple definitions and formulas.

Let us first consider hardness H , a property measured in stress units of N/m², or commonly kg/mm². Hardness is essentially a measure of a material's ability to withstand penetration by a harder indenter that is pressed into its surface with normal load P . For a spherical indenter, the hardness is often defined as the load per projected area of diameter $\bar{d}$, or $H = 4P/\pi\bar{d}^2$. We know, however, that the true mechanical bearing area is but a small fraction of the nominal contact area. As a result, the asperity contacts are subjected to large pressures even with small loads. To account for normal contacts at multiple asperities, the hardness formula is modified to

$$H = \frac{4P}{n \pi \bar{d}^2}, \quad (9-7)$$

where n is the number of such contact points or spots, and $\bar{d}$ is the average spot diameter. Reconsidering the constriction resistance, it is apparent that

$$R_c = \frac{\rho}{nd}. \quad (9-8)$$

This formula assumes that the resistor spots are effectively connected in parallel. Eliminating $\bar{d}$ between the last two equations yields

$$R_c = \rho \left(\frac{\pi H}{4nP} \right)^{1/2}, \quad (9-9)$$

with the expected result that the contact resistance decreases as the load increases. This equation holds when P is relatively large and the contact metal deforms plastically. However, when P is small such that the average pressure is roughly less than $H/3$, the contact deforms elastically; in this case theory shows that $R_c \approx P^{-1/3}$. Many experiments have been performed establishing the validity of both the $P^{-1/2}$ and $P^{-1/3}$ dependencies for R_c . Contact integrity necessitates the maintenance of appropriate contact pressures.

9.3 DEGRADATION OF CONTACTS AND CONNECTORS

9.3.1 Introduction

Reliable separable electrical connectors, like the kind shown in Fig. 9-2, require the force pressing male and female halves together to make contact, be constant, and not relax with time. Since lifetimes ranging from 10 to as much as 40 years are often desired for such connectors, the contacts must suffer little change in R_c . In this section we address practical situations that lead to an increase in R_c because of a change in one or more of the terms in Eq. 9-9. After treating the mechanics of spring contacts, the first case we consider involves creep-prone contacts. These pose a reliability issue because the contact pressure relaxes with time, raising the contact resistance. Most contacts in relative motion suffer tribological or rubbing-type damage through wear processes that potentially alter both the geometry and composition of the contacting surfaces. In this way ρ , H , n , and even P change in ways that are not easily distinguishable, but which effectively raise R_c and limit contact life.

9.3.2 Mechanics of Spring Contacts

Metal prongs that make electrical contact in separable connectors have been approximated by dual cantilevers; an example is the card edge connector. Therefore, let us consider the simple cantilever beam (6) shown in Fig. 9-3. With force

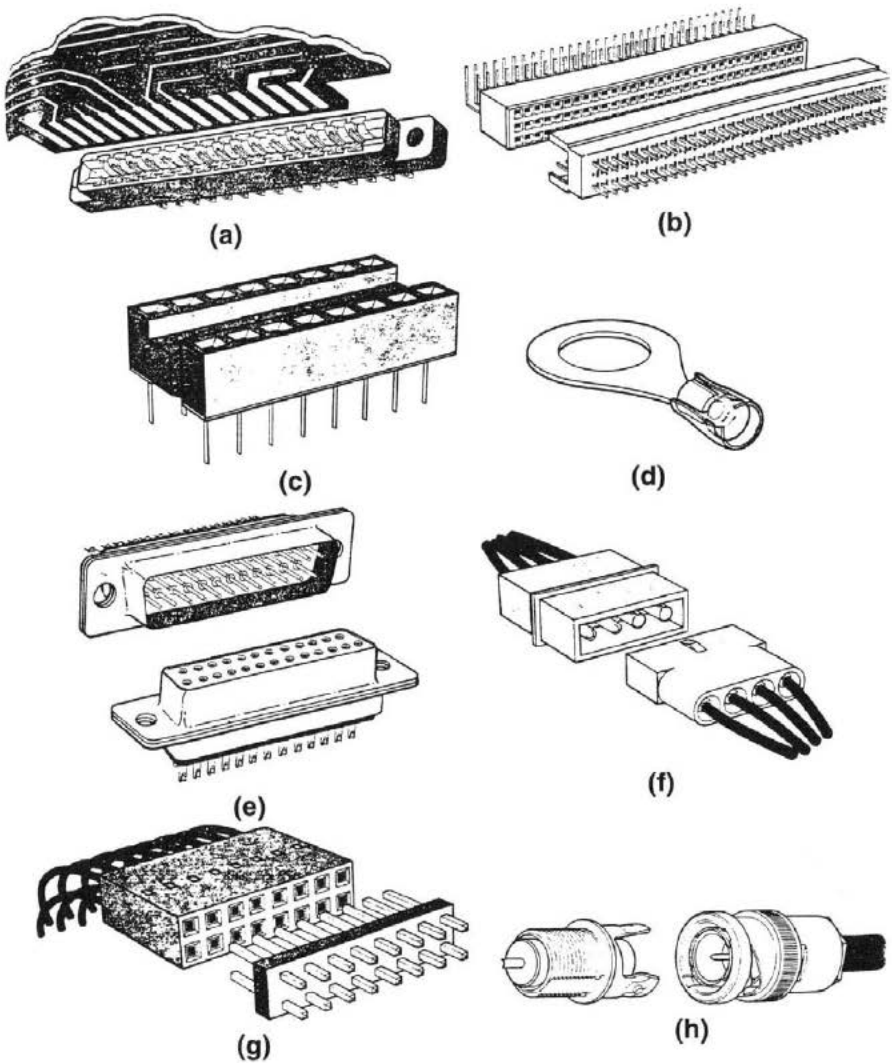


Fig. 9-2 An assortment of separable electrical connectors. (a) Edge card/PBB fingers; (b) Two-part/euro or DIN; (c) IC socket; (d) Crimp; (e) Pin and socket/track and panel; (f) Power/snap together; (g) Post and box; (h) Coaxial (BNC). From T. A Yeger and K. Nixon, *IEEE Trans. Components, Hybrids and Manuf. Tech.* 7, 370 (1984)

P acting on a such a beam of width (w) and thickness (T), the deflection (δ) is given by

$$\delta = \frac{4PL^3}{wT^3E} \quad (9-10)$$

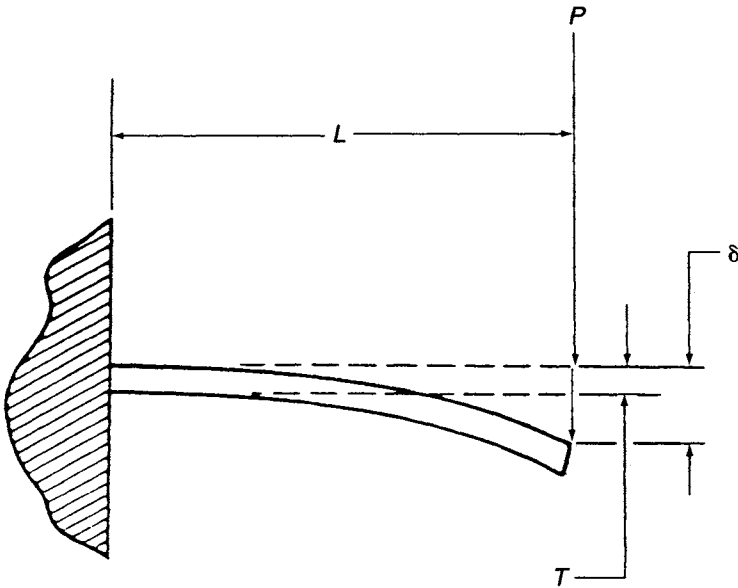


Fig. 9-3 Cantilever model of a contact spring.

in elementary texts on strength of materials, where E is Young's modulus. For spring contacts deflecting a fixed amount, a large E is desirable because P is then large, thus reducing R_c .

It is also useful to know the maximum force ($P_{\max}$) realized by an elastic spring; clearly, the flow or yield stress (σ_o) of the metal limits $P_{\max}$. The maximum stress in the outer fibers of a beam is $6 PL/(wT^2)$ and is limited by σ_o . Therefore,

$$P_{\max} = \frac{\sigma_o w T^2}{6L}. \quad (9-11)$$

For those copper alloys commonly used in connector prongs, Young's modulus varies only between 110 and 140 GPa. Therefore, one practical way to ensure large normal spring forces is to select high strength alloys; this accounts for the widespread use of Cu-Be for this purpose. In addition, there is an added reliability bonus because such alloys suffer less permanent set. Because it reduces the beam deflection, permanent set can raise R_c by relaxing the contact force. Permanent set behavior in actual contact springs is displayed in Figs. 9-4a and b, where it is seen that stiffer, stronger alloys generally behave better in this regard.

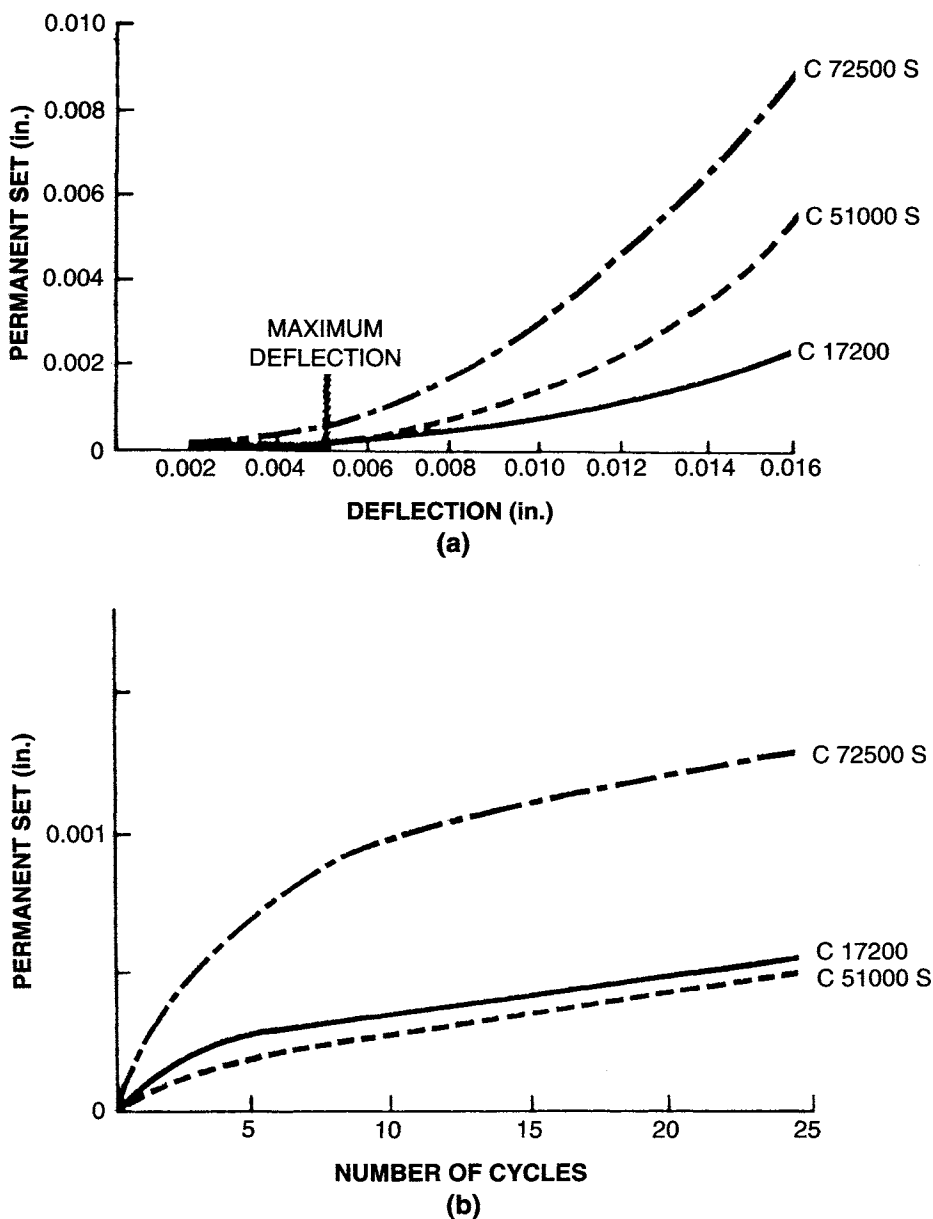


Fig. 9-4 (a) Permanent set in three different copper base alloy spring connectors as a function of deflection. Note: C17200 is beryllium copper ($E = 126\text{GPa}$, $\sigma_o = 721\text{MPa}$), C51000 is phosphor bronze ($E = 118\text{GPa}$, $\sigma_o = 658\text{MPa}$), C72500 is Cu-Ni-Sn ($E = 134\text{GPa}$, $\sigma_o = 630\text{MPa}$). (b) Permanent set in spring connectors as a function of the number of contact mating cycles. From reference 6.

9.3.3 Normal Force Reduction in Contact Springs

In order to quantitatively assess permanent set as manifested by service life, it is instructive to consider a study (7, 8) of the heat-age testing performed on gold-plated, phosphor-bronze contact springs. The latter were subjected to a matrix of accelerated aging tests spanning a temperature range of 100°C to 200°C, for time periods up to three months. Measurements of the spring deformation were made and converted to the equivalent normal force (P_N) applied.

In these two publications, it is interesting that essentially the same data were plotted in two different ways; the results are displayed in Fig. 9-5. A log-log plot was used in Fig. 9-5b to track the reduction in P_N as a function of time (t). The straight line suggests the empirical relationship, $\ln(P_N) = -n \ln(t) + b$, so that n , a constant, is the slope, and b , another constant, is the intercept on the ordinate. An equivalent explicit formula expresses the force-relaxation kinetics as

$$P_N = C t^{-n}, \quad (9-12)$$

where $C = \exp(b)$. Values of n ranged between 0.1 to 0.2 and increased with temperature. The neglect of the explicit role of temperature is a shortcoming of this analysis.

In Fig. 9-5a the data were plotted as P_N vs $\log t$, and the resulting straight line implies the linear equation, $P_N = r + s \ln t$, (r and s are constants). To account for the effect of temperature it was assumed that the rate of force recovery is thermally activated, and therefore,

$$\frac{dP_N}{dt} = -K \exp \left[-\frac{E - gP_N}{RT} \right]. \quad (9-13)$$

In this expression K and g are constants, and RT has the usual meaning. The activation energy (E) in the Boltzmann factor is linearly reduced by the instantaneous recovering force and the strain energy associated with it. A precedent for modifying E so that it is variable was established in Sect. 1.3.3 (Eq. 1-7). Integration of this differential equation yields the logarithmic recovery law

$$P_N = A \ln(Bt + C), \quad (9-14)$$

where $A = -RT/g$, $B = \frac{gK(\exp - \frac{E}{RT})}{RT}$, and C is a constant that can be evaluated from the initial condition (at $t = 0$, $P_N = P_N(0)$). When Bt is large compared to C , this solution is in accord with experiment and can be extrapolated to long times in order to predict connector lifetime. Further insight into the relaxation mechanism can be obtained by evaluating E and interpreting its magnitude.

This example illustrates the distinction between failure and reliability modeling. In the former, attempts are made to determine lifetime in terms of physical

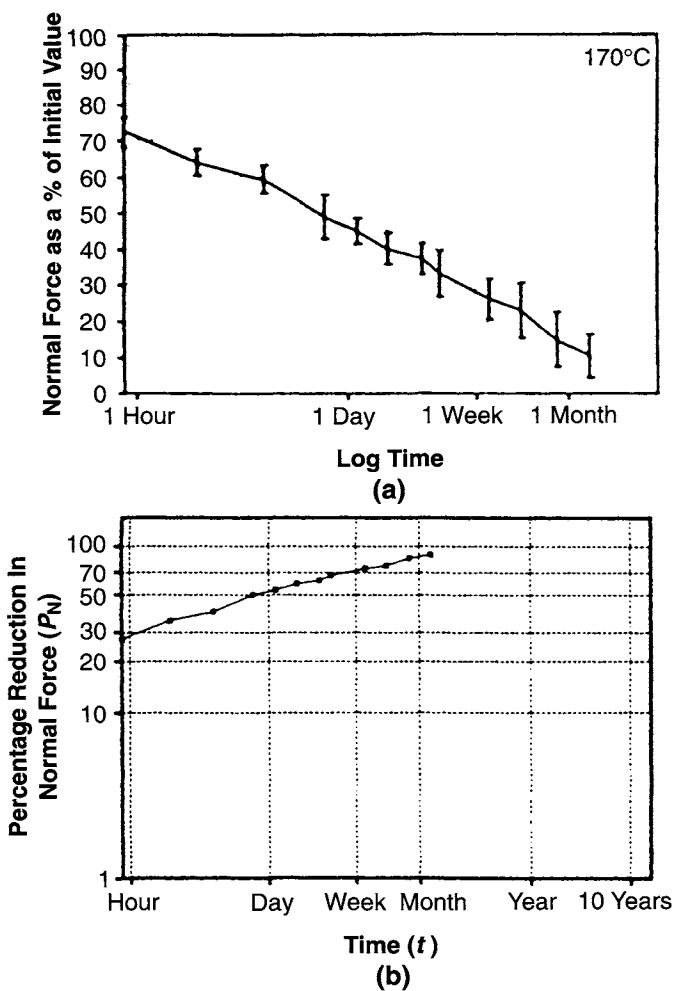


Fig. 9-5 (a) Force relaxation of a dual-cantilever electrical spring contact plotted as a function of log time. From reference 8. (b) Essentially the same data as in Fig. 9-5a plotted as log percent force reduction versus log time. From reference 7.

mechanisms that cause time-dependent changes in the material. Information provided by the failure modeling leading to Eq. 9-14 often yields ideas for corrective action that may eventually eliminate the problem. On the other hand, no such information or hope stems from the empirical statistical modeling of reliability implicit in Eq. 9-12.

9.3.4 Tribology

The subject of tribology encompasses phenomena that occur at material surfaces in contact and in relative motion. These include wear, friction, erosion, and lubrication. Of these subjects, wear is clearly most important for our needs. In electronic packaging, wear scars develop in cable connectors and on the tab portions of card edge assemblies and pluggable modules. Specific mechanisms for damage have been reviewed by Engel (8) and include:

a. Adhesive wear Surface smearing, galling, and seizure are the manifestations of the adhesive wear that results from the microwelding of asperities. Such welds fracture in shear at the original interface or below it depending on the relative magnitude of the interfacial and cohesive bond strengths. When the adhesive strength exceeds the cohesive bond strength, wear particles may form after several cycles of contact. The wear volume (V) removed in this process is given by

$$V = \frac{K_w PL}{3H}. \quad (9-15)$$

In this well-known expression due to Archard (10), P is the normal operative force, L is the length of relative surface displacement, H is the indentation hardness of the softer body, and K_w is a dimensionless constant whose magnitude depends on the materials in contact. Typical values for K_w are 32×10^{-3} for Cu-Cu pairs, 12×10^{-3} for Ag-Ag pairs, and 1.5 for Cu on steel.

b. Abrasive wear In abrasive wear, hard surface asperities move against a softer material plowing it. Often particles enter the interface, and through cutting or grinding action the softer contact material is plowed or removed. The Archard expression describes the abrasive wear volume as well, but K_w may assume different values depending on whether 2-body (contacting surfaces) or 3-body (contacting surfaces plus abrasive particles) are involved. Wear in this case depends strongly on the ratio $R_w = H_{\text{abraded}}/H_{\text{abrasive}}$, where H is hardness of the indicated material. For $R_w < 1$, extensive wear occurs, but when R_w is slightly greater than unity, wear effectively ceases.

c. Delamination wear Due to the presence of cyclic loading, cracks can develop and propagate parallel to the surface. For sufficiently high tangential frictional stresses, thin platelike particles delaminate from the wearing surface.

d. Corrosive wear Continuous removal of surface films and chemical reaction between these wear products and the environment results in oxidation wear. Oxide films are generally brittle and protective when less than 10 μm thick. Under cyclic stressing the oxide may fracture, causing modest damage if the oxide particles are soft. Hard oxide particles, however, are abrasive and accelerate wear.

e. Fretting wear This important manifestation of contact wear is defined as accelerated surface damage occurring at the interface between contacting materials subjected to small oscillatory movements. Fretting wear is often associated with adhesive, abrasive, and corrosive wear mechanisms. Because we are often concerned with the number of displacement cycles sustained prior to contact failure, low amplitude fatigue is one way to describe fretting.

9.3.5 Fretting Wear Phenomena

The micromotions that cause fretting in electronic connectors range from a few μm to as much as $100\ \mu\text{m}$ in length. They stem from mechanical vibrations, differential thermal expansion, and load relaxation of the contacting metals, as well as junction heating effects due to switching the power on and off. Manifestations of fretting damage include metal debris particles, wear scars, corrosion products, cracked surface oxide or sulfide films, and polymer formation from atmospheric organic precursors.

All metals (i.e., gold, palladium, tin, aluminum, nickel, silver, and copper) used for contacts and connectors suffer differing degrees of fretting damage depending on the particular metals involved, electrical power level, contact forces and frequency of loading, surface roughness, presence of lubricants, temperature, and environment. As Fig. 9-6a shows, fretting damage begins with cracking of surface films as a result of connector micromotions. Oxide regrows on the exposed metal but is in turn cracked with continued rubbing. Due to the accumulated damage during fretting, the contact resistance rises, as shown schematically in Fig. 9-6b. Following initial film cracking, a relatively long period of stable contact resistance ensues. Erratic resistance fluctuations are a precursor to the eventual failure, which is invariably manifested by a discontinuous resistance rise and open contact.

Tests to evaluate fretting damage usually involve measurement of contact resistance changes under the wear track as a function of the number of loading (or displacement) cycles applied by mechanical, electromagnetic, or thermal means. Accelerated current cycling tests have also been performed. The results of extensive testing has yielded a reasonably coherent qualitative picture of the material behavior and wear phenomena observed. Some of this information (11–13) is summarized below:

1. Base metal (Sn, Sn–Pb solder) contacts oxidize rapidly and produce debris that accumulates in the contact zone.
2. Composition changes occur during fretting with dissimilar metals due to mass transfer, wear, and film formation. In general, the direction of metal transfer is from the softer to harder surface. Thus a Au–Pd contact maintains

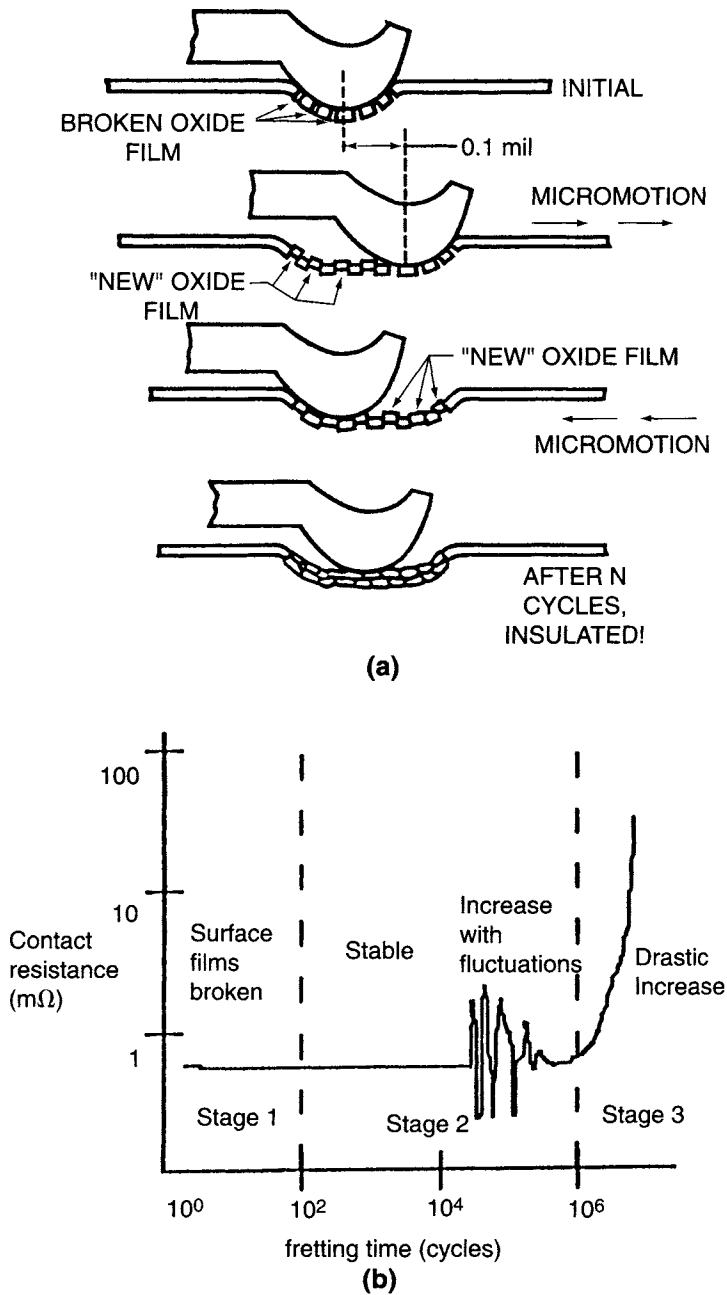


Fig. 9-6 (a) Model of fretting corrosion damage in a contact undergoing micromotions. From reference 6. (b) Stages in the rise of contact resistance during fretting. From reference 14.

low resistance because the harder palladium becomes coated with gold. Conversely, Au–Sn contacts degrade as the softer tin coats the gold, effectively leaving a base metal contact behind. Intermetallic compound formation (e.g., Al–Cu) has been observed to occur at low temperatures.

3. Simultaneous corrosion virtually always accelerates fretting damage to contacts. The reduction in contact area causes either electrical or mechanical failure or a combination of both effects. Through the use of noble metals, corrosion effects are minimized. However, corrosion of the underlying base metal through pores in gold-plated contacts has been observed.
4. Metals such as Pd, Rh, and Ru catalyze the formation of frictional polymers from adsorbed organic air pollutants. Unlike fretting corrosion, the surfaces experience little or no wear.
5. Aluminum connectors in contact with Cu–Ni and Cu–Sn alloys are prone to wear by abrasion, adhesion, and delamination during fretting. Material transfer in both directions occurs.
6. Metals like Au, Ag, and Au–Pd–Ag alloys generally suffer little fretting damage and form little polymer.
7. Lubricants (contact aid compounds) can retard the rate of fretting and stabilize the contact resistance. Commercial lubricants are available to alleviate fretting problems by dispersing polymer buildup, reducing wear and shielding surfaces from oxidation. Paradoxically, lubricants do not have to be conducting; their function is to protect the contact zone from the environment, not conduct electricity.

9.3.6 Modeling Fretting Corrosion Damage

In this section a portion of a recent quantitative model for fretting damage developed by Bryant (14) is presented. The model assumes that the contact surface consists of a large number of hills and valleys. Initially clean asperities are exposed to corrosive attack, and films are formed. After fretting begins, valleys of the lower contact begin to fill with wear debris. Upon reversal of motion, these corrosion products are sheared off and fall into the valleys. When there is enough of an accumulation, the metal surfaces are separated by this nonconductive debris, and the contact resistance rises until connector failure occurs.

To calculate a failure time, we assume that during the n th fretting cycle, a corrosive layer of thickness d_c forms over the exposed valley area A_{ex} , producing a volume of corrosion product V_c equal to $d_c A_{ex}$. The volume of reactant metal that generated V_c is equal to $1/\gamma V_c$, where γ is the ratio of the volume of the corrosion compound to volume of metal. For copper, aluminum, and chromium, $\gamma > 1$. Therefore, the asperity volume lost per cycle is $V_{asp} = V_c - \frac{V_c}{\gamma} = \left[\frac{\gamma-1}{\gamma}\right]V_c$. When the

total asperity volume decrease is equal to a critical valley void volume V^* , failure occurs, or

$$V^* = \sum_{n=1}^N \left[\frac{\gamma - 1}{\gamma} \right] V_c(n) = \sum_{n=1}^N \frac{\gamma - 1}{\gamma} d_c A_{ex}(n). \quad (9-16)$$

For average values over fretting cycles, this equation can be rearranged to yield the number of cycles (N_f) to failure

$$N_f = \frac{V^*}{\left(\frac{\gamma - 1}{\gamma} \right) d_c A_{ex}}. \quad (9-17)$$

In typical connectors, N_f is predicted to range between 10^3 to 10^6 cycles. The normalized contact-resistance increase, as a function of corrosion debris volume, was theoretically shown to be constant until it reaches a value of 0.22 times the volume of the plastically deformed layer. A rapid resistance rise then signals failure in a behavior that parallels that of Fig. 9-6b.

9.4 CREEP AND FATIGUE OF SOLDER

9.4.1 Introduction

Rather than have contacts that open and close, it is often necessary to have permanent interconnections. As we have seen in Chapter 8, solders, perhaps the most ubiquitous of packaging materials, are almost universally employed for this purpose. While our previous concern was primarily with solder processing defects, we now focus on the reliability implications of soldered connections. For this purpose it is worthwhile reviewing Sect. 8.3, dealing with the metallurgical change in phase microstructure and composition as well as intermetallic-compound growth in solders exposed to elevated temperatures. Troublesome thermal stresses, treated in Sect. 8-5, also arise to promote a variety of microscopic structural changes within the context of much broader macroscopic plastic-flow effects.

In practice, thermal cycling of electrical hardware generates corresponding stresses that couple both creep and fatigue effects to produce dynamic strains that often have important consequences for packaging reliability. The resulting deformation will generally vary in a complex way depending on the frequency and amplitude of the thermal stress cycle. Once creep and fatigue are individu-

ally well understood, thermal-cycling phenomena, which combine differing proportions of creep and fatigue deformation, will be more intelligible. That is why this section is primarily reserved for separate descriptions of creep and fatigue phenomena in solders.

In discussing stress relaxation and fretting damage in contacts, we have already treated manifestations of creep and fatigue previously in this chapter.

9.4.2 Creep—An Overview

When loads are applied to materials at elevated temperatures, they creep or plastically strain, provided that both the stresses and temperatures are sufficiently high. In crystalline solids, creep produces a number of irreversible deformation effects, e.g., dislocation migration, and annealing, viscous grain-boundary sliding, a reduction in residual stress, void formation, softening, and mechanical relaxation. Creep in metals and ceramics generally means stress levels near the yield point and temperatures of approximately $0.5 T_M$ and above, where T_M is the melting point in degrees K. Electrical solder (60 Sn–40Pb) melts at $\sim 458\text{K}$ (185°C), and it is easy to appreciate its potential for creep even at room temperature, because $298\text{K}/458\text{K} = 0.65$. Metal lead wires and tungsten filaments of incandescent bulbs operate at very high temperatures and are therefore strengthened to resist creep. Ceramic substrates that typically melt above 2000°C will obviously not creep when used in electronic packaging. At the other extreme, polymers commonly creep in a viscous fashion at low temperatures, i.e., between the glass transition and melting points. Even the tiny Si seed crystal that supports the hot, heavy single crystal being pulled during Czochralski growth does not appear to creep.

Creep is usefully viewed as an admixture of two simultaneous effects; one is strain hardening, and the other is softening due to recrystallization and aging effects. Three regimes of creep deformation can be distinguished, depending on the magnitudes of the hardening and softening rates. When a tensile or shear load is applied under creep conditions, the material instantaneously stretches or strains the predictable elastic amount, after which the strain increases with time at an ever decreasing rate. In the *primary creep* stage the hardening rate exceeds that for softening. This primary-deformation stage then gives way imperceptibly to a regime of *steady-state*, or *secondary, creep*, where the hardening rate is balanced by the softening rate. As a result, the overall creep strain rate (de/dt) is constant. Design of operating conditions for creep-prone components is usually performed in this regime because time-dependent deformation can be confidently projected into the future. Steady-state creep phenomena also pose reliability concerns because potentially large plastic strains can develop. Finally, in

the *tertiary creep* stage the strain increases rapidly with time, leading to failures characterized by much stretching and thinning, void cavitation at grain boundaries, and intergranular cracking. In our subsequent discussion of solder creep we will be exclusively concerned with these last two deformation regimes.

9.4.3 Constitutive Equations of Creep

In the broadest sense, the creep strain (ϵ) that develops is a function of time (t) as well as applied stress (σ) and temperature (T). Often, shear stresses (τ) and strains (γ) are involved, and we shall simply represent these by the generic σ and ϵ for simplicity. Furthermore, ϵ is assumed to be a product of three separate factors, one of which is the ubiquitous Boltzmann factor; therefore, the constitutive equations for the creep strain and strain rates assume the respective forms

$$\epsilon = f(\sigma) g(t) \exp \left[\frac{-E_c}{RT} \right] \quad (9-18a)$$

and

$$\frac{d\epsilon}{dt} = \dot{\epsilon} = f(\sigma) \left[\frac{d g(t)}{dt} \right] \exp \left[\frac{-E_c}{RT} \right]. \quad (9-18b)$$

Explicit expressions for the $f(\sigma)$ and $g(t)$ factors, and the researchers associated with them, are (15)

$f(\sigma) = A \sigma^n$	Norton	$g(t) = (1 + et^{1/3})$	Andrade (primary creep)
$= B \sinh(b \sigma)$	Prandtl	$= t$	Steady-state (secondary creep)
$= C \exp(c \sigma)$	Dorn	$= t^m$	Bailey (tertiary creep)
$= D \sinh(f \sigma)^n$	Garafalo,		

where A , B , C , D , b , c , e , f , n , and m are constants, E_c is the creep activation energy, and RT has the usual meaning. For crystalline solids, E_c is close in magnitude to a diffusional activation energy because creep cannot occur without simultaneous atomic motion. Depending on whether deformation primarily occurs within grains or at grain boundaries, activation energies applicable to these mechanisms would be appropriate. As we shall see, the stress exponent n typically ranges from 1 to 8, depending on the material, stress/temperature range, and creep mechanism.

In each creep regime an equation for both ϵ and $d\epsilon/dt$ can now be constructed by selecting the appropriate forms of (σ) and $g(t)$. For short deformation times

where primary creep occurs, the strain rate decreases with time. Therefore, such behavior can be described by the Andrade form of $g(t)$, where $d\epsilon/dt \approx t^{-2/3}$. Shortly prior to failure, in the tertiary creep regime, the reverse behavior is operative, and the strain rate increases with time. The use of the Bailey time dependence is consistent with deformation in this range because $d\epsilon/dt \approx t^{m-1}$, and $m > 1$.

Between these extremes in behavior, the extended steady-state creep regime for many materials is very commonly described by the phenomenological equation

$$\frac{d\epsilon}{dt} = A \sigma^n \exp \left[-\frac{E_c}{RT} \right]. \quad (9-19)$$

This expression generally holds up to intermediate stress levels ($\sigma/G < 10^{-3}$, where G is the shear modulus) at temperatures above $0.5 T_M$. Among the more significant terms absorbed within constant A is $(1/d)^p$, where d is the grain size and p is a number dependent on the type of creep. So-called diffusional creep prevails at high temperatures and low stress levels ($\sigma/G < 10^{-4}$) such that $n = 1$. In such a case creep can be understood in terms of the Nernst–Einstein model (Eq. 1-6), with stress the driving force and velocity the strain response. Two categories of diffusional creep have been distinguished. Herring–Nabarro diffusional creep, with $p = 2$, occurs at the higher temperatures, where lattice diffusion occurs. On the other hand, Coble creep, with $p = 3$, occurs at lower temperatures, where grain-boundary diffusion dominates.

At higher stress levels, power-law creep due to diffusion-aided dislocation motion is operative, and values of $n = 3$ to 7 have been predicted. However, as we shall see in the next section, Eq. 9-19 does not fully cover steady-state creep phenomena in solders at all stress levels and temperatures.

9.4.4 Solder Creep

In response to thermal cycling of electronic equipment, stresses develop that cause creep deformation in solder interconnections. Two books edited by J. H. Lau (16, 17) are devoted to this subject; they are highly recommended for their broad review of all aspects of thermal stress and creep in solders, as well as the reliability implications in electronic packaging. In addition, much creep test data for solder alloys used in packaging has accumulated in the literature (18–20). Creep behavior is most commonly plotted as creep strain versus time, and a sample curve is shown in Fig. 9-7. Data points for the three creep regimes lie below (primary), above (tertiary), or on the straight line (secondary).

In an extensive study (18) of the behavior of five different solder compositions, creep testing was performed on an array of flip-chip solder ball connections in both

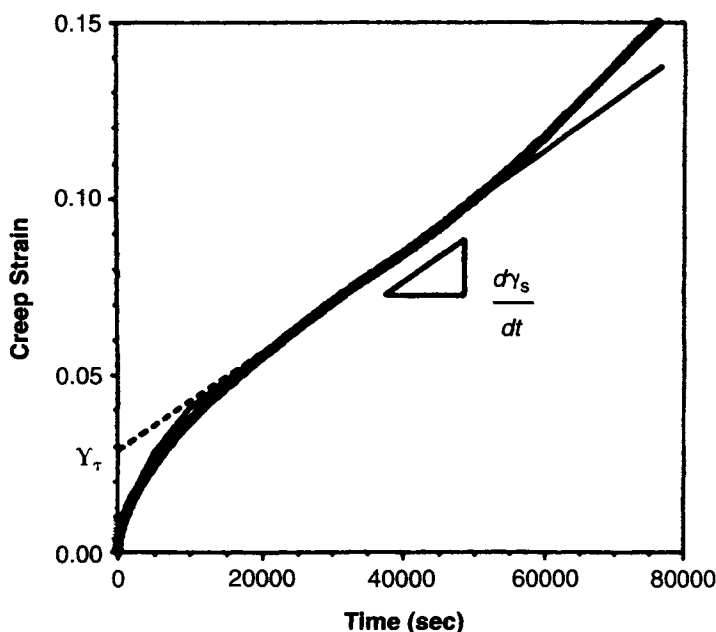


Fig. 9-7 Creep shear strain as a function of time in a constant load creep test. The solder containing 62Sn-36Pb-2Ag was tested at 28°C with a shear stress of 1575lbs/in². From reference 18.

shear (loading parallel to the chip-substrate plane) and tension (loading perpendicular to the chip-substrate plane). While the power law fits data spanning some five orders of magnitude of strain deformation, it breaks down at stress levels of σ/G higher than 10^{-3} . Therefore, rather than Eq. 9-19, the experimental results in this intermediate stress range were analyzed by the Garafalo dependence in the form

$$\frac{d\epsilon}{dt} = C \left(\frac{G}{T} \right) \sinh \left(\frac{a\sigma}{G} \right)^n \exp \left[-\frac{E_c}{RT} \right], \quad (9-20)$$

where C , a , and n are constants. An experimental test of this equation in dimensionless shear strain-rate-shear stress units is shown in Fig. 9-8, demonstrating a good fit over the entire stress range for 60Sn-40Pb solder.

Accurately predicting the extent of solder deformation is a matter of precisely determining its creep characteristics. Experimentally measured values for the constants of a number of solder alloys are entered in Table 9-1. Theory has shown that stress exponents (n) of ~ 3 imply viscous dislocation glide, while n values of $\sim 5-7$ are associated with vertical climb of dislocations past barriers. Later we will consider the physical implications of these exponents on stress

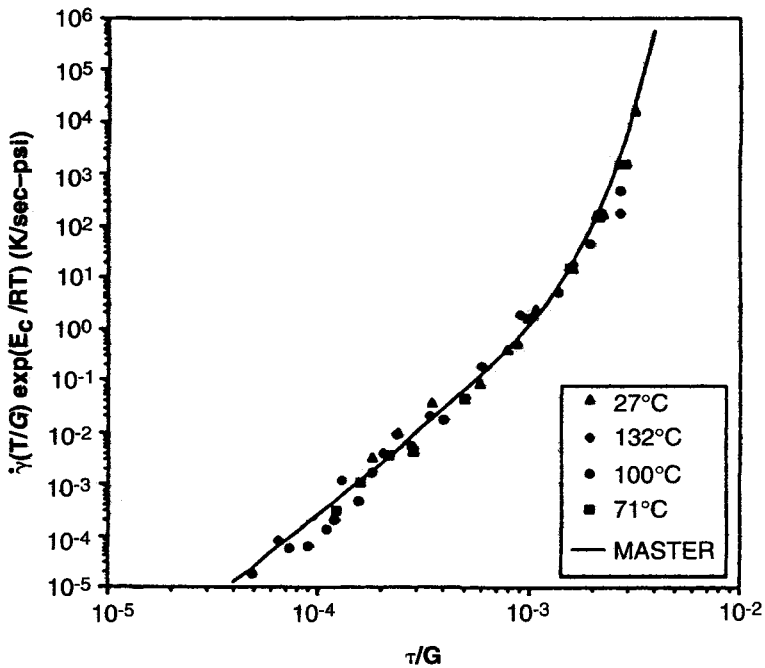


Fig. 9-8 Steady-state creep behavior of 60Sn-40Pb solder plotted as $\frac{\dot{\gamma}(T)}{a(G)} \exp\left[\frac{E_c}{RT}\right]$ vs τ/G . When the data are fit to Eq. 9-20, i.e., $\dot{\gamma} = C(T/G) [\sinh(a \tau/G)]^n \exp[-E_c/RT]$, values of constants derived from the plot are $C = 0.198$ K/psi-s, $a = 1300$, $n = 3.3$ and $E_c = 0.548$ eV. From reference 18.

relaxation of solders during thermal cycling. The activation-energy values for these alloys generally support stress-enhanced diffusional or creeping motion of atoms through grain boundary paths.

In general, homologous-temperature arguments suggest that lower melting temperature solder alloys deform more readily at a given stress than those with higher melting points. Thus, if the electronic package environment demanded that solder joints accommodate large amounts of strain, the low melting temperature compositions should be chosen; under these conditions the higher melting point solders may stress the joints by not relaxing sufficiently.

Example 9-2 In flip-chip interconnections using 95Pb-5Sn solder, suppose that stresses encountered in service fluctuate between $\sigma/G = 7 \times 10^{-4}$ and 1×10^{-3} at temperatures between 25°C and 35°C. A measure of contact degradation is taken as the ratio of creep strains at the two stress levels. What is the difference in the predicted strain ratio using Eq. 9-19 as opposed to Eq. 9-20?

Table 9-1 Creep Constants for Solder Alloys

Composition	C (K/psi-s)	a	n	E _c (kcal/mol)
96.5Sn-3.5Ag	4.89 × 10 ⁻⁴	1300	5.5	9.2
62Sn-36Pb-2Ag	98.9 × 10 ⁻³	1300	3.3	12.6
60Sn-40Pb	198 × 10 ⁻³	1300	3.3	12.6
97Pb-3Sn	5.04 × 10 ⁷	1200	7.0	27.7
95Pb-5Sn	1.21 × 10 ⁷	1000	7.0	27.7
96.5Sn-3.5Ag	3.13 × 10 ⁻³	1500	5.5	11.5

From references 17, 18, and 19.

Answer We refer to the high stress/temperature state as 2 and the low stress/temperature state as 1. In the case of Eq. 9-19, noting that $n = 7.0$ and $E_c = 27.7$ kcal /mol,

$$\frac{\epsilon_2}{\epsilon_1} = \frac{A \sigma_2^n \exp[-E_c/RT_2]}{A \sigma_1^n \exp[-E_c/RT_1]} = \frac{(1 \times 10^{-3})^7 \exp[-27,700/(1.99 (308))]}{(7 \times 10^{-4})^7 \exp[-27,700/(1.99 (298))]} = 55.3.$$

In contrast, the use of Eq. 9-20 where $a = 1000$ and $n = 7.0$, yields

$$\frac{\epsilon_2}{\epsilon_1} = \frac{(298) (\sinh 1)^7 \exp[-27,700/(1.99 (308))]}{(308) (\sinh 0.7)^7 \exp[-27,700/(1.99 (298))]} = 9.4.$$

Therefore, depending on which constitutive equation is chosen, a difference by a factor of 5.9 can result in estimating strain acceleration.

9.4.5 Fatigue—An Overview

Failures due to fatigue are the most common and insidious that metals suffer from. After propagating for some time, unnoticed cracks suddenly and without warning cause fracture of components. Although fatigue is popularly associated with the failure of highly-stressed rotating machinery, it also occurs in electrical equipment in the form of damage to moving contacts as well as vibrating and thermally cycled components. The cyclic loading can be sinusoidal about a mean zero stress level, but any other high- or low-frequency quasi-periodic stress distribution can also cause fatigue failure. Some materials, like steel, display an endurance limit, a critical stress level below which failure does not occur irrespective of the number of cycles; aluminum and polymers show no such endurance limit. These as well as other fatigue phenomena are reviewed in standard texts on the mechanical properties of materials (21).

A microscopic model for fatigue damage in crystalline materials starts with cyclic dislocation displacements on slip planes in concert with the applied load.

As stacks of slip planes emerge on the surface, they form intrusions and extrusions of material that become nuclei for incipient cracks. One such favored surface crack begins to open as it extends into the interior, inexorably driven by the external cyclic stresses. In round sections, the smoothly parted crack surfaces grow outward in concentric circles amid the continuous back-and-forth stressing and rubbing action. Ultimately, very little of the original material remains intact to support the load, and rapid fracture ensues due to overloading. Surface microcracks, scratches, hard phases, notches, and other stress concentrators facilitate fatigue-crack formation, especially in corrosive environments.

In examining cyclic loading effects more carefully, it has been observed that metals are metastable and may either harden or soften with time (22). Whether metals cyclically harden or soften is apparently related to whether they respectively exhibit large or low strain hardening effects in a simple tensile test. Another way to predict what a metal will do cyclically is to take the ratio of the ultimate tensile strength (σ_{UTS}) to the 0.2% offset yield stress (σ_o). If $\sigma_{UTS}/\sigma_o > 1.4$, hardening usually occurs, and if $\sigma_{UTS}/\sigma_o < 1.2$, softening is expected; in between, either or mixed behavior occurs. Solders cyclically soften, and depending on composition and cycling conditions the magnitude of the effect can be quite pronounced.

The case of a cyclically softening metal is shown in Fig. 9-9, where the plastic strain amplitude is held fixed during periodic tensile and compressive straining cycles (Fig. 9-9a). In concert, the stress follows the triangular strain cycle, but with a decreasing amplitude (Fig. 9-9b). When time is eliminated from both cycling responses, the resulting pairs of stress-strain variables exhibit hysteresis behavior upon plotting one against the other. The hysteresis loops obtained are shown exaggeratedly in Fig. 9-9c, where interestingly, their shapes vary with the number of cycles. But after the first few percent of the cycles-to-failure, the hysteresis tends to stabilize, and the loop retraces itself. Loading materials between fixed stress-amplitude limits also generates hysteresis curves that vary in shape with time. This time the width of the loop generally widens. Much fatigue testing and service loading is of this kind. Irrespective of the conditions of cyclic stressing, fatigue failures occur when enough cycles accumulate. Some of the stress and strain quantities used in the discussion below are defined within the context of the hysteresis loop displayed in Fig. 9-10. To create this loop, the metal is first loaded elastically, deformed to a total tensile strain of $\Delta\epsilon/2$ (or $\Delta\epsilon_e/2 + \Delta\epsilon_p/2$), and then elastically unloaded. Next, elastic and plastic compression occurs until a total strain of $-\Delta\epsilon/2$ is attained. Finally, elastic unloading followed by tensile stretching to strain $+\Delta\epsilon/2$ completes the cycle.

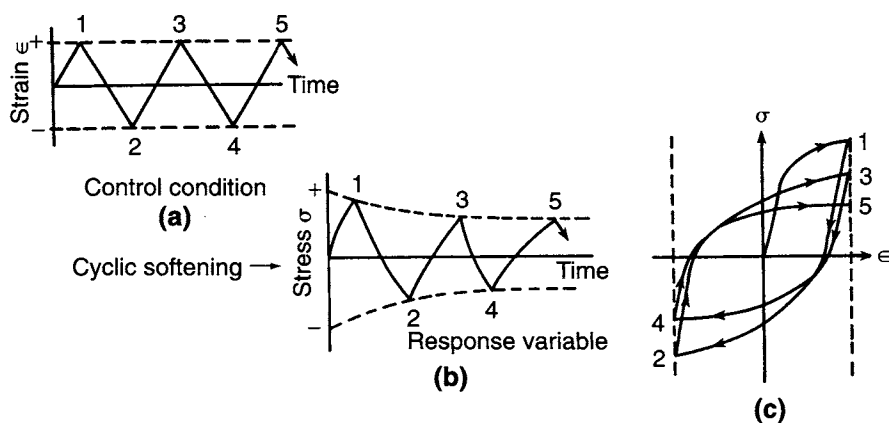


Fig. 9-9 Response of cyclic softening material to cyclic strain cycles. (a) Imposed triangular strain-time cycling. (b) Corresponding stress amplitude as a function of time illustrating cyclic softening. (c) Resultant stress-strain hysteresis loops shown exaggeratedly. From reference 21.

9.4.5.1 Quantifying Fatigue Failure Engineers have traditionally presented fatigue data in the form of S-N curves, where the applied stress amplitude (σ_a) is plotted versus the log of the number of cycles to failure ($\log N$). Large stress amplitudes mean failure in a relatively small number of loading cycles, while fatigue life is extended to many cycles when stress levels are low. Such curves are generally not linear. But semilog plots can often be straightened when the data are plotted in log-log fashion. Thus Basquin demonstrated that the $\ln \sigma_a$ - $\log N$ plot was linear and suggested the exponential law of fatigue. This is normally written in the form

$$\sigma_a = \sigma_f (2N_f)^b, \quad (9-21)$$

where N_f is the number of stress reversals (1 reversal = $\frac{1}{2}$ cycle), σ_f is the fatigue strength coefficient, and b is Basquin's exponent. Typically, σ_f is the true fracture stress, and b varies between -0.05 and -0.12 for engineering materials.

More than a half century later, in 1955, Coffin and Manson suggested that a log-log plot of plastic strain ($\Delta \epsilon_p$) extension in fatigue versus $2N_f$ was also linear. Again, plastic strain-life data are related by a power law, i.e.,

$$\frac{\Delta \epsilon_p}{2} = \epsilon_f (2N_f)^c, \quad (9-22)$$

where ϵ_f is the fatigue ductility coefficient, a constant, and c is the fatigue ductility exponent. Typically, ϵ_f is the true strain at fracture, and c varies between

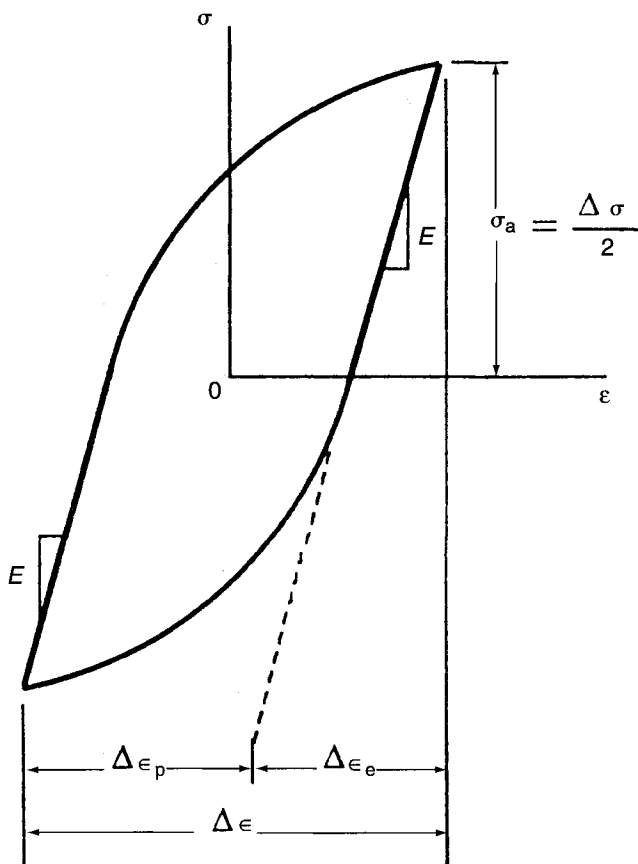


Fig. 9-10 Stress and strain hysteresis loop illustrating elastic and plastic strain contributions.

−0.5 and −0.7. Considerable reference to the Coffin–Manson equation will be made subsequently.

During fatigue the total strain ($\Delta\epsilon$) that develops is the sum of elastic ($\Delta\epsilon_e$) and plastic ($\Delta\epsilon_p$) contributions, so that $\Delta\epsilon = \Delta\epsilon_e + \Delta\epsilon_p$. By assuming that $\epsilon_e = \sigma_a/E$, where E is Young's modulus, we obtain

$$\frac{\Delta\epsilon}{2} = \frac{\sigma_f}{2E}(2N_f)^b + \epsilon_f(2N_f)^c. \quad (9-23)$$

The total strain and its component elastic and plastic contributions are schematically depicted in Fig. 9-11 as a function of the number of stress reversals; data for solder fatigue that obey Eqns. 9-21 and 9-22 are plotted in Fig. 9-12. Note

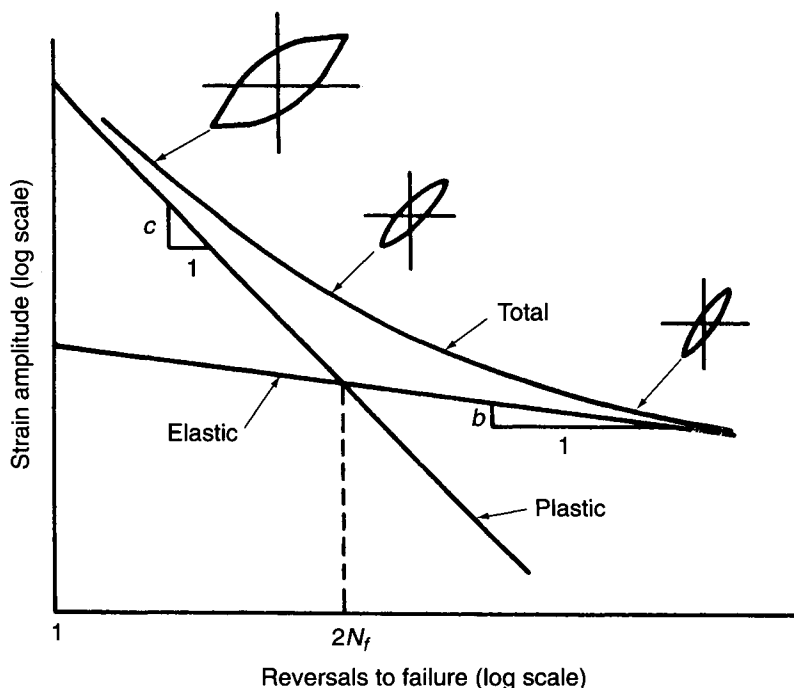


Fig. 9-11 Total strain and individual elastic and plastic contributions depicted as a function of the number of stress reversals. Changes in loop shape and size are schematically indicated. Note that the slopes of linear portions yield b and c .

that wide hysteresis loops are generated when plastic effects predominate, while narrow loops occur when elastic behavior is the rule. Because ϵ_p dominates at short failure times, ductility or toughness controls performance in low-cycle fatigue. However, for long-lived behavior during high-cycle fatigue, performance is controlled by strength.

By recasting Eq. 9-22, we obtain

$$N_f = \frac{1}{2} \left(\frac{\Delta\epsilon_p}{2\epsilon_f} \right)^{1/c}, \quad (9-24)$$

an equation of the form that has been frequently used to predict solder fatigue life-times. With no loss in generality, elastic and plastic shear strains (γ) could be substituted for the tensile strains. This formula expresses the important connection between fatigue failure and the plastic strain that accumulates. Strictly speaking, Eq. 9-23 might be a more suitable criterion for failure, but since elastic strains are small, total and plastic strains are interchangeable for all practical purposes.

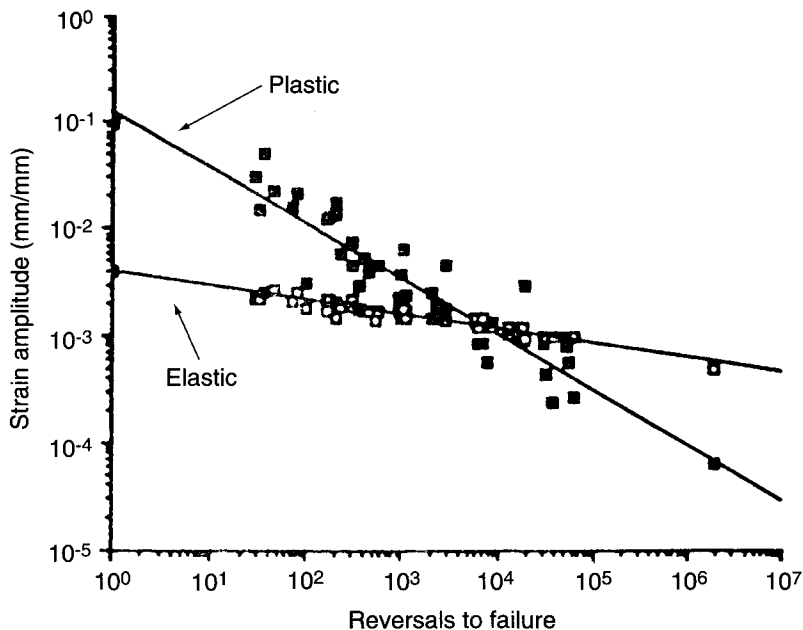


Fig. 9-12 Elastic and plastic strain contributions to solder fatigue versus number of reversals to failure. From N. F. Enke, T. J. Kilinski, S. A. Schroder, and J. R. Lesniak, *IEEE Trans Comp. Hybrids and Manuf. Tech* **12** (4) 459 (1989).

Fatigue does not usually occur at a single stress amplitude but is the result of cumulative damage occurring at a set of different stress or strain levels. For such cases the Palmgren–Miner linear cumulative damage rule is often invoked. This rule states that the fractional amount (R) of failure damage produced in $2N_i$ reversals is expressed by $R = 2N_i/2N_f$, where $2N_f$ represents the number of reversals to failure at the particular applied stress or strain. Since failure is the cumulative sum of n fractional amounts of damage, we write the Palmgren–Miner rule as

$$1 = \Sigma R = \sum_{i=1}^n \frac{2N_i}{2N_f} \quad (9-25)$$

Example 9-3 The solder of Fig. 9-12 undergoes 10^3 reversals at a strain of 1.3×10^{-3} . How many more cycles can the solder sustain before it fails at the larger strain of 10^{-2} ?

Answer Noting that $2N_f$ at these respective strain levels is $\sim 10^4$ and $\sim 10^2$, Eq. 9-25 yields $1 = 10^3/10^4 + 2N_i/10^2$. Solving, $2N_i = 90$ reversals, or 45 cycles.

Nevertheless, fatigue-life parameters for a number of solders are entered in Table 9-2 for a Coffin–Manson relation of the form $N_f^{-c} \Delta\gamma = C$, where the values of the constants c and C depend on ϕ .

9.4.6.2 Frequency Dependence The effect of cycling frequency (f) on fatigue life is depicted in Fig. 9-14 for 60Sn–40Pb solder. Amid considerable scatter two trends are discernable. At high frequency there is little or no effect; at lower frequencies, fatigue life is strongly dependent on frequency. A similar effect has been observed in the fatigue of superalloys. Such findings prompted the Coffin (24) frequency-modified version of the Coffin–Manson equation, i.e.,

$$N_f = \frac{1}{2} f^{1-K} \left(\frac{\Delta\epsilon_p}{2\epsilon_f} \right)^{1/c}, \quad (9-26)$$

where K , a constant, is a measure of the strength of the effect. Thus at 35°C, $K \approx 1$ for frequencies ranging from $\sim 3 \times 10^{-4}$ to 1 Hz, while for f below approximately 3×10^{-4} Hz (or ~ 1 cycle an hour), $K \approx 0$. Furthermore, the frequency where the transition from $K = 1$ to $K = 0$ occurs rises with increasing temperature. It has been common to use a value of K extrapolated from accelerated high-frequency testing to apply under use conditions where much lower frequencies are operative. In such cases fatigue lifetimes are dangerously overestimated.

9.4.6.3 Temperature Dependence It is abundantly documented that for solders, ultimate tensile strengths fall and creep (plastic) strain rates rise with increasing temperatures. In view of this, one expects fatigue life, or the number

Table 9-2 Coffin–Manson Fatigue Parameters for Solders

Composition	$-c$	C	Comments
10Sn–90Pb	0.42	0.93	Air or vacuum, 300K 95% load drop (a)
63Sn–37Pb	$0.24 + 0.10\phi$	$0.23 + 0.78\phi$	300K (b)
42Sn–58Bi	$0.32 + 0.15\phi$	$0.10 + 0.80\phi$	300K (b)
43Sn–43Pb–14Bi	$0.19 + 0.08\phi$	$0.23 + 0.27\phi$	300K (b)
55Sn–35Pb–10In	$0.32 + 0.14\phi$	$0.14 + 2.76\phi$	300K (b)
59.3Sn–39.6Pb–1.1Sb	$0.23 + 0.13\phi$	$0.38 + 1.04\phi$	300K (b)
96Sn–4Ag	$0.72 + 0.08\phi$	$0.48 + 8.06\phi$	300K (b)

(a) Z. Guo, A. F. Sprecher, D. Y. Jung, and H. Conrad, *IEEE Trans. Comp., Hybrids, Manuf. Tech* **14** (4) 833 (1991).

(b) Z. Guo, A. F. Sprecher, H. Conrad, and M. Kim, *Materials Developments in Electronic Packaging Conf. Proc.*, Montreal (1991).

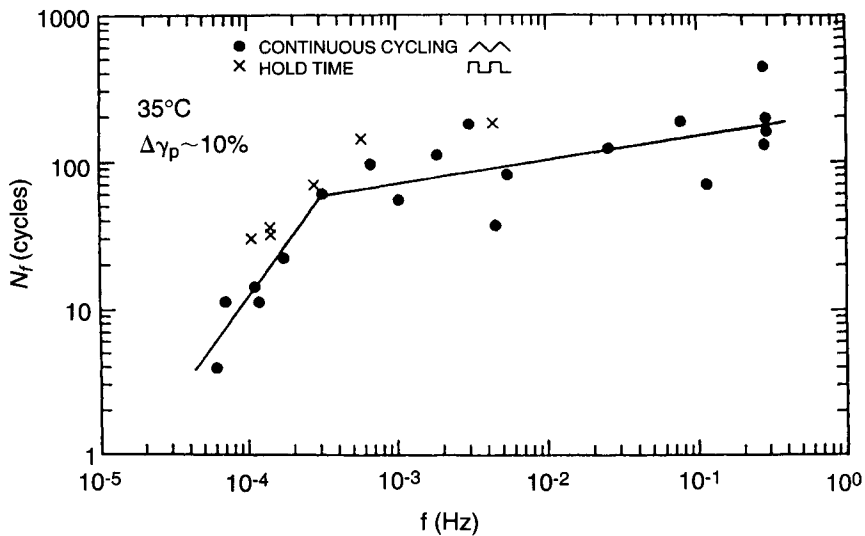


Fig. 9-14 Comparison of 35°C fatigue life (N_f) for 60Sn-40Pb solder versus cycling frequency. From reference 23.

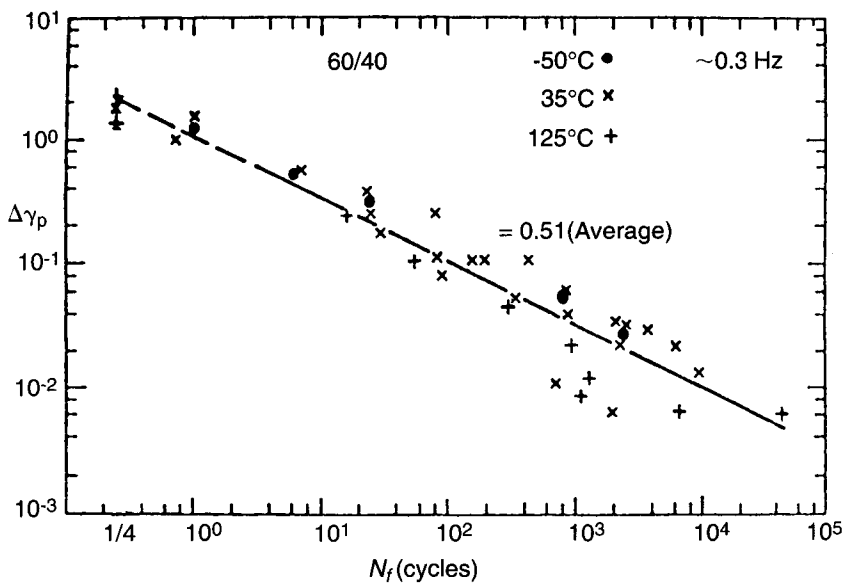


Fig. 9-15 Temperature dependence of low-cycle fatigue life for 60Sn-40Pb solder. A 50% load drop was taken as the failure criterion. From reference 23.

of cycles to failure, to be strongly temperature dependent. Confounding intuition, while there is a temperature dependence, it is not a large one nor, considering the statistical uncertainties in the data for 60Sn–40Pb solder shown in Fig. 9-15, a particularly significant one (23). The same is true of the high-lead solders. Solomon has suggested that claims for larger reported temperature dependencies in high-lead–tin solders may be exaggerated due to scatter in the data.

9.5 RELIABILITY AND FAILURE OF SOLDER JOINTS

9.5.1 Introduction to Thermal Cycling Effects

The previous sections dealing with creep and fatigue assumed that 1. isolated (unattached) solder materials were involved, and 2. deformation occurred under isothermal conditions. Now we address the more complicated issues associated with thermally cycled solder joints attached to IC chips, packages and circuit boards. The objective is to understand the processes that lead to mechanical failure. We recall that the thermal stresses in solders (adhesives) that we calculated at great length in Chapter 8 were elastic shear stresses. However, in this chapter we have already noted that accumulated plastic strain is what causes failure of solder during creep or fatigue. How are elastic shear stresses, which after all produce small elastic strains, related to the much larger plastic strains? In addressing this question and others related to thermal cycling of solder joints, a review of thermal expansion mismatch effects is a good place to begin.

9.5.1.1 Thermal Cycling Distortion of Solder Joints Following Solomon (23), consider a chip or component (c) attached to printed wiring board (PWB) substrate (s) by means of a solder post of height h , all at temperature T_o . The thermal expansion coefficients $\alpha(c)$ and $\alpha(s)$ are such that $\alpha(s) > \alpha(c)$ for the geometry shown in Fig. 9-16a. We note that heating and cooling essentially fatigues the solder joints through the shear strain reversals depicted in Figs. 9-16b and 9-16c, respectively. It is not hard to calculate the strain that develops. First, taking the difference in the expanded lengths on heating to T_{MAX} yields the net displacement of both joints as $L [\alpha(s) - \alpha(c)](T_{MAX} - T_o)$. Similarly, during cooling to T_{MIN} , the net displacement of both joints is $L [\alpha(s) - \alpha(c)](T_{MIN} - T_o)$. The net total thermal displacement (Δ) after heating followed by cooling is the difference in these, or $\Delta = L [\alpha(s) - \alpha(c)](T_{MAX} - T_{MIN})$. By definition, the shear strain (γ) per joint is $\gamma = \Delta/(2h)$, so that

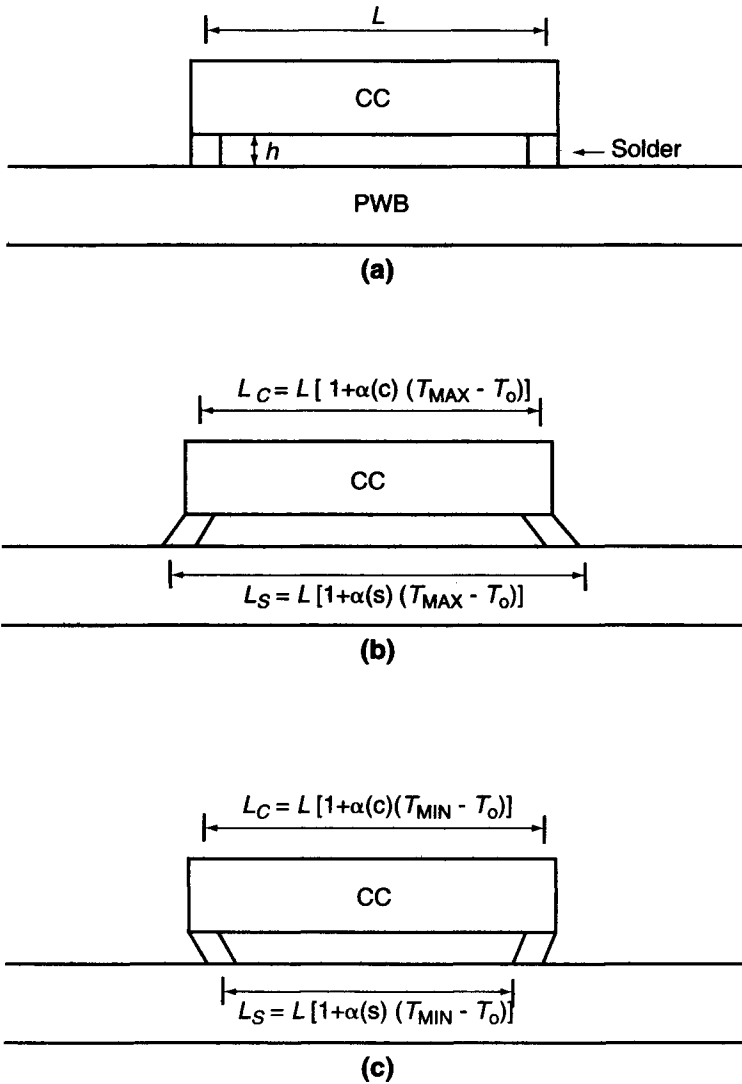


Fig. 9-16 Assorted thermal effects in soldered joint connecting a chip carrier (CC) to a substrate (PWB). Assembly at temperature T_o with thermal expansion coefficients $\alpha(c)$ and $\alpha(s)$, such that $\alpha(s) > \alpha(c)$. (a) Initially unstressed solder joint. (b) Solder joint is heated to T_{MAX} , expanding PWB relative to CC. (c) Solder joint is cooled to T_{MIN} , contracting PWB relative to CC. From reference 23.

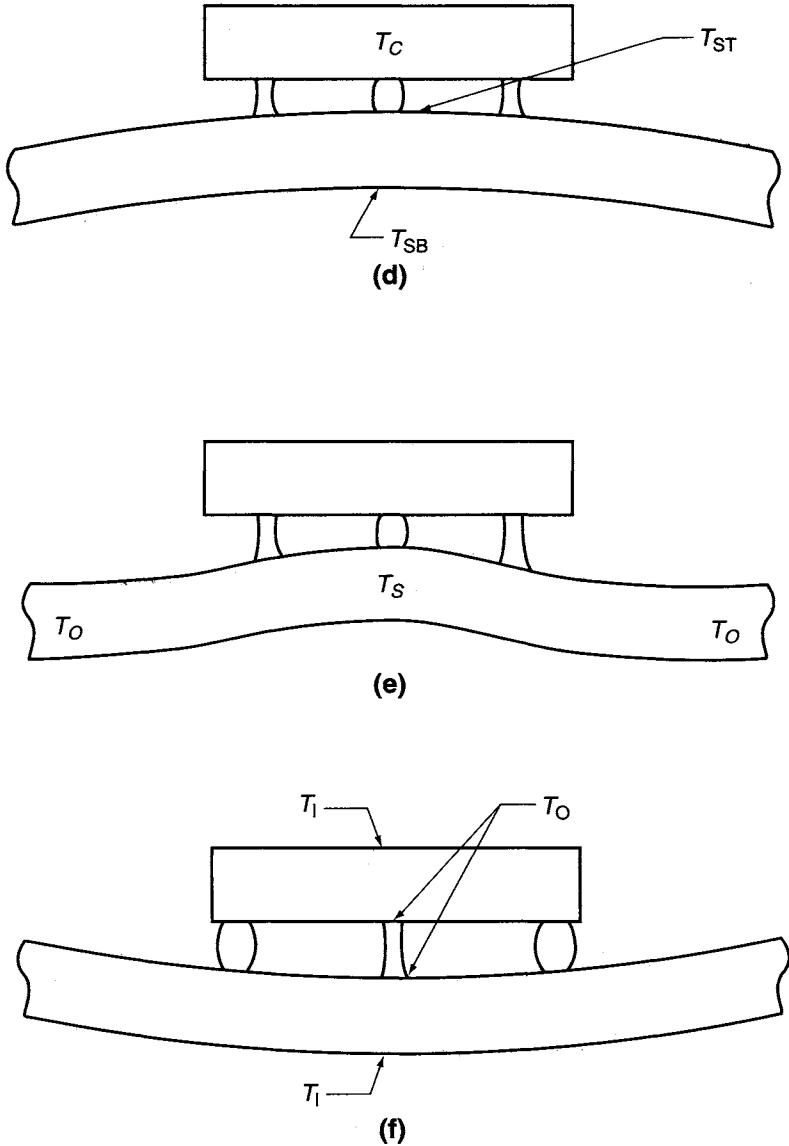


Fig. 9-16 (d) Idealized thermal strain for a through-board temperature gradient where $T_C > T_{ST} > T_{SB}$. (e) Idealized thermal strain for an in-plane temperature gradient where $T_S > T_O$. (f) Deformation of assembly due to thermal shock caused by temperature reversal. From P. M. Hall, in *Solder Joint Reliability*, ed. J. H. Lau, Van Nostrand Reinhold, New York (1991).

$$\gamma = \left(\frac{L}{2h} \right) [\alpha(s) - \alpha(c)](T_{\text{MAX}} - T_{\text{MIN}}). \quad (9-27)$$

9.5.1.2 Conversion of Elastic Strain to Plastic Strain It might be imagined that thermal flexing of the assembly in Fig. 9-16 would yield the same bend angles (positive and negative) for identical temperature excursions above and below the ambient (stress-free) temperature datum. This, after all, is how bimetallic strips respond to temperature. But as seen in Fig. 9-17, this does not happen. Instead of reversible elastic bending, an asymmetrically shaped hysteresis loop develops, indicating that the thermal behavior of the solder is influencing the overall response of the assembly.

This behavior is related to the question raised at the beginning of this section, which is now addressed. First we note that solder joints are elastically strained as well as stressed during thermal cycling. With time at elevated temperature, however, the elastic strain is essentially converted into plastic strain. This effec-

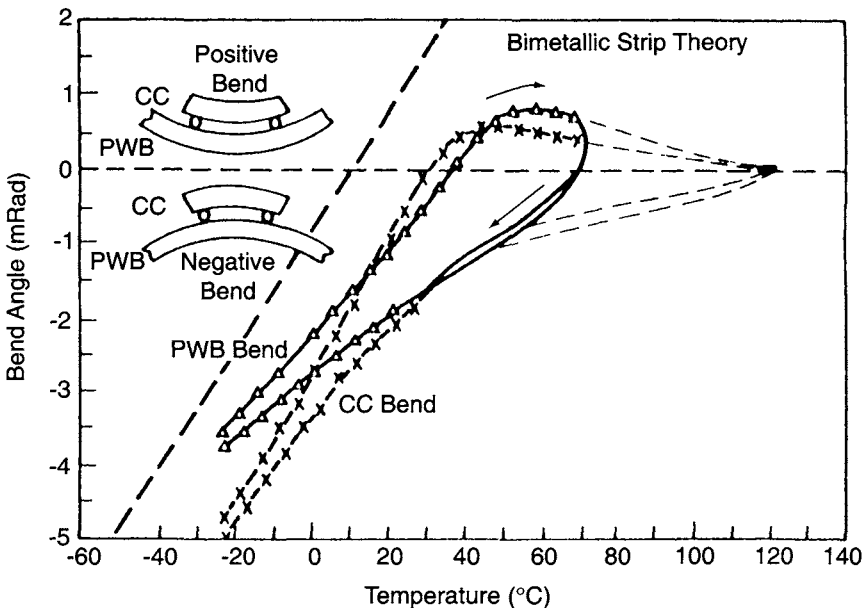


Fig. 9-17 Bend angles as a function of temperature in a soldered plate assembly subjected to reversed bending. Deviation from bimetallic strip theory is indicated. Hysteresis in bend angles (positive and negative) occurs due to temperature excursions above and below the ambient temperature (stress-free) datum. From P. M. Hall, *IEEE Comp. Hybrids and Manuf. Tech.* 7 (4) 314 (1984).

tively limits the bend angle at elevated temperatures. To see why, we model the viscoelastic solder by a spring (of constant k_s) displaying elastic behavior, and by a dashpot (of viscosity η) that embodies its viscous nature. Solder behaves viscously during creep when dislocations and vacancies move or grain boundaries slide. In the classic Maxwell model of a viscoelastic solid (24) consisting of a spring and dashpot in series, consider what happens when this combination is stretched to a total strain (ϵ_o) that remains fixed thereafter. This is like the problem of a bolt initially tightened by force P_o to a high stress level (σ_o) that then relaxes with time through viscous flow. The subsequent force (stress) relaxation behavior as a function of time (t) is described by

$$P(t) = P_o \exp \left[\frac{-k_s t}{\eta} \right], \text{ or } \sigma(t) = \sigma_o \exp \left[-\frac{t}{t_o} \right], \quad (9-28)$$

where the decay time constant t_o is related to the ratio of Young's modulus to the effective viscosity. Again, with no loss in generality, shear stresses (τ) can be substituted for tensile stresses in Eq. 9-28. Since the elastic strain (ϵ_e) is $\sigma(t)/E$, and $\epsilon_o = \epsilon_e + \epsilon_p$, the plastic strain (ϵ_p) is given by

$$\epsilon_p = \epsilon_o - \left(\frac{\sigma_o}{E} \right) \exp \left(-\frac{t}{t_o} \right). \quad (9-29)$$

Thus, with time the elastic strain is irreversibly diminished and replaced by plastic strain. It is this accumulated plastic strain that eventually causes fatigue failure when it reaches a critical level.

9.5.1.3 Thermal Distortion of Nonuniformly Heated Solder Joints

In the cases of solder-joint distortion just treated, the chip, solder, and substrate were all assumed to be at the same temperature whether it was T_{MAX} , T_{MIN} , or T_o . But other scenarios for circuit-board distortion occur in practice. For example, in Fig. 9-16d the chip carrier dissipates heat, so that there is a thermal gradient in the vertical direction. As a result of the hot spot, the top of the PWB is hotter (T_{ST}) than the bottom (T_{SB}), causing it to bow downward as shown. In addition to through-board gradients there may be in-plane thermal gradients generated by neighboring components. This causes similar convex bowing that buckles the PWB in the manner shown in Fig. 9-16e. In both of these cases it is assumed that the solder is totally relaxed, so that the outer solder posts are extended while the inner one is compressed.

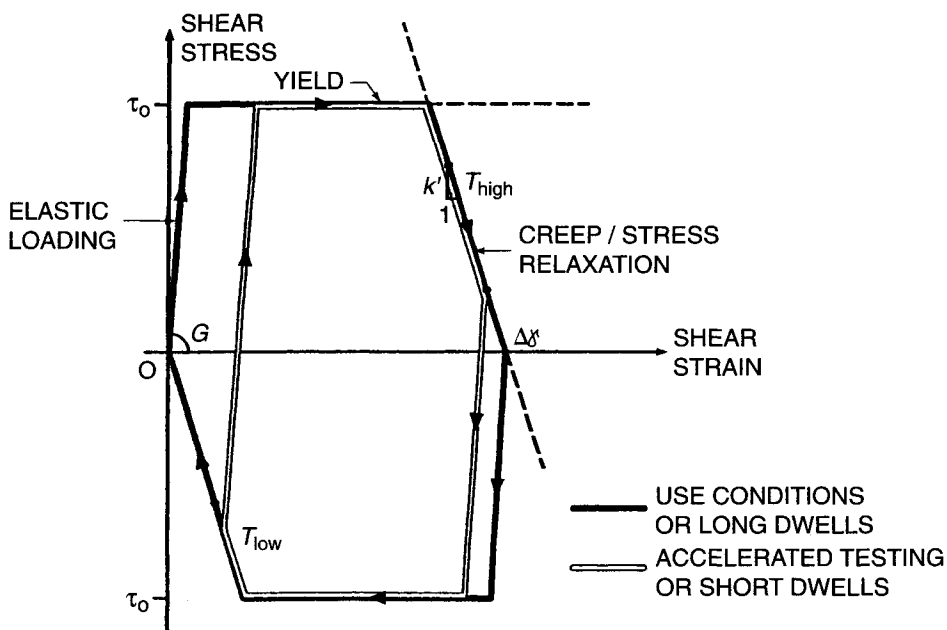
Another environment of interest in packaging is thermal shock. Here the temperature is changed so rapidly that the inner parts (T_o) thermally lag the response of the outer parts (T_i). Therefore, by inverting the temperature distribution, the resulting distortion in Fig. 9-16f is reversed.

9.5.2 Thermal-Stress Cycling of Surface-Mount Solder Joints

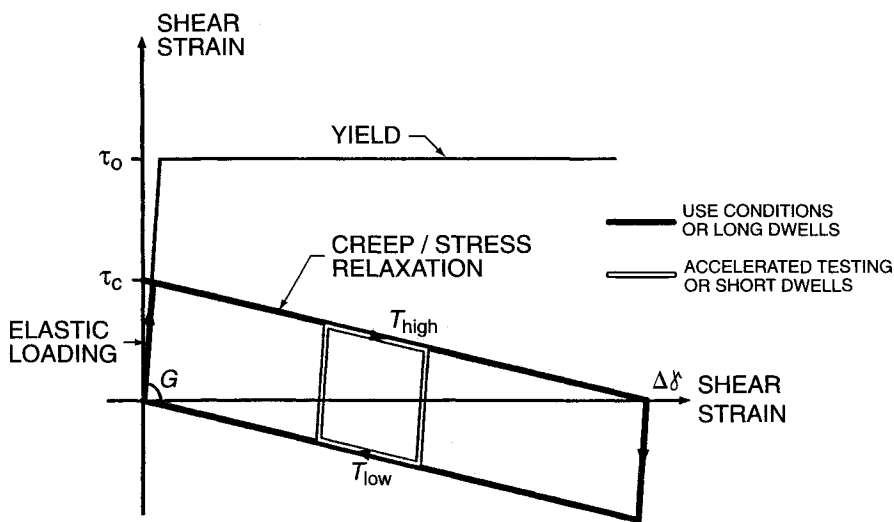
9.5.2.1 Scope We now treat the case of thermal fatigue that occurs during thermal cycling of surface-mounted packages. Merely turning electrical equipment on and off is the simplest way to generate such periodic thermal stressing. The isothermal loading cycle that was considered in Sect. 9.4.6 is clearly a simpler process than the case now being addressed. Engelmaier and coworkers (26–28) have extensively studied the problem of thermal fatigue in surface-mount (SM) solder joints and the related issue of design for long-term reliability. Two types of SM joints were considered—leadless and leaded (see Fig. 8-18). Leadless joints are stiff and sustain high stress levels, but occupy little space on the board. On the other hand, leaded joints are compliant, enabling them to reduce stress levels below the solder yield stress, thus minimizing cyclic fatigue damage as a result. The lower stress levels, in turn, slow and may even arrest harmful viscoplastic stress-relaxation and creep processes. As a result, leaded joints suffer lead fracture, while solder fatigue is the usual failure mechanism in leadless joints. The difference in these two joints is reflected in their response to thermal cycling, as indicated in the hysteresis loops of Fig. 9-18a and b. Simulated hysteresis loops for a number of different thermally cycled solders are plotted in Fig. 9-19 and reveal similar behavior.

9.5.2.2 Leadless Joints In the case of the leadless surface-mount solder joint at high temperature, elastic-plastic, shear stress-strain deformation occurs. Beyond the initial elastic response in Fig. 9-16a, the solder essentially yields at a constant stress level (τ_0), and depending on the ductility, large strains can be induced. With continued exposure to elevated temperature (T_{high}), the solder relaxes along a line dependent on the assembly stiffness. During this time, stress relaxation and creep processes are activated to extend the plastic strain. Given enough time the stress can even relax to zero. These stress and strain effects are reversed in the low-temperature (T_{low}) portion of the thermal cycle. When the initial temperature is again reached, the stresses and strains are ideally restored to zero; but the damage has been done in the form of plastic strain accumulation in the solder. The fact that the hysteresis loop appears symmetric does not mean that equal times were spent at high and low temperatures. In fact, longer times are required at T_{low} than at T_{high} to produce equivalent plastic strains. It should be noted that during accelerated thermal cycling, dwell times are shorter and stress relaxation may be incomplete. This results in the reduced hysteresis curves.

Due to the large thermal stresses in leadless joints, rapid and extensive stress relaxation causes plastic-strain accumulation. In such a case the relation for isothermal cyclic life (Eq. 9-22) is applicable, but it must be modified by the operative temperature–time variables. The equation



(a)



(b)

Fig. 9-18 (a) Shear stress and strain hysteresis response in a stiff, surface-mounted leadless joint. Use (long dwell) as well as accelerated (short dwell) cycles are indicated. (b) Shear stress and strain hysteresis response in a compliant, surface-mounted leaded joint. From reference 26.

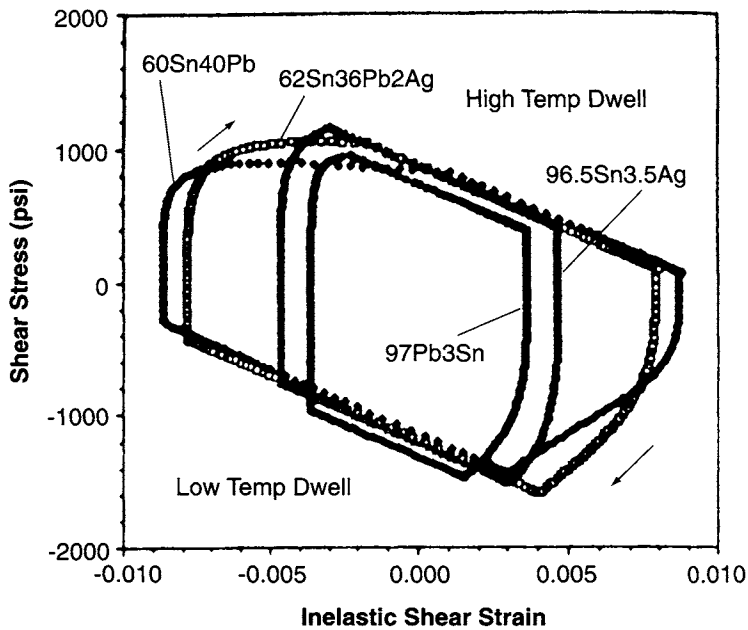


Fig. 9-19 Simulated hysteresis loops for a number of different solders, thermally cycled once each day, plotted as shear stress versus shear strain. The high and low temperature dwells are at 25°C and -75°C, respectively. From reference 18.

$$N_f = \frac{1}{2} \left(\frac{\Delta\gamma}{2\epsilon_f} \right)^{1/c} \quad (9-30)$$

has been used to model thermal fatigue, where for 63Sn-37Pb solder, $2\epsilon_f = 0.65$ and $c = -0.442 - 6 \times 10^{-4}T_m + 1.74 \times 10^{-2} \ln(1 + 360/t_D)$. This complicated form of the fatigue ductility exponent accounts for both the mean cyclic solder-joint temperature (T_m) and half cycle dwell time (t_D in minutes). To use Eq. 9-30, the plastic shear range ($\Delta\gamma$) must be known, and Eq. 9-27 can be used for this purpose. In rectangular packages L is taken as the diagonal length of the chip because solder joints beneath the chip corners suffer the maximum stresses and are most susceptible to fatigue.

Finally, the mean cyclic life of leadless joints can be expressed by

$$N_f = \frac{1}{2} \left\{ L \frac{[\alpha(s) - \alpha(c)](T_{MAX} - T_{MIN})}{2h(2\epsilon_f)} \right\}^{1/c} \quad (9-31)$$

Quantitative agreement between this formula and the fatigue life of ceramic chip carriers of varying size on assorted substrates has been achieved (26).

9.5.2.3 Leaded Joints In an entirely parallel manner we now consider the corresponding issues for leaded surface-mount solder joints, starting with the hysteresis behavior (Fig. 9-18b). Initially the stress-strain response is elastic at elevated temperature. But if the lead is sufficiently compliant, it deflects easily and keeps stress levels below those required to cause yielding in the solder. The stress rises only to τ_c and then relaxes as the solder now creeps along isothermal lines of lesser slope than for the stiffer leadless joints. Given sufficient time, the stress relaxes to zero. Again reversal of stress and strain at lower temperature completes the loop.

Less plastic strain energy, as measured by the loop area, accumulates in leaded joints because the lower stress levels prevent extensive plastic straining. Therefore, Eq. 9-30 is an inappropriate criterion for joint lifetime. Instead, the viscoplastic strain energy (ΔW), proportional to the area of the hysteresis loop, has been proposed as a measure of fatigue damage. Since ΔW effectively substitutes for $\Delta \epsilon_p$, the formula

$$N_f = \frac{1}{2} \left(\frac{\Delta W}{W_j} \right)^{1/c}, \quad (9-32)$$

with W_j a constant, has been used to predict fatigue lifetimes. Basically, the loop area (ΔW) is a product of the largest stress ($\Delta \tau_{\max}$) and $\Delta \gamma$. We have already developed a formula for $\Delta \gamma$ in Eq. 9-27 and will assume that $\Delta \tau_{\max}$ is proportional to $\Delta \gamma$ according to Hooke's law. Thus, $\Delta \tau_{\max} = K_F(L/A) [\alpha(s) - \alpha(c)] (T_{\max} - T_{\min})$, where K_F is the diagonal flexural stiffness (spring constant) of the package and A is the effective solder area. Combining these terms,

$$\Delta W \approx \Delta \tau_{\max} \Delta \gamma \approx \frac{K_F}{Ah} \{L[\alpha(s) - \alpha(c)](T_{\max} - T_{\min})\}^2. \quad (9-33)$$

Through substitution in Eq. 9-32, we finally express the mean cyclic life of leaded joints by

$$N_f = \frac{1}{2} \left[\frac{K_F \{L[\alpha(s) - \alpha(c)](T_{\max} - T_{\min})\}^2}{Ah W_j} \right]^{1/c}. \quad (9-34)$$

Agreement between this formula and cyclic tests for various surface-mount component assemblies has been realized (29).

In summary, strategies used to extend surface-mount solder-joint life essentially follow two directions, namely,

1. tailor the coefficients of thermal expansion (CTE), and
2. increase attachment compliancy.

Tailoring to reduce expansion mismatch shrinks the cyclic hysteresis loop primarily by reducing the strain range. On the other hand, attachment compliancy works by reducing the cyclic stress level.

9.5.3 Acceleration Factor for Solder Fatigue

The simplest way to transform the results of accelerated stress-cycling testing (s) to predict fatigue behavior under use conditions (u) is through a direct ratio of the involved failure times. Therefore, we may write for the acceleration factors,

$$AF = \frac{N_f(u)}{N_f(s)} = \frac{[\Delta\gamma(u)]^{1/c(u)}}{[\Delta\gamma(s)]^{1/c(s)}} \quad (9-35a)$$

and

$$AF = \frac{[\Delta W(u)]^{1/c(u)}}{[\Delta W(s)]^{1/c(s)}} \quad (9-35b)$$

for leadless and leaded joints, respectively. These expressions implicitly account for temperature and cycle duration in the values selected for c .

An early and still popular method for determining AF in solder-joint fatigue processes is due to Norris and Landzberg (30), who proposed the formula

$$AF = \frac{N(u)}{N(s)} = \left[\frac{f(u)}{f(s)} \right]^{1/3} \left(\frac{\Delta T(s)}{\Delta T(u)} \right)^2 \left(\frac{MTTF(u)}{MTTF(s)} \right). \quad (9-36)$$

The symbols N , f , ΔT and MTTF refer to the number of cycles to failure, cycling frequency, temperature range, and isothermal fatigue-test lifetime, respectively. Often ΔT is taken as the maximum temperature in °C. With respect to MTTF(s) the isothermal test is performed at the peak temperature used in accelerated thermal cycling. As suggested by Eq. 9-26, a single value for $K = 0.67$ has been selected here; in view of considerations raised in Section 9.4.6.2, this assumption requires scrutiny.

To end our discussion of acceleration factors it is instructive to present the following example due Lau *et al.* (31).

Example 9-4 Accelerated temperature cycling of solder joints on a 256 pin quad flat package (QFP) was carried out at a frequency of 1 cycle/45 min over the range -40 to 125°C . The Weibull cumulative failure distribution was found to be of the form $F_s(t) = 1 - \exp[-(t/42,900)^{1.27}]$, where t is the time in minutes. This chip will be used in a computer, where it undergoes 1 thermal cycle per day and is exposed to $\Delta T(u) = 85^\circ\text{C}$. If it is assumed that $MTTF(u)(85^\circ\text{C})/MTTF(s)(125^\circ\text{C}) = 1.7$,

- What is the acceleration factor?
- Predict the failure rate under use conditions.
- Derive an expression for $F_u(t)$.

Answer

- a. Substituting in Eq. 9-36, noting that 1 cycle/45 min = 32 cycles/day,

$$AF = (1/32)^{1/3} (165/85)^2 (1.7) = 2.02.$$

- b. From Eq. 4-42 and the definition of Weibull failure rate (Eq. 4-15),

$$\lambda_u(t) = (1/2.02)^{1.27} \times (1.27/42900)(t/42900)^{0.27} \text{ or } 0.0000121(t/42900)^{0.27}.$$

- c. From Eq. 4-39, $F_u(t) = F_s(t/AF)$. Therefore,

$$F_u(t) = 1 - \exp [-(t/2.02 \times 42,900)^{1.27}] = 1 - \exp [-(t/86700)^{1.27}].$$

9.5.4 Fracture of Solder Joints

9.5.4.1 Fracture Mechanics—An Overview Mechanical fracture of semiconductor chips, plastic packages, and ceramic substrates are no doubt the most severe processing defects and reliability concerns we encounter. The subject of fracture mechanics is a very fruitful way to address these failures as well as those of solder joints. By assuming that defects are an unavoidable consequence of manufacturing, fracture mechanics addresses their implications with respect to mechanical failure of the surrounding matrix. The traditional analysis of the energetics at play when a defect or incipient crack of critical size (a) rapidly opens under applied stress σ , yields the quintessential formula of fracture mechanics

$$K_{IC} = \sigma (\pi a)^{1/2}. \quad (9-37)$$

It concisely relates processing defects (a), material properties (K_{IC}) and design stresses (σ) in a neat way. In the application of this equation, the material property known as the fracture toughness (K_{IC}) plays a key role. Essentially, when K_{IC} is greater than $\sigma (\pi a)^{1/2}$ the material is tough enough to withstand the effect of stress concentration about the crack. When both sides of Eq. 9-37 are equal, a precarious mechanical balance is maintained and the mechanically sensitive crack, though on the verge of propagating, is still stable. But when $\sigma(\pi a)^{1/2}$ exceeds K_{IC} , mechanical instability causes the crack to open rapidly.

The critical difference in toughness between metals and other classes of materials stems from the presence of nondirectional bonds. Dislocation motion enables atoms in metals to slide readily by one another, facilitating the plastic deformation that blunts the advance of cracks. No such mechanism exists in the case of other materials. Therefore, metals are tough and have large K_{IC} values. Conversely, ceramics, semiconductors, and polymers are not tough, with K_{IC} values an order of magnitude or more lower than for metals; even small defect cracks are critical and readily lead to brittle fracture.

In solder-contact applications we are concerned not only about the fracture toughness of solders, leads, and substrates, but also with the intermetallic phases

that form at interfaces between them. Values of K_{IC} for some of these materials are listed in Table 9-3. Although they are not easy to measure and subject to wide variations, the K_{IC} values for bulk solders are probably larger than those for the Cu-Sn intermetallic compounds. Values for the latter, in turn, appear to be larger than for other intermetallics, which are barely tougher than typical ceramics. Importantly, it is clear that thicker interfacial compound layers are more fracture-prone than thinner layers.

An interesting way to think about fracture was already presented in Section 1.3.5 dealing with load-strength interference effects. The fracture strength, or toughness, of materials and structures in solder joints is statistically distributed because of the assorted phases and processing defects present, and their respective size variations. Applied loads are also subject to statistical spread because of the their irregular static and dynamic character, and the variety of joint geometries and associated stress concentrations. Initially there is no load-strength interference because the respective Gaussian distributions do not overlap. However, through cyclic thermal stressing, the strength Gaussian distribution of the solder falls and probably broadens while a similar, but perhaps lesser change in loading

Table 9-3 K_{IC} (MPa-m^{1/2}) Values for Solder and Related Materials

Solder	Layer Thickness				
	1 μm	4 μm	8 μm	10 μm	>10 μm
60Sn-40Pb	8 (a)	6 (a)	6 (a)	6 (a)	71(d)
Intermetallics					
Cu ₆ Sn ₅ (b,c)	12, 27	7, 22	6, 19		1.4
Cu ₃ Sn (b)	12	6	6		1.7
Ni ₃ Sn ₄ (c)	21	—	—		
Sn-Ag (a)					1
Sn-Bi (a)		5			
Sn-Ag-Bi (a)	5			2	3
Substrates					
Cu					>100 est.
Ni					>100 est.
Al ₂ O ₃					3.7

From

(a) D. R. Frear, *J. of Materials* 49, May (1996).
(b) R. E. Pratt, E. I. Stromswold, and D. J. Quesnel, *IEEE Trans. Comp. Pack. Manuf. Tech.*, A(19), 134 (1996).
(c) D. R. Frear, F. M. Hosking, and P. T. Vianco, *Proc. Conf. Materials Developments in Microelectronic Packaging*, Montreal. p. 229 Aug.19 (1991).
(d) R. B. Clough, A. J. Shapiro, A. J. Bayba, and G. K. Lucey, *J. Electronic Packaging, Trans ASME* 117, 270 (1995).

occurs. The centroids of the two distributions approach each other, raising the probability of overlap and joint fracture.

9.5.4.2 Crack Propagation Rates Cracks in solders have been reported to nucleate at solder-base metal interfaces and interfacial intermetallic compounds, as well as at highly stressed surface points (32). Once nucleated, crack propagation occurs through the bulk of the solder as well as along the solder-base metal interface. Some examples of solder fractures in through-hole and surface-mount joints are reproduced in Figs. 9-20 and 9-21, respectively. These are taken from a larger collection of beautiful solder structure and failure photographs by de Kluizenaar (33). Irrespective of where it occurs, fatigue-fracture propagation has been modeled to yield the number of cycles to failure. We start by noting that at arbitrary, noncritical stress levels the stress intensity (K) is given by $K = (\pi a)^{1/2} \sigma$ (see Eq. 9-37). The assumption is then made that crack growth depends on some power of cyclic stress intensity range to yield the Paris equation,

$$\frac{da}{dN} = B (\Delta K)^m = B(\pi a)^{m/2} (\Delta \sigma)^m, \quad (9-38)$$

where B and m are constants derived directly from experiment. Physically, the crack growth rate is necessarily proportional to the amplitude of the applied peak-to-peak stress, $\Delta \sigma$, or difference between the maximum and minimum applied stress. Distinctions in this formula may arise depending on whether mode I (tension) or II (shear) cracks are involved. In addition, corrections for cracks of different geometries alter the constants slightly. This differential equation then directly links crack growth to applied stress, material constants, and instantaneous crack size. Below the indicated stress levels for applicability of Eq. 9-38, cracks do not grow appreciably, while above them crack growth is nonlinear.

If the crack (defect) size is initially a_o the number of cycles N_f required to extend it to a final length a_f is easily obtained by first separating variables a and N , and integrating, i.e.,

$$\int_0^{N_f} dN = \int_{a_o}^{a_f} \frac{da}{B[\Delta \sigma(\pi a)^{1/2}]^m}$$

Substitution of limits yields

$$N_f = \frac{2\{a_o^{-(m-2)/2} - a_f^{-(m-2)/2}\}}{(m-2) B \pi^{m/2} (\Delta \sigma)^m}, \quad m \neq 2. \quad (9-39)$$

The sparse amount of data on B and m values makes it difficult to apply this formula for particular solder-joint failure applications. Crack growth-rate data

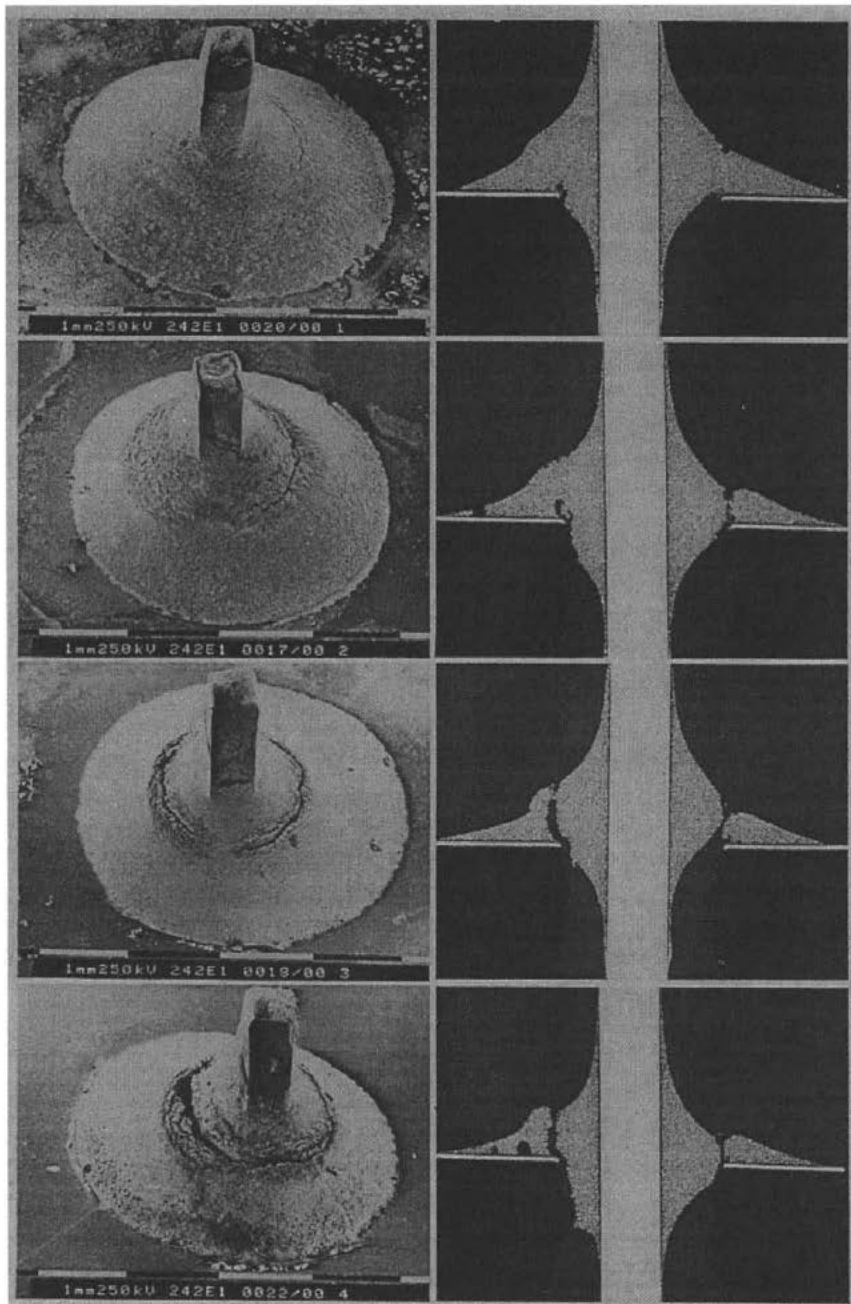
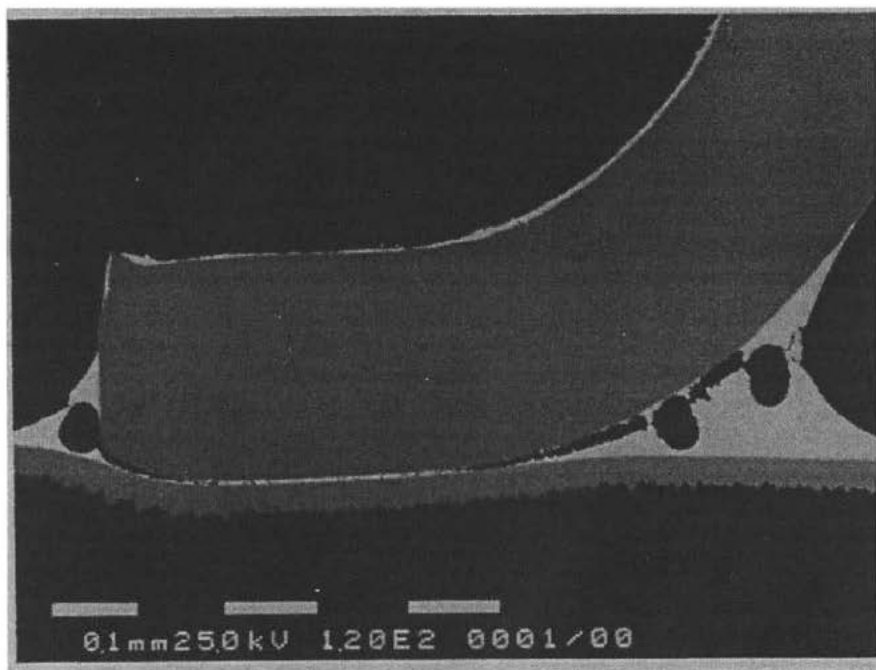
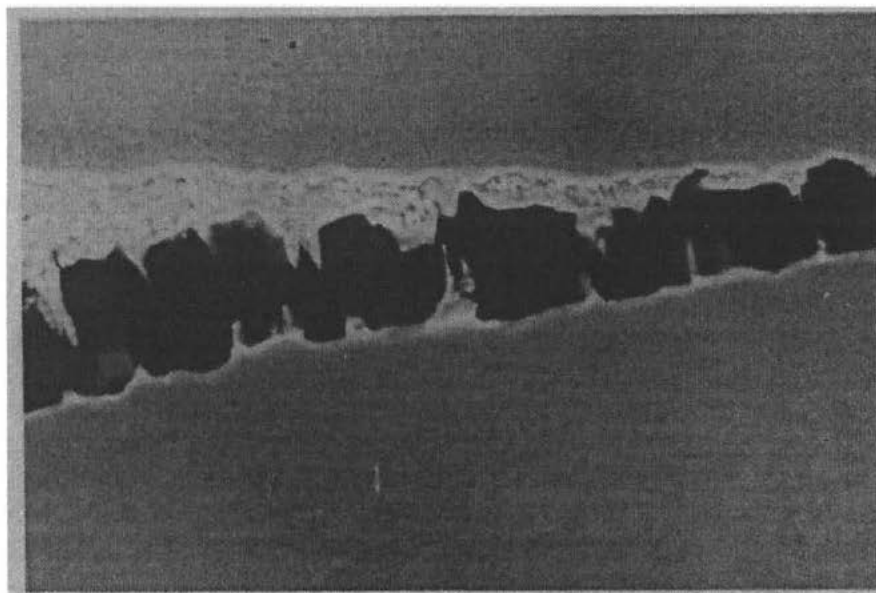


Fig. 9-20 Various stages of fatigue fracture in a through-hole solder joint connecting a lead to a single-sided circuit board. From E.E. de Kluizenaar, *Soldering and Surface Mount Tech* 5, 56, June (1990).

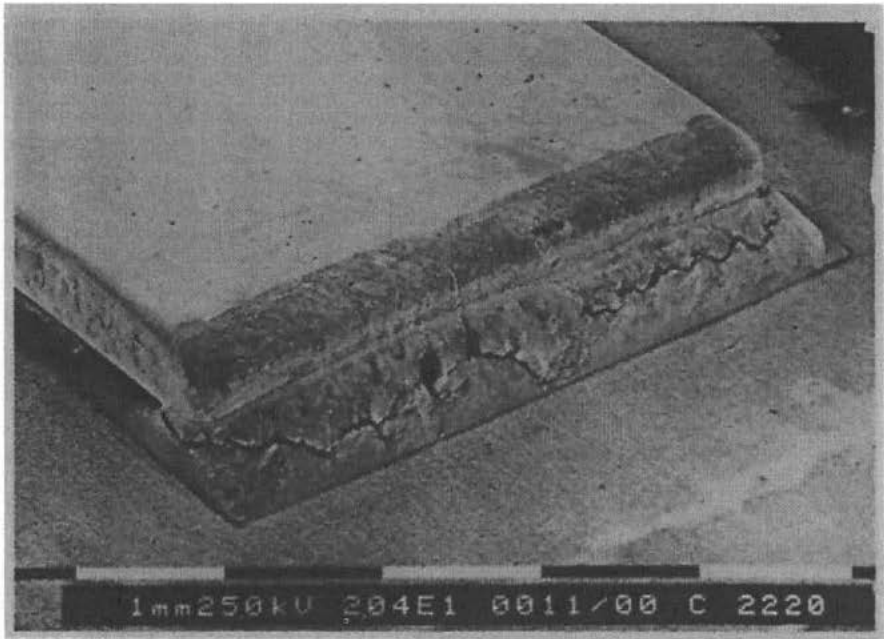


(a)

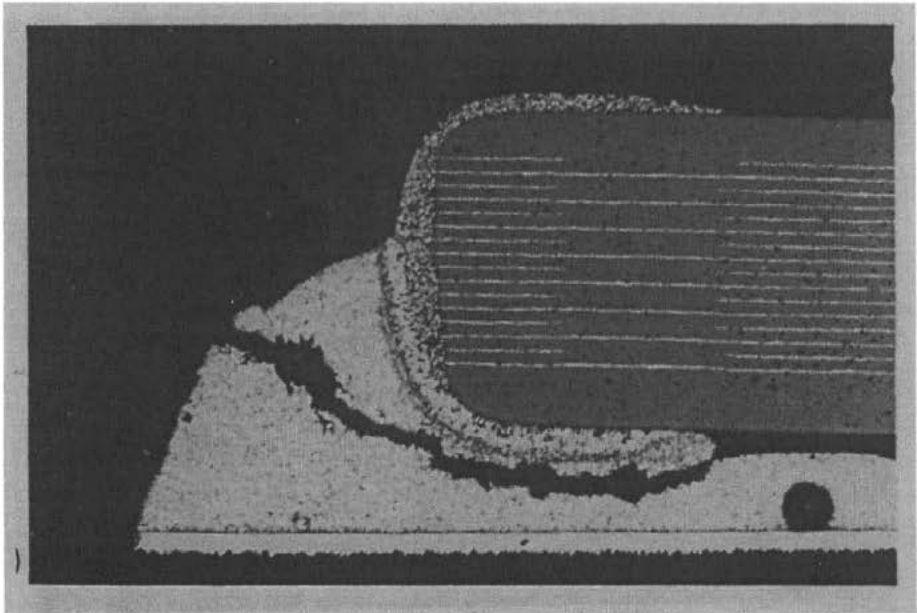


(b)

Fig. 9-21 A collection of surface-mount solder-joint fractures. (a) Porosity at solder-lead interface. (b) Creep rupture of reflowed solder joint with lead containing a large elastic deformation.



(c)



(d)

Fig. 9-21 (c) General view of fatigue fracture of a soldered joint. (d) Cross-sectional view of fatigue fracture shown in Fig. 9-21c. From E. E. de Kluizenaar, *Soldering and Surface Mount Tech* 5, 56, June (1990).

for eutectic Pb–Sn solder are shown in Fig. 9-22, where amid much scatter Eq. 9-38 appears to be followed. From such curves both B and m can be extracted. But the question of what constitutes failure means the use of ϕ . These same authors (34) also claim that eutectic solder is more resistant to crack growth than the 96Sn–4Ag and 42Sn–58Bi compositions. In using Eq. 9-39 to solve practical problems, a_o and a_f must be selected. The latter may be taken as the joint length, while a measure of a_o is the interdendritic spacing within the microstructure, e.g., 0.01 to 0.1 μm .

A second approach to the problem of crack propagation rates in solder joints was proposed by Satoh *et al.* (35). Instead of Eq. 9-38 they suggested the simple equation

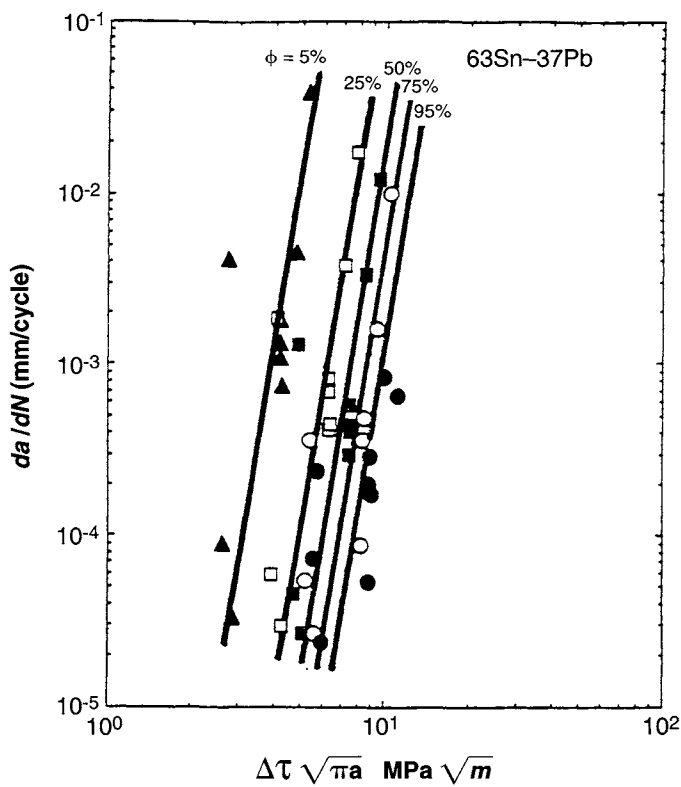


Fig. 9-22 Crack growth-rate data at 300K for eutectic Pb–Sn solder plotted as da/dN versus shear fracture toughness range. Triangular stress-time cycling was performed at a frequency of 0.1 Hz, and data are plotted for different values of ϕ . From reference 34.

$$\frac{da}{dN} = Aa + B, \quad (9-40)$$

where constants A and B are related to crack propagation resistance and crack nucleation, respectively. Integration of Eq. 9-40 gives the fatigue life N_f as

$$N_f = \int_0^{N_f} dN = \int_{a_o}^{a_f} \frac{da}{Aa + B} = A^{-1} \ln \left(\frac{Aa_f + B}{Aa_o + B} \right). \quad (9-41)$$

A fit of this relation to observed fatigue striations in QFP solder joints shown in Fig. 9-23 reveals that $A = 1.48 \times 10^{-3}$ and $B = 0.18$. In addition, three-dimensional finite-element modeling of the plastic strain versus number of fatigue cycles obeyed the Coffin-Manson relationship. The authors suggest a synthesis of all the pertinent parameters, i.e., frequency, temperature, plastic strain, and crack dimensions, that affect thermal fatigue to yield

$$N_f = C \left\{ \ln \left(\frac{Aa_f + B}{Aa_o + B} \right) \right\} (\Delta\epsilon)^{-c} f^m \exp \left[-\frac{E_f}{kT} \right], \quad (9-42)$$

with C a constant and E_f an appropriate activation energy for fatigue. By now we should all realize the appropriateness and influence of the individual factors in this equation.

A parallel, finite-element analysis of crack growth in solder joints tested by Satoh *et al.* (35) has been performed by Lau (36). This time the thermal fatigue

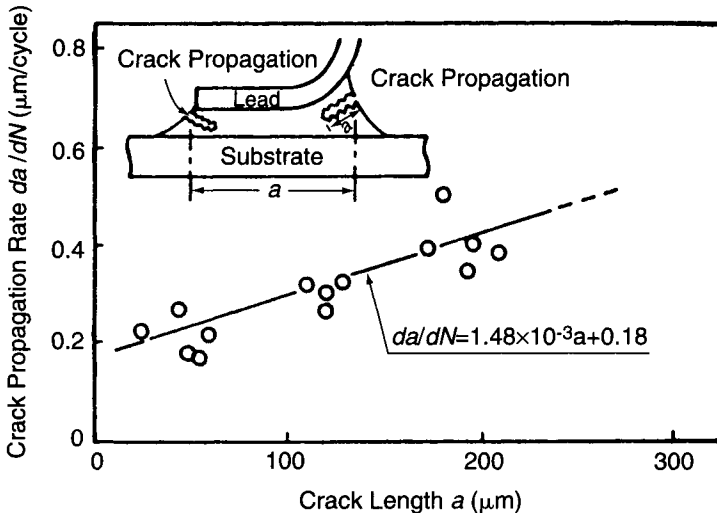


Fig. 9-23 Solder crack propagation rate as a function of crack length in a surface mounted QFP joint. From reference 35.

life of self-stretching flip-chip joints connecting a silicon die to an Al_2O_3 substrate was modeled. Assuming a temperature excursion from -50 to 150°C and one cycle per hour, the approximate three-dimensional stress distribution of the uncracked corner joint is shown in Fig. 9-24a. Employing a fracture mechanics

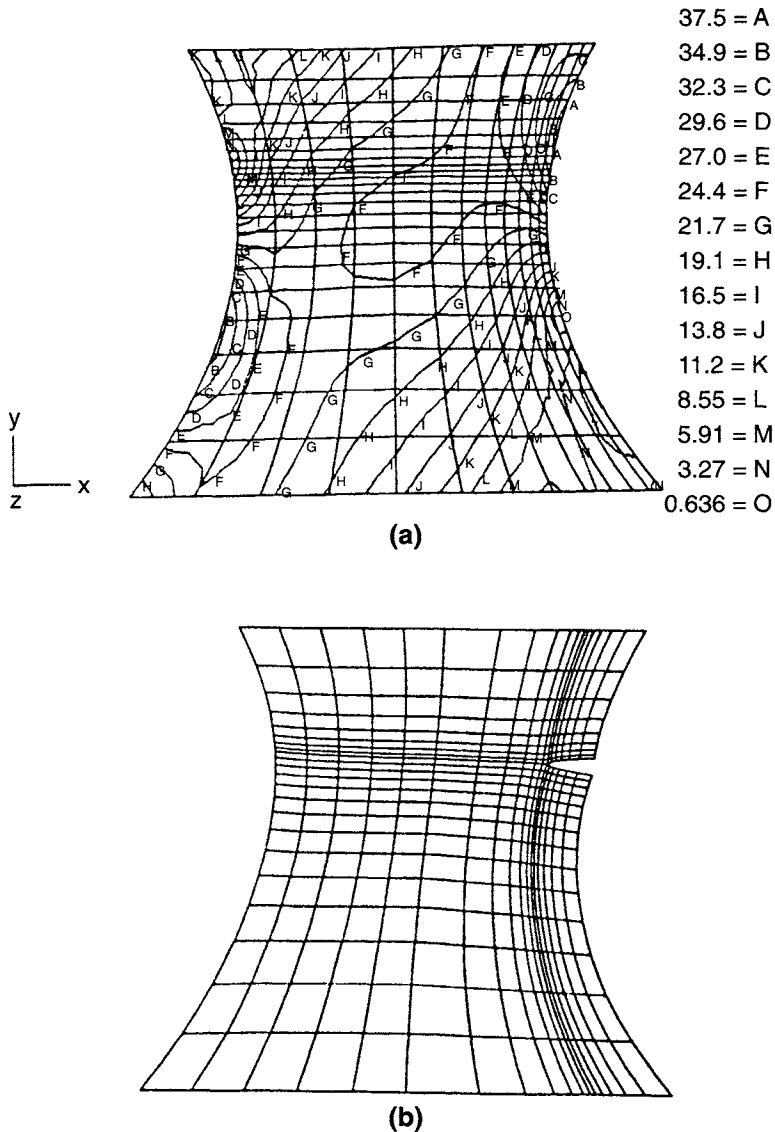


Fig. 9-24 (a) Two-dimensional stress distribution in the corner flip-chip solder joint (no crack). Stresses given in MPa are 11% less than three-dimensional results. (b) Deformation cracking of the flip-chip solder joint. From reference 36.

approach, failure of the cracked joint (Fig. 9-24b) was predicted to occur in 3000 cycles. This estimate compared well with the experimentally observed 2700 cycle failure life. Significantly, but perhaps fortuitously, thermal fatigue life prediction by both fracture mechanics and Coffin–Manson approaches were in good agreement.

9.6 DYNAMIC LOADING EFFECTS IN ELECTRONIC EQUIPMENT

9.6.1 Introduction

We now address some other reliability concerns that stem from periodic stressing of structures and components. Rather than cyclic stresses arising from thermal expansion mismatch, external dynamic loading and the resultant displacements they produce in electronic hardware are of interest now. Electronic equipment is subject to many different forms of vibration over wide frequency ranges and acceleration levels (37). These include dynamic-loading effects such as impact, shock, and acoustic noise by the action of cooling fans, repeated impact of keys during typing, accidental misuse, and vibration surrounding military activities, etc. The kinds of failure that result are due to flexure of circuit boards, fatigue, loosening and even opening of contacts, and the shorting of leads, attached components, and devices. Complete loosening of screw and nut fasteners due to vibration can eventually cause them to rattle around inside equipment with the potential for producing serious damage. Loosening is an interesting phenomenon that is not entirely understood. Apparently, the dynamic loading stretches the bolt, momentarily reducing the friction between the nut and threads; after it unwinds a bit, the bolt relaxes, retightening the nut. This cyclic action is the driving force that causes fasteners to loosen; when they do, electrical contact (e.g., grounding) may be lost.

Characteristics that distinguish dynamically generated from thermal-mismatch-induced damage include:

1. Thermal-mismatch damage and failure is more common and a greater reliability concern. Vibrational stresses tend to be small by comparison.
2. Dynamic-loading effects influence the mechanical reliability of second (e.g., circuit boards, leads) and third (e.g., board contacts, ground planes) levels of packaging more than chip-level structures and components.
3. External rather than internal stresses are the sources of damage.
4. Dynamic effects occur and are modeled at constant temperature, e.g., room temperature.

A useful review of vibration analysis as it applies to electronic packaging is due to Suhir and Lee (38), and their treatment of the subject is presented here.

9.6.2 Vibration of Electronic Equipment

9.6.2.1 Vibrating Mass The behavior of a particle of mass m attached to a spring whose spring constant is k_s is well known and provides a convenient base from which to launch the discussion. In such a one degree-of-freedom system the vibrational displacement (x) is described by the equation

$$m \frac{d^2 x}{dt^2} + k_s x = 0, \quad (9-43)$$

where the first and second terms are the inertial and restoring forces, and t is the time. The solution is of the form $x = A \sin \omega_o t$, where A is a constant and $\omega_o = (k_s/m)^{1/2}$ is related to the natural frequency (f_o) of the free vibrations by $\omega_o = 2\pi f_o$.

It often happens, however, that vibrational motion of structural elements within electronic equipment is driven by periodic forces (F) of the form $F = F_o \sin \omega t$, where F_o is the force amplitude and ω is the excitation frequency. If it is further assumed that vibrations are damped by frictional or viscous-deformation effects that are proportional to the velocity, then for damping constant η ,

$$\frac{d^2 x}{dt^2} + \frac{\eta}{m} \frac{dx}{dt} + \frac{k_s x}{m} = \frac{F_o}{m} \sin \omega t. \quad (9-44)$$

This second-order, linear differential equation is identical to that which describes an R-L-C circuit powered by ac voltage. Under the most general conditions the solution to Eq. 9-44 contains a combination of terms of the form $B \sin (\omega t + \alpha)$ and $C \sin (\omega t + \beta) \exp [-(\eta/2m)t]$, where α and β are phase angles and B and C are constants. Several physical cases of interest can be distinguished.

1. Free vibrations. If for the moment the driving force is neglected (i.e., $F_o = 0$), the displacement relative to the equilibrium position generally executes a damped oscillatory motion. Under the special condition that $\omega_o^2 = \eta/m$, the motion is not oscillatory but is critically damped and decays exponentially. Further, when $\eta/m > \omega_o^2$, the nonoscillatory motion is overdamped.

2. Steady-state vibrations. When C is zero, the amplitude of vibration is

$$B = \frac{F_o}{m\{(\omega_o^2 - \omega^2)^2 + (\eta/m \omega)^2\}^{1/2}}. \quad (9-45)$$

In practical applications, larger amplitudes generally imply less reliable function of the vibrating body. Because the dynamic displacement is inversely proportional to the frequency squared, shifting ω_o upward is an obvious strategy to

limit vibration. It is instructive to compare B with the static displacement of the mass, $x_s = F_o/k_s$ under constant load F_o . The ratio B/x_s , which is known as the amplification factor A , is then given by

$$A = \frac{k_s}{m\{(\omega_o^2 - \omega^2)^2 + (\eta/m \omega)^2\}^{1/2}}, \quad \text{or} \quad \frac{1}{\{(1 - \omega^2/\omega_o^2)^2 + [\eta/m(\omega/\omega_o)^2]\}^{1/2}}.$$

Among the possible behaviors are the following:

1. When $\omega_o > \omega$, $A \approx 1$. In this case, the amplitude of forced vibrations is equal to that of the static displacement.
2. If $\omega \approx \omega_o$, A reaches its largest value of $[\eta/m]^{-1/2}$. This is the resonant condition, a state that should be avoided because dynamic loads and stresses are magnified.
3. When $\omega \gg \omega_o$, $A \approx (\omega_o/\omega)^2$. Thus at very high applied frequencies, A approaches zero, and the system is relatively immune to vibrational damage.

9.6.2.2 Vibration of Beams Instead of the point masses above, we now consider the vibration of components that can be idealized as beams. The resulting vibrations cause transverse displacements $u(x, t)$ normal to the beam, or x -axis, that obey the partial differential equation (39)

$$\frac{m \partial^2 u(x, t)}{\partial t^2} = \frac{EI \partial^4 u(x, t)}{\partial x^4}. \quad (9-46)$$

Here $u = u(x, t)$ is the displacement function, E is Young's modulus of the material, I is the moment of inertia of the beam cross-section (EI is the flexural rigidity), and m is the mass per unit length. General solutions for the natural frequencies of vibration are

$$f_n = \frac{K_n}{2\pi} \left(\frac{EI}{m l^4} \right)^{1/2}. \quad (9-47)$$

where the value of K_n depends on the particular boundary conditions. Several specific beam geometries are shown in Fig. 9-25, where corresponding K_n values are given. For example, if the beam is freely supported or hinged at opposing ends (Fig. 9-23a), the displacements are those of the simple vibrating string. In this case $K_n = (n\pi)^2$, where n is an integer 1, 2, 3, . . . that denotes the first, second, third, . . . mode of vibration. Normally we are interested only in the first mode, because its vibrational amplitude exceeds that of the other modes.

From the standpoint of reliability, low-frequency oscillations should be avoided because they lead to large vibrational amplitudes. Therefore, other

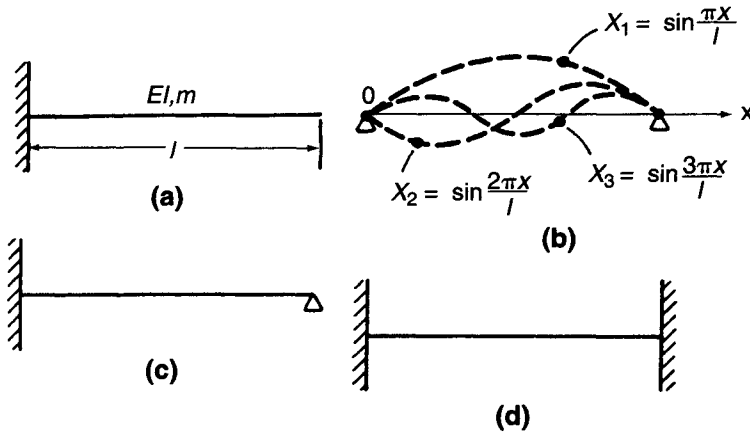


Fig. 9-25 Vibrational frequencies ($2\pi f_n = K_n[El/mt^4]^{1/2}$) of various beams. (a) cantilever; $K_1 = 3.516$, $K_2 = 22.04$, $K_n = \{(2n+1)\pi/2\}^2$ for $n = 3, 4, \dots$. (b) freely supported (hinged at both ends); $K_1 = 9.870$, $K_2 = 39.48$, $K_n = (n\pi)^2$ for $n = 3, 4, \dots$. (c) clamped-hinged; $K_1 = 15.42$, $K_2 = 49.96$, $K_n = \{(4n+1)\pi/2\}^2$ for $n = 3, 4, \dots$. (d) clamped at both ends; $K_1 = 22.37$, $K_2 = 61.67$, $K_n = \{(2n+1)\pi/2\}^2$ for $n = 3, 4, \dots$. From reference 38.

things being equal, stiffening the beam or board, reducing its mass and sectional area, and particularly, shortening its length, will all help to bring about the desideratum of higher-frequency vibrational modes. Clamping beams at either one or both ends also raises the vibrational frequencies. Electronic components whose geometry and structural behavior can be idealized as a cantilever are least desirable from the standpoint of mechanical reliability. This situation occurs in circuit boards plugged in or clamped at only one end. A simple comparison of the structures depicted in Fig. 9-25 shows that cantilevers vibrate at the lowest frequencies.

9.6.2.3 Vibration of Packaging Components The formulas just presented apply equally well to elongated plates such as long, narrow printed circuit boards. But in order to use them, EI in Eq. 9-47 must be replaced by the flexural rigidity of the plate, i.e., $Eh^3/[12(1-\nu^2)]$, where δ is the plate thickness. For wide plates of dimension $l \times W$, the formula that applies is

$$f_n = \frac{K_n}{2\pi} \left\{ \frac{Eh^3}{12(1-\nu^2)m(l^4 + W^4)} \right\}^{1/2}, \quad (9-48)$$

where the mass is now on a per unit area basis. If the plate is simply supported, $K_1 = 9.87$.

Actual components employed in packaging have complex shapes that do not lend themselves to simple vibration analyses that yield closed-form solutions. As an example let us reconsider the gull-wing joint that was introduced earlier, in section 8.5.3.3. Lau (40) has again done the finite-element analysis, which consists first in identifying the geometry and interfaces of interest and the boundary conditions that apply. Then the natural frequencies were determined in a free-vibration analysis. A more complete study would require estimation of the loads on the structure and their effect on the resultant forced vibrations.

Calculated fundamental frequencies of free vibrations are shown in Fig. 9-26 where results for soldered (or good) joints are contrasted with unsoldered (or bad) joints. The fundamental frequencies of reliably soldered gull-wing leads are approximately six times those of the leads with cracked solder joints. In effect, beams fixed either at both ends or at only one end are being roughly compared;

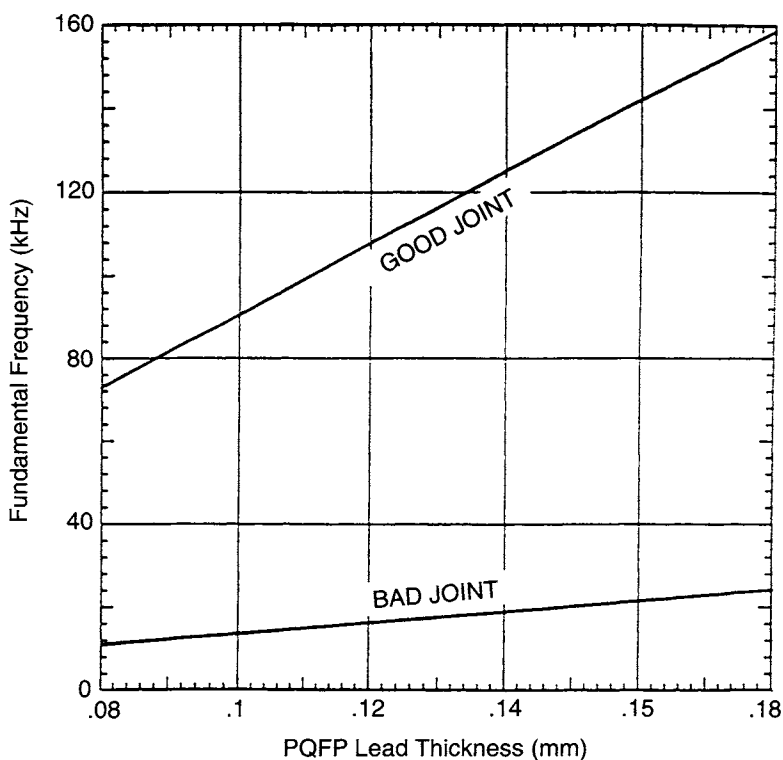


Fig. 9-26 Calculated fundamental frequencies of free vibrations for various quad flat packages having different lead thicknesses. Results for soldered (or good) joints are contrasted with unsoldered (or bad) joints. From reference 40.

beneficial low-amplitude–high-frequency vibrations characterize the former relative to the unconstrained lead with the cracked solder end. A semiquantitative calculation of the frequency ratio of the fully clamped beam (cb) to the cantilever (c) is, from Eq. 9-47, $K(\text{cb})/K(\text{c}) = 22.37/3.516$, or 6.37. The good agreement with the exact solution indicates the value of using simple beam geometries for crude estimates. An additional finding is the small change in fundamental frequency with lead thickness. Similar results apply to gull-wing leads in plastic small-outline packages, and J leads in plastic leaded chip-carriers.

The traditional method for raising the natural frequency of PCBs has been to add stiffening ribs or provide point supports. Rib stiffeners are common in computer memory, logic, and processor cards, while point supports are used to stiffen power-supply boards. A design approach to locate such support points has been advanced by Pitarresi and Di Edwardo (41). If the PCBs vibrate both in the x and y directions, the frequencies will generally differ because of the board geometry and the nonuniform placement of electronic components and devices. Optimizing arrangement of components through finite-element modeling has been explored by Chang and Magreb (42) as a way to maximize f_n .

EXERCISES

- 9-1 Consider 1 cm lengths of metal wire with diameter d making contact.
- Derive an expression for the ratio of the constriction resistance to total resistance if contact is made through a circular area of radius a .
 - For a 1 mm diameter Cu wire, what value of a is required for the constriction resistance to reach a value of 10 m Ω ?
- 9-2 The contact voltage that produces a temperature equal to the melting point of the metal is known as the melting voltage. Correspondingly, the contact voltage that produces a temperature equal to the boiling point of the metal is known as the boiling voltage. What are the melting and boiling voltages for gold and platinum contacts?
- 9-3 An a-spot in a contact shrinks in radius as a ring of oxide grows around it consuming metal in the process. If the oxide grows by a diffusional reaction, derive an expression for the time-dependent change in contact resistance.
- 9-4 Accelerated steady-state creep testing of solder yielded the following results:

Test	Temperature (°C)	Stress (MPa)	Strain rate (s ⁻¹)
1	22.5	6.99	2.50×10^{-5}
2	22.5	9.07	6.92×10^{-5}
3	46	6.99	2.74×10^{-4}

- From these data determine the creep parameters n and E_c .
- What is the maximum stress that can be applied to a 0.5 mm high solder joint of this material to keep the creep extension less than 0.05 mm after 5 years of exposure to a 30°C ambient?

9-5 The tensile stress-strain behavior for solder pulled at a rate of 3.26×10^{-5} /s gives a maximum tensile stress of 31 MPa. At a strain rate of 1.65×10^{-3} /s the maximum stress is 47 MPa. How do you account for this difference in strength?

9-6 In a study of the thickness dependence of Cu-Sn intermetallic compounds (IMC) on the thermal fatigue of surface-mount solder joints, it was found that failures were described by Weibull statistics. In particular, the data points were fitted to the function $F(x) = 1 - \exp[-(x/\alpha)^\beta]$, where $F(x)$ is the cumulative distribution of failures, and x is the number of cycles (N) to failure. Analysis of the results yielded the following:

Compound thickness (μm)	α	β	$N_{50\%}$ cycles
0.95	1727	3.31	1475
1.16	1304	3.5	1123
1.43	922	2.5	749
1.7	767	2.87	639
1.97	723	2.91	605

Data from P. L. Tu, Y. C. Chan, and J. K. L. Lai, *IEEE Trans. Components, Packaging, and Manuf. Tech.* 20, 87 (1997).

- Plot $F(x)$ versus x at the different IMC thickness values.
- Replot the data in any manner that yields straight lines.
- Provide a possible reason for the observed dependence of failure on IMC thickness.

9-7 An electrical-probe assembly consisting of springy metal contactors undergoes compressive deformation during operation. The length of the contactor is L_i initially and then shortens to L_o when constant contacting load P is applied to it. When the load is removed, the contactor length is L_f instead of the desired L_i ($L_i > L_f > L_o$). The observed permanent set or contact deformation appears to worsen when contact load is maintained for longer

time, or if the contact tips get hot. Provide an explanation for these effects and mention the material properties you would recommend for set-free contact behavior.

- 9-8 A leadless chip carrier soldered to a circuit board is thermally fatigued by turning the board on and off three times a day. Each cycle requires 30 minutes to reach the steady-state operating temperature of 50°C. Predict how many cycles the eutectic solder can sustain prior to failure if the strain in each cycle is 0.005.
- 9-9 Cracks in copper-plated through-holes were observed to grow during thermal cycling. The crack length ($2a$) varied with the number of cycles (N) as $2a = 1.16 \times 10^{-12} N^{3.77}$. Derive expressions for
- da/dN as a function of N .
 - da/dN as a function of a .
 - da/dN as a function of ΔK .
- 9-10 A high-power resistor is soldered at each end close to a PCB by means of solder posts. During operation, heat is emitted by the resistor. Discuss how the PCB warps in response.
- 9-11 A leadless ceramic chip carrier is soldered to an FR4 printed circuit board and thermally cycled between 0°C to 100°C, with 20 minute dwells at the maximum temperature. The chip carrier measures 0.635 cm $\times$ 0.635 cm and the joints are 0.013 cm high. If the CTE for the ceramic is 6×10^{-6} and that for FR4 is $12 \times 10^{-6} \text{ } ^\circ\text{C}^{-1}$, predict how many cycles will be sustained prior to failure?
- 9-12 It is desired that an electronic assembly, consisting of a chip carrier bonded through a fully compliant solder to a substrate, operate at a chip temperature T_c and a substrate temperature T_s , where $T_c > T_s$. To minimize the shear strain produced in heating the chip from T_o from T_c , and the substrate from T_o from T_s , what value of thermal expansion coefficient (α_s) would you recommend? The chip has a thermal expansion coefficient of α_c .
- 9-13 Eutectic solder fatigued at 25°C and 10 Hz displayed a crack-growth dependence on the number of cycles given by $da/dN = 2.77 \times 10^{-7} \Delta K^{3.26}$, where ΔK has units of $\text{MPa}\cdot\text{m}^{1/2}$. A ball-bond joint composed of this solder was subjected to fatigue loading, and finite-element analysis revealed that ΔK depends linearly on crack length. Specifically, $\Delta K = 0.151 \text{ MPa}\cdot\text{m}^{1/2}$ at the initial crack length of $5.65 \times 10^{-6} \text{ m}$, and $\Delta K = 0.261 \text{ MPa}\cdot\text{m}^{1/2}$ at a crack length of $1.02 \times 10^{-4} \text{ m}$ corresponding to joint fracture.
- How many cycles did the solder sustain prior to fatigue failure?

- b. If the joint is cycled once per hour and reaches a maximum temperature of 150°C, estimate how long it will take for the solder to fail.
- 9-14 At the 1996 Surface Mount International Conference it was reported that plastic ball-grid array (PBGA) solder joints whose solder balls contained void defects (up to 24% of the pad area) had no adverse effect on board-level reliability; “In fact, PBGA solder joints with voids had 16 percent better reliability than those without voids” !
- What are the likely sources of voids in freshly soldered joints?
 - Provide scientific explanations for these paradoxical findings.
- 9-15 Suppose incandescent-bulb life is governed by low-cycle thermal fatigue. The thermal strain is induced by a 2000°C rise or drop in the tungsten-filament temperature each time the bulb is turned on or off, respectively. If the equation governing lamp life is $N_f^{0.5} \Delta\epsilon_p = 0.2$, predict the number of cycles to failure.
- 9-16 Calculate the natural frequency of a vibrating cantilever spring that is loaded at the free end.
- 9-17 A simply supported beam of length L with a concentrated load P at the center exhibits a deflection given by $\delta = PL^3/(48EI)$. If such a beam can be simulated as a spring-mass system, calculate the natural frequency of vibration.
- 9-18 Experience shows that actual resonant frequencies of vibrating beams are lower than those theoretically calculated. Suggest a reason why this is so.
- 9-19 Rapid variations in the temperature of packages can theoretically induce stress because of the acceleration forces associated with dimensional changes. Consider a 30 mm long package of specific gravity 2 that is 10 mm wide and 1 mm thick. If the package undergoes a temperature rise of 100°C in 0.1 sec, what is the stress from this source? How does this value compare with typical thermal stresses?
- 9-20 A vibration test was conducted on circuit cards installed within a computer mainframe. The cards lay in the horizontal plane cantilevered off the motherboard. In ramping the frequency up to 60 Hz one of the circuit cards began to vibrate excessively at 50Hz. The involved card made of fiberglass ($E = 38\text{GPa}$), weighs 80 grams and is 30 cm long, 3 cm wide and 0.1 cm thick. How thick should the card be to displace the first mode of vibration to 120Hz? Mention another way to increase the vibrational frequency of the card.

- 9-21 Determine the resonant frequency of a rectangular plug-in circuit card assembly that is simply supported on all four sides. The epoxy-glass printed circuit board weighs 0.545 kg, has dimensions of 22.9 cm $\times$ 17.8 cm $\times$ 0.2 cm and elastic properties defined by $E = 14$ GPa and Poisson's ratio of 0.12.

REFERENCES

1. R. Holm, *Electrical Contacts-Theory and Applications*, 4th ed., Springer Verlag, New York (1967).
2. F. Llewellyn Jones, *The Physics of Electrical Contacts*, Clarendon Press, Oxford (1957).
3. M. Braunovic', *Proc. 40th IEEE Holm Conf. on Electrical Contacts*, p. 1, (1994).
4. M. Runde, *IEEE Trans. Components, Hybrids and Manuf. Tech.*, CHMT-10, 89 (1987).
5. M. Runde, H. Kongjorden, J. Kulsetas, and B. Todtal, *IEEE Trans. Components, Hybrids and Manuf. Tech.*, CHMT-9, 77 (1986).
6. R. S. Mroczkowski, *Proc. First Electronic Materials and Processing Congress*, p. 159, Chicago, Sept. 24-30 (1988).
7. N. A. Stennett, T. P. Ireland, and D. S. Campbell, *IEEE Trans. Components, Hybrids and Manuf. Tech.* **14**, 45 (1991).
8. N. A. Stennett and D. S. Campbell, *IEEE Trans. Components, Packaging and Manuf. Tech.* **17A**, 128 (1994).
9. P. A. Engel, *IEEE Trans. on Reliability* **42** (2), 262 (1993).
10. J. F. Archard, *J. Appl. Phys.* **24**, 981 (1953).
11. M. Antler, *IEEE Trans. Components, Hybrids and Manuf. Tech.* **8**, 87 (1985).
12. M. Antler, *IEEE Circuits and Devices* **3**, 8 March (1987).
13. M. Braunovic', *IEEE Trans. Components, Hybrids and Manuf. Tech.* **15**, 205 (1992).
14. M. D. Bryant, *IEEE Trans. Components, Packaging and Manuf. Tech.* **17**, 86 (1994).
15. J. H. Lau, in *Thermal Stress and Strain in Microelectronics Packaging*, ed. J. H. Lau, Van Nostrand Reinhold, New York (1993).
16. J. H. Lau, ed., *Solder Joint Reliability*, Van Nostrand Reinhold, New York (1991).

17. J. H. Lau, ed., *Thermal Stress and Strain in Microelectronics Packaging*, Van Nostrand Reinhold, New York (1993).
18. R. Darveaux and K. Banerji, *IEEE Trans. Components, Hybrids and Manuf. Tech.* **15**, 1013 (1992).
19. J. H. Lau, *Soldering and Surface Mount Tech.* **13**, 45, Oct. (1993).
20. Z. Guo, Y.-H. Pao, and H. Conrad, *J. Electronic Packaging, Trans AIME* **117**, 100, June (1995).
21. G. F. Dieter, *Mechanical Metallurgy*, 3rd ed., McGraw-Hill, New York (1986).
22. D. F. Socie, M. R. Mitchell, and E. M. Caulfield, *Fundamentals of Modern Fatigue Analysis*, FCP Report No. 26, U of Illinois (1978).
23. H. D. Solomon in *Solder Joint Reliability*, ed. J. H. Lau, Van Nostrand Reinhold, New York (1991).
24. L. F. Coffin, *ASTM-STP* **520**, 5 (1973).
25. M. Ohring, *Engineering Materials Science*, p. 340, Academic Press, San Diego (1995).
26. W. Engelmaier, in *Solder Joint Reliability*, ed. J. H. Lau, Van Nostrand Reinhold, New York (1991).
27. W. Engelmaier, *Soldering and Surface Mount Tech.*, **No. 1**, 14, Feb. (1989).
28. J.-P.M. Clech and J.A. Augis, in *Solder Joint Reliability*, ed. J. H. Lau, Van Nostrand Reinhold, New York (1991).
29. J. W. Balde, *IEEE Trans. Components, Hybrids and Manuf. Tech.* **10**, 463 (1987).
30. K. C. Norris and A. H. Landzberg, *IBM J. Res. and Develop.* **13**, 266 (1969).
31. J. H. Lau, Y.-H. Pao, C. Lerner, R. Govila, S. Twerefour, D. Gilbert, E. Erasmus, and S. Dolot, *Soldering and Surface Mount Tech* **16**, 42, Feb. (1994).
32. R. B. Clough, A. J. Shapiro, A. J. Bayba, and G.K. Lucey, *J. Electronic Packaging, Trans ASME* **117**, 270 (1995).
33. E. E. de Kluizenaar, *Soldering and Surface Mount Tech* **4**, 27, Feb. (1990). and E. E. de Kluizenaar, *Soldering and Surface Mount Tech* **5**, 56, June (1990).
34. Z. Guo, A. F. Sprecher, H. Conrad, and M. Kim, *Proc. Conf. Materials Developments in Microelectronic Packaging*, Montreal, p. 155, Aug. (1991).
35. R. Satoh, K. Arakawa, M. Harada and K. Matsui, *IEEE Trans. Components, Hybrids and Manuf. Tech.* **14** (1), 224 (1991).
36. J. H. Lau, ed., *Flip Chip Technologies*, McGraw Hill, New York (1996).
37. D. S. Steinberg, *Vibration Analysis for Electronic Equipment*, 2nd ed, Wiley, New York (1988).
38. E. Suhir and Y. C. Lee, in *Electronic Materials Handbook, Volume 1, Packaging*, p. 45. ASM International, Materials Park, OH (1989).

39. M. Paz, *Structural Dynamics*, Van Nostrand Reinhold, New York (1982).
40. J. H. Lau, in *Solder Joint Reliability* ed. J. H. Lau, Van Nostrand Reinhold, New York (1993).
41. J. M. Pitarresi and A. V. Di Edwardo, *J. Electronic Packaging, Trans ASME* **115**, 119 (1993).
42. T.-S. Chang and E. B. Magreb, *J. Electronic Packaging, Trans ASME* **115**, 312 (1993).