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Presented at the Department of Navy 13th Annual Electronics Manufacturing Seminar, China Lake, CA, March 2, 1989, and to be published in the Proceedings

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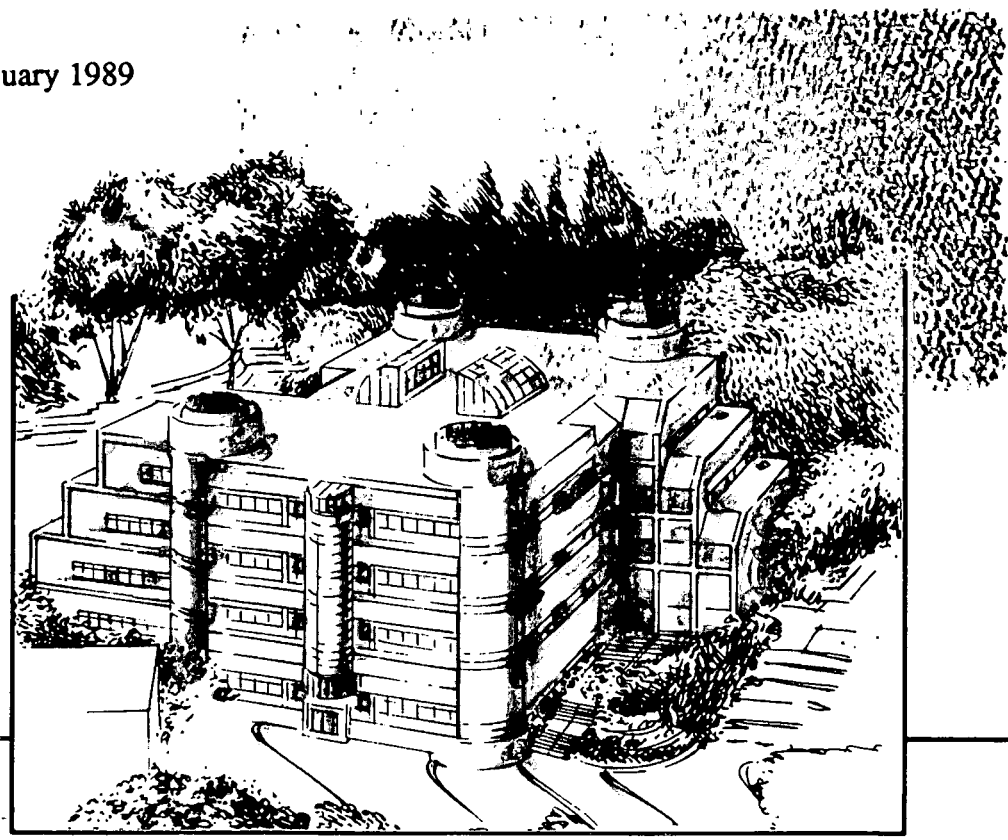
January 1989



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Prepared for the U.S. Department of Energy under Contract DE-AC03-76SF00098

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LBL 26565

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This work was supported by Director, Office of Energy Research, Office of Basic Energy Sciences, Materials Science Divison of the U. S. Department of Energy under Contract No. *DE-AC03-76SF00098*

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Abstract

This paper discusses the microstructures of solder joints and the mechanisms of thermal fatigue, which is an important source of failure in electronic devices. The solder joints studied were near-eutectic Pb-Sn solder contacts on copper. The microstructure of the joints is described. While the fatigue life of near-eutectic solder joints is strongly dependent on the operating conditions and on the microstructure of the joint, the metallurgical mechanisms of failure are surprisingly constant. When the cyclic load is in shear at temperatures above room temperature the shear strain is inhomogeneous, and induces a rapid coarsening of the eutectic microstructure that concentrates the deformation in well-defined bands parallel to the joint interface. Fatigue cracks propagate along the Sn-Sn grain boundaries and join across the Pb-rich regions to cause ultimate failure. The failure occurs through the bulk solder unless the joint is so thin that the intermetallic layer at the interface is a significant fraction of the joint thickness, in which case failure may be accelerated by cracking through the intermetallic layer. The coarsening and subsequent failure is influenced more strongly by the number of thermal cycles than by the time of exposure to high temperature, at least for hold times up to one hour. Thermal fatigue in tension does not cause well-defined coarsened bands, but often leads to rapid failure through cracking of the brittle intermetallic layer. Implications are drawn for the design of accelerated fatigue tests and the development of new solders with exceptional fatigue resistance.

1. Introduction

Solder contacts are critical to electronic packaging since they provide both the electrical and the mechanical connection between different levels of the package. As device miniaturization has led to increasingly dense electronic packages the mechanical integrity of the solder contacts has become a serious concern [1-10]. A major problem is thermal fatigue, which arises from the thermal expansion mismatch between the materials joined by the solder. The device is heated and cooled by its own power cycle. Current flow leads to Joule heating when the device is turned on. The on-off cycle induces a temperature change that may range from room temperature to near 100°C. Specialty devices, such as those used in aerospace avionics, may experience an even larger thermal cycle that results from

exposure to the environment, for example, "on" on the ground on a hot summer day to "off" in chill air at 40,000 ft. Military specifications suggest an environmental thermal cycle that varies between extremes at -55°C and 125°C.

The research that is reviewed here concerns the metallurgical mechanisms of the thermal fatigue of near-eutectic Pb-Sn solder contacts on copper, since this is the most common class of contacts in the industry. There are two principal reasons to be concerned about the mechanism of thermal fatigue, and they address the two problems that need to be overcome to ensure joint reliability in service.

First, the industry needs good analytic models or accelerated fatigue tests to guide and verify package designs. While useful tests can be designed empirically, and several have been proposed [10-14], these become uncertain when they are applied to devices, geometries or operating environments outside the set for which they have been specifically verified. A theoretical model or accelerated test is only valid to the extent that it reproduces the mechanisms that lead to failure in service; otherwise it may yield results that are badly misleading.

Second, the industry needs fatigue-resistant solders that can survive the severe conditions that will be experienced in very dense microelectronic packages. The design of improved solders also requires that the metallurgical mechanisms of thermal fatigue be well understood so that appropriate metallurgical modifications can be introduced to defeat them.

From the metallurgical perspective the problem of thermal fatigue in solder contacts is as complex as any fatigue problem that has been researched. Since the solder is mechanically soft and is used at a high homologous temperature (a large fraction of its melting point) deformation is introduced by plasticity and creep (stress relaxation) at both ends of the strain cycle. The cyclic deformation is affected by the fact that the solder contact is a thin layer bonded to relatively rigid materials with a complex intermetallic reaction layer at the bonding line. The long-term deformation behavior is further complicated by the microstructural changes that inevitably occur in the solder as it is cycled and aged. A review of the literature will show that the first-order problem of fatigue crack growth at low temperature under simple periodic loading is only partly understood. There is no reliable predictive theory for thermal fatigue under creep conditions for even simple monolithic materials. The problem of predicting the rate of thermal fatigue in a solder joint is more formidable still, and is unlikely to be solved in the near future. Nonetheless mechanistic research on the fatigue of solder joints can provide valuable information to help guide the development of accelerated tests and the creation of improved solders.

In work published elsewhere we have studied the strength of near-eutectic solders in tension and shear [15], the isothermal and thermal fatigue behavior in tension and shear [16-18], and the creep and stress relaxation of bulk solders and solder joints under shear loads [19,20]. In this paper we concentrate on the mechanisms of thermal fatigue and their association with the microstructure of the solder joint.

2. Experimental Details

To conduct mechanistic research on the fatigue of solder joints one needs metallographic sample preparation techniques to characterize the state of the joint and mechanical testing procedures that can subject solder joints to known loads and loading geometries. Both of these experimental needs present significant problems, and deserve some comment.

The preparation of good metallographic samples from eutectic solder joints is complicated by the fact that the joint consists of a mechanically soft Pb-Sn phase that wets a relatively hard substrate (Cu) through the formation of brittle intermetallic compounds. Metallographic sample preparation is difficult. Some useful techniques are described in reference [15].

The mechanical tests conducted as part of this work include tests to failure in tension and shear to determine the strengths of solder joints in these two configurations, isothermal and thermal fatigue tests under cyclic loading in tension and shear, and isothermal creep and stress relaxation tests under fixed stress (creep) or strain (stress relaxation) in tension or shear.

To accomplish these mechanical tests we require specimens that ensure loading in simple tension or shear under isothermal or thermal cycling conditions. Figures 1-4 show the mechanical test specimens that we have used. The specimen shown in Fig. 1 is designed for isothermal strength and fatigue testing in tension, and has a solder joint in the center of a tension specimen [15,21]. Fig. 2 shows the specimen used for isothermal strength, fatigue, creep, and stress relaxation testing in balanced shear. It consists of a three-layer sandwich of copper plates joined by solder. The sample is notched so that the solder joints in the central region of the plates are in simple shear when the sample is loaded in tension, as can be verified by using scribed specimens [16,17]. Fig. 3 shows the specimen used for thermal fatigue tests in shear [18]. It is a three-layer sandwich of copper-aluminum-copper joined by solder lines. The central aluminum plate is plated with copper, so that the joint is solder-copper on both sides. Thermal fatigue testing is done by alternately immersing the sample in constant temperature baths (usually -55°C and 125°C) for selected periods of time; the strain is imposed by the thermal expansion mismatch between copper and aluminum. Fig. 4 shows the specimen used for thermal fatigue in tension [22]. A specimen that consists of two copper plates joined by solder is inserted into the aluminum fixture shown in the figure. When the temperature is changed, for example, by immersing the specimen into a heated or cooled bath, the thermal expansion mismatch between the copper and the aluminum imposes a tensile strain across the solder joint. These specimens are useful for fundamental research in that they establish a known macroscopic strain across a solder joint. Other specimen configurations which will not be described here are used to study more realistic joint configurations.

3. The Microstructure of Near-Eutectic Solder Joints

The central principle of Materials Science states that the properties of a material are determined by its composition and its *microstructure*, the manner in which the atoms are arranged on the microscopic scale. It follows that if we are to understand the mechanical behavior of solder joints we must begin with an understanding of the microstructures in which they would most commonly be found. Figure 5 shows a cross-section through a 60Sn-40Pb solder contact on copper at relatively low magnification. The cross-section contains three distinct elements: the copper, the solder, and a thin intermetallic layer along the interface between them. In the cases that interest us here the microstructure of the copper is irrelevant to the behavior of the joint; the stresses imposed are so far below the yield strength of the copper that it behaves essentially as a rigid body. However, both the solder and the intermetallic layer may influence the mechanical behavior of the joint. We therefore describe the microstructures of these.

The interfacial intermetallic layer

The strong adhesion that is established between Pb-Sn solders and copper is due to a chemical reaction between tin and copper that establishes a thin, adherent layer of copper-tin intermetallics along the interface. The intermetallic layer between a near-eutectic solder and copper is, in fact, a bilayer, as shown in Figure 5. The copper surface is coated by a thin layer of the ε-phase intermetallic, Cu_3Sn, which is separated from the solder by a thicker layer of the η-phase intermetallic, Cu_6Sn_5. The ε-phase consists of small, columnar grains that grow outward from the copper surface. When the ε-phase exists alone, as it does, for example, at the interface of a 95Pb-5Sn solder with copper, the grains are compact and faceted, as shown in Figure 6. The intermetallic is extremely brittle, and fractures easily along the grain boundaries of the Cu_3Sn grains when the surface is bent. The η-phase, on the other hand, presents a very rough interface to the solder. Grains of Cu_6Sn_5 grow out into the solder in spiky, whisker-like crystals of hexagonal morphology, which reflect the hexagonal crystal structure of the η-phase. The morphology of an interface with 60Sn-40Pb is shown in Figure 7. This intermetallic is also brittle. Grains of Cu_6Sn_5 cleave easily under stress.

Large η-phase precipitates are also found in the bulk of the solder. These have the curious morphology shown in Figure 8; they are long, hexagonal rods that are hollow along their axes and filled with solder. Specific studies [23] show that these intermetallic precipitates form as whiskers on the Cu surface during initial wetting, break off the surface, and redissolve slightly in the liquid solder before solidification to acquire their hollow appearance. Despite their size, these intermetallic precipitates usually have no more than a secondary influence on the mechanical properties of the solder joint.

Bulk solder

It is important to recognize near-eutectic Pb-Sn solder does not have a unique microstructure, but may be found in any one of several characteristic microstructures depending on the manner in which it was processed. The most common microstructure is

shown in Figure 9, and can be understood in terms of the Pb-Sn phase diagram that is shown in Figure 10. A eutectic solder (63Sn-37Pb) solidifies near the eutectic temperature in the phase diagram to form a classic eutectic microstructure that consists of parallel lamellae of Pb-rich phase in a matrix of Sn. The microstructure is divided into grain-like colonies within each of which the Pb-rich lamellae are nearly parallel. These colonies are the features that are referred to as "grains" in much of the technical literature on solder; they are not true grains at all, since they contain two distinct phases and many distinct crystallites.

When the solidified solder is cooled to room temperature the solubility of Sn in the Pb-rich phase decreases. The Sn supersaturation is relieved by the formation of Sn precipitates within the Pb-rich lamellae. Transmission electron microscopic studies [27] show that these are plate-like particles of Sn that have stepped surfaces and grow or shrink easily as the temperature is changed to establish an equilibrium Sn concentration within the Pb-rich phase. When the solder is slightly off-eutectic in composition, for example, 60Sn-40Pb, distinct (pro-eutectic) grains of Pb-rich phase also appear.

As the solder is made to solidify at increasingly high cooling rates the microstructure of the solder changes in that the Pb-rich regions of eutectic become more rod-like and closely spaced and the colony size becomes smaller, as shown in Figure 11. At very high cooling rates the Pb-rich phase may appear in the form of small, equiaxed particles.

The eutectic microstructure of as-solidified solder is unstable because of the very high surface to volume ratio of the grain it contains. If the solder is held at room temperature or above, the eutectic microstructure coarsens through a continuous, diffusion controlled process. The coarsening is particularly pronounced near the colony boundaries, where the grains can reconfigure with relative ease. The coarsening process is slow, but continuous, and has the consequence that the appearance and the mechanical properties, such as strength or hardness, change continuously with time after the solder joint is made.

The coarsening process is strongly accelerated if the solder is deformed mechanically. Mechanical deformation triggers recrystallization at room temperature and above. The microstructure changes into an equiaxed mixture of Sn and Pb-rich grains, as illustrated in Figure 12. Recrystallization is followed by grain growth at a rate that increases with temperature. Recrystallization dramatically changes the mechanical properties of the solder [24]. The recrystallized material is much softer than the same material in the eutectic microstructure, and deforms in creep at a very high relative rate. Recrystallized solder is superplastic, while solder in the eutectic microstructure is not.

Given the variety of microstructures in which near-eutectic solders can be found and the strong influence of the microstructure on mechanical properties, it is hardly surprising that compilations of the quantitative mechanical properties of eutectic solder show a substantial scatter in the measured results.

4. The Mechanism of Thermal Fatigue in Shear

Because the thermal fatigue of a solder joint is driven by the differential thermal expansion of the surfaces that bind the joint, the dominant load on the joint is often a simple shear, and the failure (or, at least, the initial failure) of solder joints in service often occurs in the portion of the joint at which the shear load is applied. For these reasons thermal fatigue under shear loads is of particular interest.

Since there are many possible solder joint configurations and several possible microstructures of the solder in solidified joints, it is quite possibly premature to talk about "the" mechanism of thermal fatigue of eutectic solder in shear. Nonetheless, every test of which we are aware in which a eutectic solder joint was deformed to failure in thermal fatigue, isothermal fatigue, or creep at high temperature under predominantly shear loading led to failure by the same basic mechanism.

The basic mechanism of fatigue in shear

The mechanism is illustrated in Figure 13, which shows the behavior of a 60Sn-40Pb joint as it is cycled from -55°C and 125°C. As the cycling proceeds the shear deformation concentrates into bands that are clearly marked because of the associated rapid coarsening of the microstructure. Figure 14 is a detailed view of a coarsened band. Fatigue cracks grow through these bands, primarily along the Sn-Sn grain boundaries as shown in Figure 14. A review of published cross-sectional micrographs of solder contacts that failed during actual or simulated device operation shows that this fatigue mechanism has been widely observed in near-eutectic solders [9,10], though it is rarely noted.

The mechanism is intimately connected with the evolution of the microstructure and is accelerated by the destructive interference between inhomogeneous shear deformation and microstructure coarsening. Inhomogeneous shear deformation in the solder triggers inhomogeneous coarsening of the eutectic microstructure that is concentrated in the bands of maximum shear. Since the microstructure softens as it coarsens, shear deformation concentrates in the coarsened regions, which therefore coarsen more rapidly. Eventually, cracks form and propagate along the Sn-Sn grain boundaries. Once the Sn grains have failed the Pb-rich phase easily separates to complete fatigue failure.

The microstructural mechanism of thermal fatigue

A closer study of the sequence of microstructural changes that occur during shear failure, whether it is through thermal fatigue [18], isothermal fatigue above room temperature [16,17] or high temperature creep [20,25] reveals the same sequence of internal changes, which can be described as follows.

The microstructural event that initiates the pattern of failure is the inhomogeneity of plastic deformation of a eutectic microstructure that is deformed in shear. Figure 15 shows the deformation pattern that forms when a near-eutectic solder is deformed slowly in shear at elevated temperature, as revealed by relief at the surface of the specimen. The deforma-

tion pattern includes a nearly straight band of deformation that lies parallel to, but slightly displaced from the solder-copper interface together with several irregular deformation bands that lie approximately perpendicular to the interface and cross the solder joint. A metallographic examination of the same surface after a light polish and etch establishes the correspondence between the deformation bands, Figure 15 right, and the bands of coarsened microstructure, Figure 15 left. Closer inspection shows that the parallel band in Figure 15 crosses eutectic colonies while the perpendicular bands tend to follow colony boundaries. In all cases the parallel deformation band originates at one side of the lateral surface of the solder joint. Finite element analyses [28] suggest that the point of origin is near the principal stress concentration point in the sample.

The mechanism that initiates the coarsened band seems straightforward. It is well known [24] that the eutectic microstructure in the Pb-Sn system is unstable with respect to recrystallization if it is worked at a high homologous temperature or exposed to high temperature after deformation. Inhomogeneous shear strain creates a band of concentrated plastic deformation, which is greatest at the free surface near the stress concentration point. The eutectic recrystallizes there to relieve the accumulated deformation. Since the deformation is local and largely confined to a thin band, only this band of material recrystallizes.

The mechanism of growth of the recrystallized band also seems straightforward. The band grows by the progressive recrystallization of the deformed material. Recrystallization occurs predominantly at the tip of the planar band of previously recrystallized material, and is expected to appear there because this should be the most highly deformed element of material. Since the recrystallized material is soft compared to the eutectic material surrounding it, a narrow band of recrystallized or coarsened material in a eutectic matrix behaves mechanically very much like a narrow crack. Since the applied stress is a simple shear in the plane of the crack the strain field at the crack tip is a mode II deformation zone. The plastic strain field for a mode II crack has been solved for a Von Mises material by McClintock, et al. [26] and is diagrammed in Figure 16. Note that the deformation field is narrow and tends to confine the deformation to the plane of the crack. The next increment of recrystallization near a propagating band of recrystallization that behaves like a mode II crack should occur in the most highly strained material, which lies in the plane of the crack near the crack tip. It follows that the recrystallized band tends to remain plane and narrow while it extends itself along a shear plane, as is observed.

Inhomogeneous shear deformation in the solder triggers inhomogeneous coarsening of the eutectic microstructure that is concentrated in the bands of maximum shear. Since the microstructure softens as it coarsens, shear deformation concentrates in the coarsened regions, which therefore coarsen more rapidly. The preferential coarsening of the Sn grains in the shear bands is pronounced (Figure 17). Eventually, the accumulated strain cracks the Sn-Sn grain boundaries. The propagating cracks separate the Pb-rich material to complete fatigue failure.

Features of the shear failure mechanism

There are several features of the microstructural mechanism of shear failure in solder that affect the lifetime of particular solder joints. These include the influence of colony size, joint thickness and cycle frequency on the rate of failure.

It has often been observed that decreasing the apparent grain size of a eutectic solder (which is actually the eutectic colony size) tends to increase its resistance to fatigue [9]. While there is, in fact, very little quantitative data supporting this conclusion, the observation makes sense in terms of the microstructural mechanism of shear failure. A solder with fine colony size deforms more homogeneously in plastic deformation, and is hence more resistant to the formation of the deformation bands that trigger coarsening and failure.

Second, arguments based on continuum mechanics suggest that the fatigue resistance of a solder joint should increase dramatically if the joint is made thicker, since this decreases the overall shear deformation per cycle. While there is merit in this approach the thickening of the solder joint is not nearly as effective as one might suppose when the joint is loaded in shear [22]. There are two reasons for this. First, the shear deformation remains inhomogeneous when the joint is made thicker. Once well-developed deformation bands appear, the shear deformation concentrates there, so the local shear in the planar band where failure is actually occurring depends much more on the total strain across the joint than on the joint thickness. The value of a thicker joint is in the smaller initial shear strain, which may retard the formation of bands of concentrated shear. In practice, this beneficial effect is counterbalanced by the slower cooling rate of a thicker joint, which leads to a larger colony size and a greater tendency toward inhomogeneous deformation.

However, there is a significant deterioration in thermal fatigue resistance when the joint is made so thin, or the intermetallic layer is so coarsened by reflow, that the thickness of the brittle intermetallic layer becomes a significant fraction of the thickness of the joint [22]. In this case the crack propagation path may branch out of the ductile solder into the brittle intermetallic layer, causing a relatively rapid failure. An example is shown in Figure 18, which is a micrograph of a cross-section of a 2 mil eutectic solder joint on Cu after thermal fatigue. The crack grows partly through the intermetallic layer.

Third, since the failure mechanism in shear is associated with recrystallization and grain coarsening, it is driven by shear strain. High temperature contributes to failure because it increases the rate of recrystallization and coarsening, but does not lead to banding or failure by itself. This is illustrated in Figure 20, which shows a cross-section through a solder joint in which the solder continues beyond the end of the joint. Thermal fatigue creates an obvious band of coarsened material in the sheared region within the joint, but results in only a small, overall coarsening in the material outside the joint that has experienced the same thermal cycle without the cyclic shear stress. This has the consequence, which is documented by the thermal fatigue studies discussed in [22] that the joint lifetime in thermal fatigue is much more sensitive to number of shear cycles than to the total time of exposure

at high temperature, and suggests that shear fatigue tests can be accelerated without substantial loss of quantitative information.

5. The Mechanism of Thermal Fatigue in Tension

The mechanism of thermal fatigue under cyclic tensile loads perpendicular to the plane of the joint differs significantly from that observed under shear loads [22]. The fatigue failure in tension initiates more rapidly than in shear, and tends to occur through crack growth through the brittle intermetallic layer, as shown in Figure 21. The final fatigue failure contains cracks through the intermetallic that are joined by cracks through the bulk solder near the interface. Bands of coarsened microstructure do not form in the striking way observed for fatigue in shear. The coarsening is more general, and tends to concentrate along colony boundaries perpendicular to the interface, as shown in Figure 22.

The fatigue behavior of solder joints under tensile loads contrasts with that of bulk eutectic solder. The data suggest that solder joints are more liable to failure under tensile load than shear loads, while bulk solder fails more quickly in shear. The reason is the tensile failure of the brittle intermetallic layer. These results suggest that solder joints be designed to minimize tensile loading across the intermetallic. On the other hand, tensile loads are less damaging if they occur in the bulk solder away from the intermetallic, as they do, for example, in the outer regions of the solder attachment of a chip carrier to a circuit board.

6. Conclusion

The purpose of the work reported here was to clarify the microstructural mechanisms of fatigue in solder to assist in the design of probative accelerated tests and the possible design of fatigue-resistant solders. Its principle conclusions are the following.

While the fatigue life of near-eutectic solder joints is strongly dependent on the operating conditions and on the microstructure of the joint, the metallurgical mechanisms of failure are surprisingly constant. When the cyclic load is in shear at temperatures above room temperature the shear strain is inhomogeneous, and induces a rapid coarsening of the eutectic microstructure that concentrates the deformation in well-defined bands parallel to the joint interface. Fatigue cracks propagate along the Sn-Sn grain boundaries and join across the Pb-rich regions to cause ultimate failure. The failure occurs through the bulk solder unless the joint is so thin that the intermetallic layer at the interface is a significant fraction of the joint thickness, in which case failure may be accelerated by cracking through the intermetallic layer. The coarsening and subsequent failure is influenced more strongly by the number of thermal cycles than by the time of exposure to high temperature, at least for hold times up to one hour. Thermal fatigue in tension does not cause well-defined coarsened bands, but often leads to rapid failure through cracking of the brittle intermetallic layer.

These results suggest that good accelerated tests for the solder joint life in thermal fatigue can be developed, but must be studied with some care to ensure that the mechanism of fatigue is not altered by the acceleration. It appears that thermal fatigue tests can be replaced by isothermal fatigue tests, which are much easier to perform, provided that the isothermal tests study solder joints under controlled load conditions, and are done at sufficiently high temperatures that the microstructural changes that accompany thermal fatigue can occur. Otherwise the test is of limited value. In particular, it would appear that isothermal fatigue tests on bulk solder in tension, which have been proposed as a means for estimating joint fatigue life, are demonstrably useless for that purpose.

The results also suggest that it may be possible to develop solders with fatigue properties that are substantially better than those of the materials now in use. The guidelines for alloy design are relatively clear. The improved solders should deform more homogeneously under shear loads, hence minimizing the strain concentrations that lead to recrystallization and coarsening, and should be compositionally modified to resist recrystallization and coarsening at normal operating temperatures. The challenge is to achieve these material properties.

Acknowledgement

This work was supported by the Director, Office of Energy Research, Office of Basic Energy Sciences, U.S. Department of Energy, under Contract NO. DE-AC03-76SF00098.

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FIGURE CAPTIONS

Figure 1. Copper/solder composite specimen used for isothermal strength and fatigue testing in tension. The solder joint is the dark shaded region in the center of the specimen. All dimensions are in mm.

Figure 2. Specimen configuration used for isothermal strength, fatigue, creep, and stress relaxation testing. The specimen is a multilayer copper/solder sandwich. On axial loading the two solder joints, shaded regions in figure, are loaded in nearly simple shear. All dimensions are in mm.

Figure 3. Specimen used for thermal fatigue experiments. The dark central regions are the solder joints. The mismatched thermal expansion characteristics of the constraining copper and aluminium strain the joint in shear with each temperature excursion.

Figure 4. Specimen jig used to perform thermal fatigue in tension. The specimen shown on the right fits into the Al jig on the left. On thermal cycling, the solder joint, shaded region in specimen, is deformed in tension as a result of the uneven thermal expansion between the Al jig and the Cu sample. All dimensions are in mm.

Figure 5. Optical micrograph of a 60Sn-40Pb solder joint on copper. Both intermetallic layers are clearly visible.

Figure 6. Optical micrograph of a 95Pb-5Sn/Cu interface showing the Cu_3Sn intermetallic.

Figure 7. SEM micrograph of a 60Sn-40Pb/Cu interface. The solder has been etched away to reveal the spiky, whisker like crystals of the Cu_6Sn_5 intermetallic.

Figure 8. SEM micrograph of a polished 60Sn-40Pb surface etched to reveal the intermetallic morphology. Note that the Cu_6Sn_5 intermetallic takes the shape of a hollow hexagonal rod whose core is filled with solder.

Figure 9. Optical micrograph of a typical 60Sn-40Pb solder joint. The dark regions are the Pb-rich phase. Note the phase distribution into grain-like features, i.e. colonies.

Figure 10. The Pb-Sn phase diagram.

Figure 11. SEM micrographs of Pb-Sn solder cooled at different cooling rates. The upper most figure underwent the slowest cooling rate, the bottom specimen was quenched. Note the disappearance of lamellar and colony-like features with increasing cooling rate.

Figure 12. Optical micrograph showing a worked and annealed 60Sn-40Pb microstructure. The thermomechanical treatment results in a recrystallized microstructure consisting only of equiaxed Pb-rich and Sn-rich grains. The recrystallized microstructure bears no resemblance to the initial as-cast one.

Figure 13. A series of optical micrographs showing the evolution of a coarsened microstructure within the solder joint as a function of thermal cycles.

Figure 14. A detailed view of the microstructure within the coarsened band. Note that fatigue cracks grow primarily along the Sn-Sn grain boundaries. (Optical micrograph).

Figure 15. Optical micrograph of a solder joint deformed in creep. On the right is the deformation pattern that develops on the surface of the joint as it is slowly deformed. The micrograph on the left is the same joint after a light polish. Note correspondence between deformation pattern and location of coarsened bands.

Figure 16. The plastic strain field for a mode II crack.

Figure 17. SEM micrograph of a solder joint showing a well developed coarsened band.

Figure 18. An optical micrograph of a 2 mil eutectic solder joint. Note that the joint failure occurs partly through the intermetallic layer.

Figure 19. Optical micrograph of a solder joint in which the solder extends beyond the length of the joint. The coarsened band forms only in the sheared solder.

Figure 20. Optical micrograph of a solder joint that has been thermally fatigued in shear. The fatigue failure contains cracks through the brittle intermetallic layer.

Figure 21. A series of micrographs showing the evolution of solder microstructure as the joint is thermally cycled in tension. The coarsening is general and tends to concentrate at the colony boundaries.

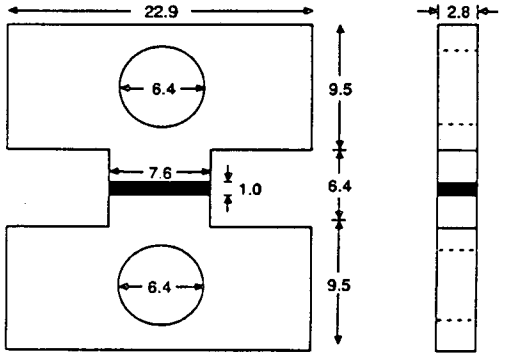
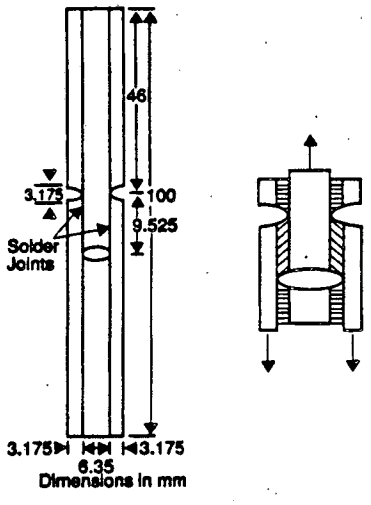


Figure 1.



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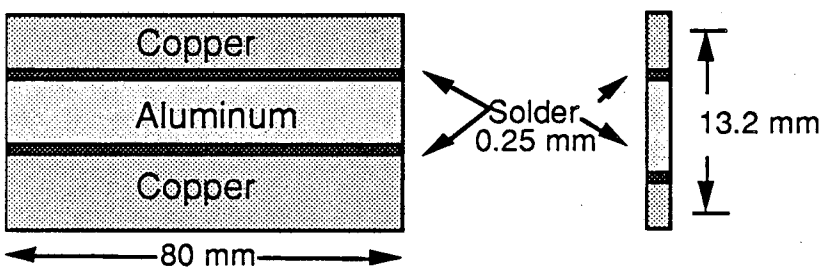

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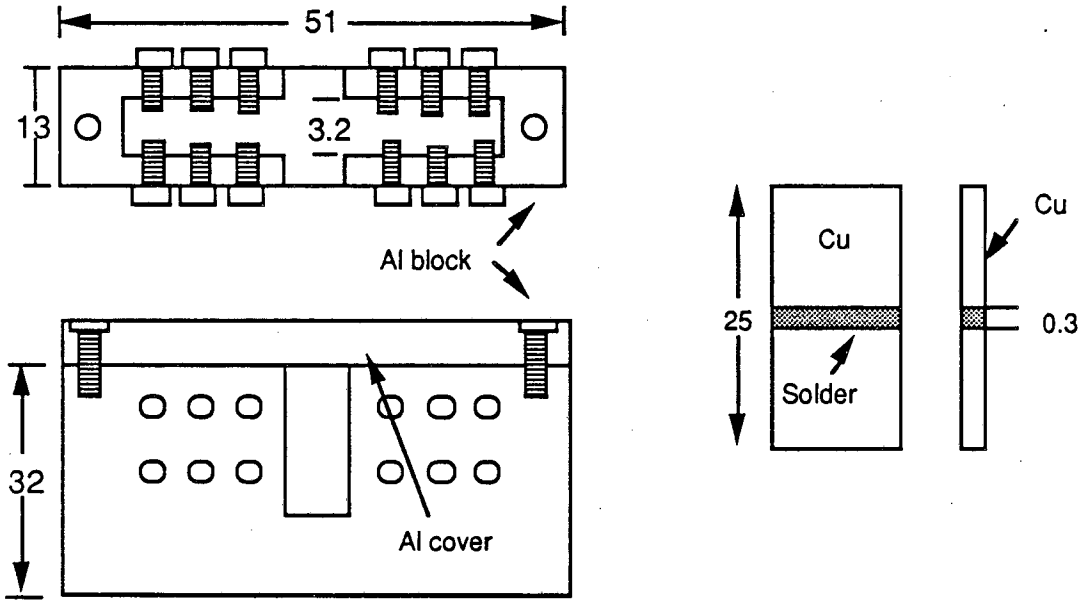
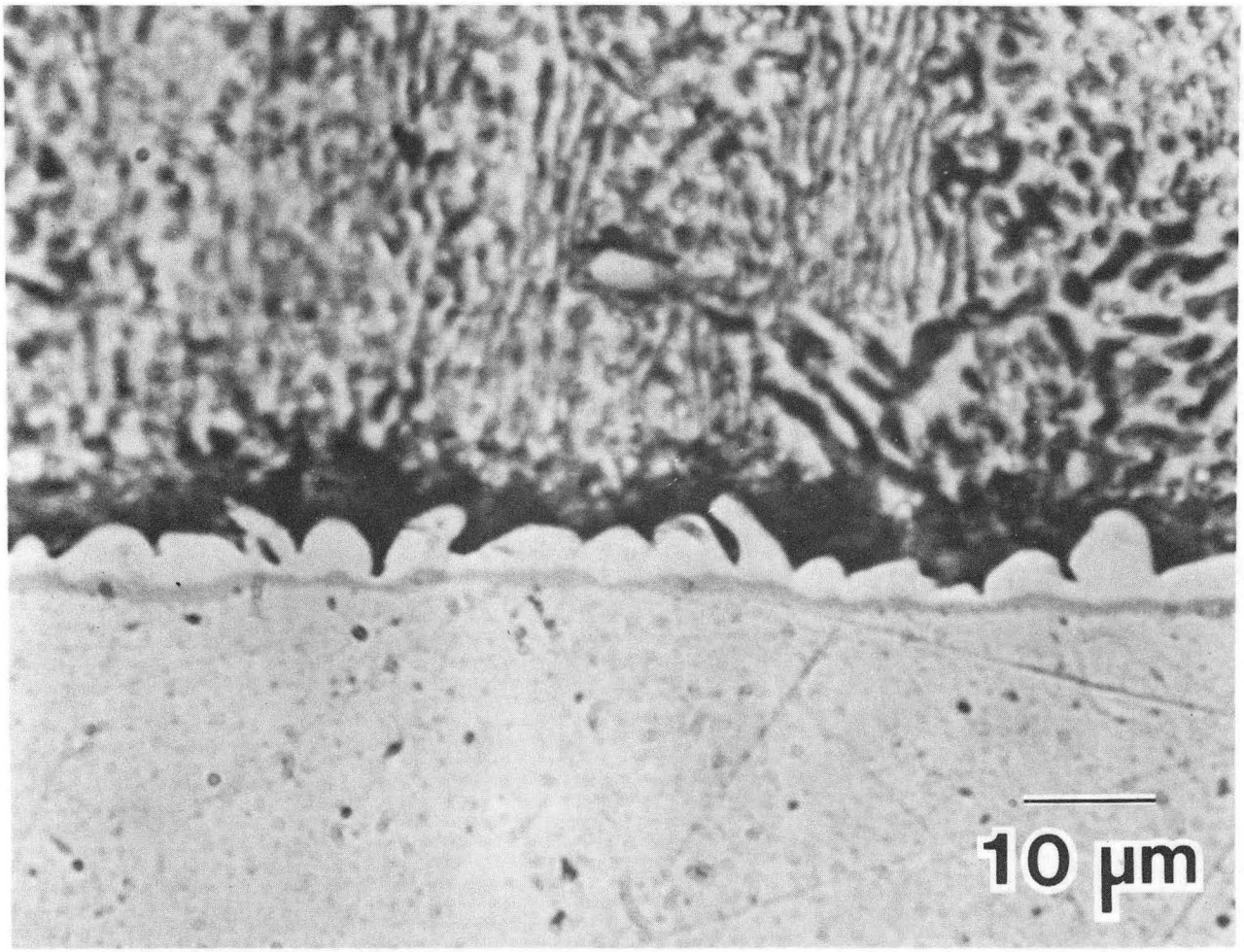
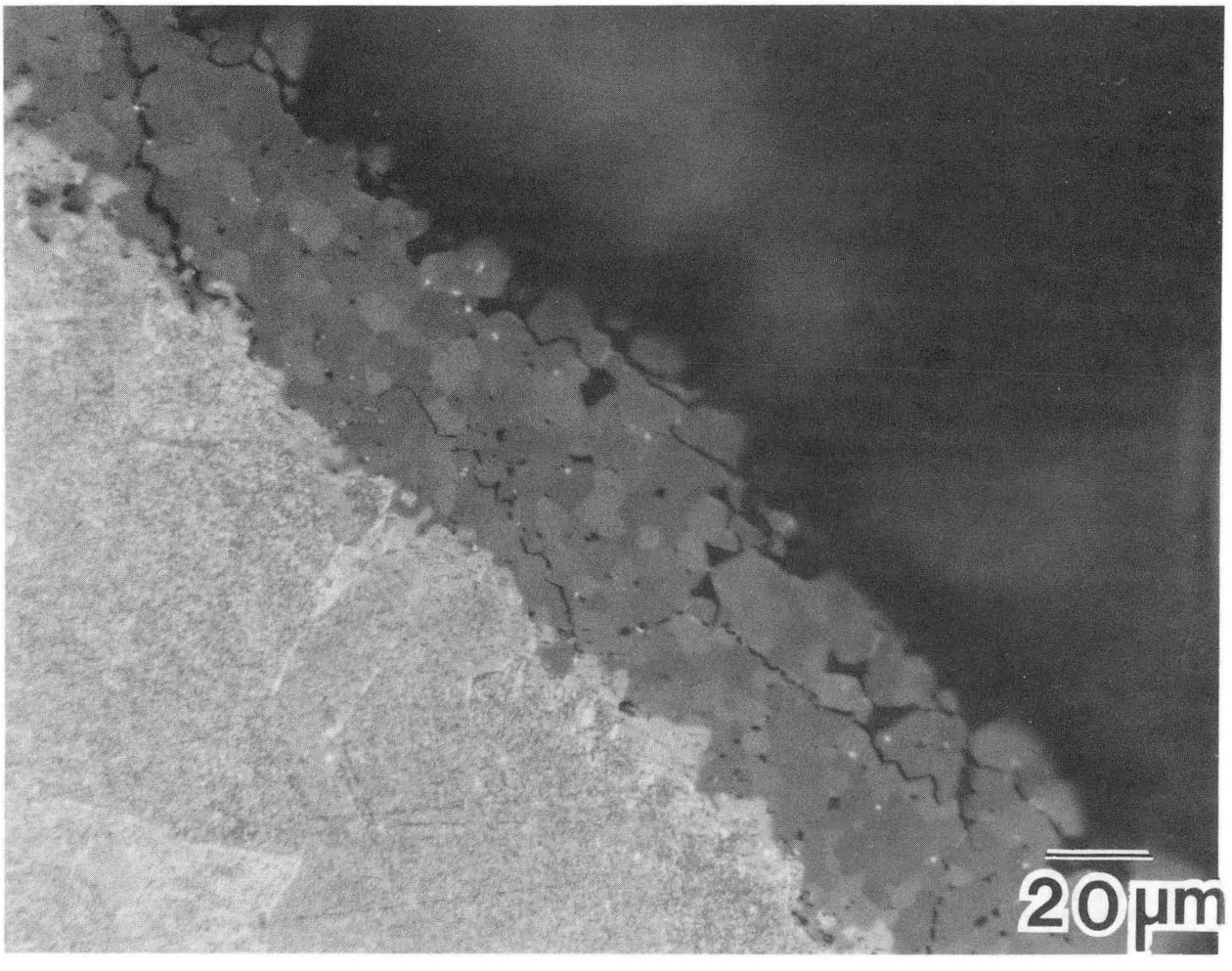


Figure 4.



XBB 871-439

Figure 5.



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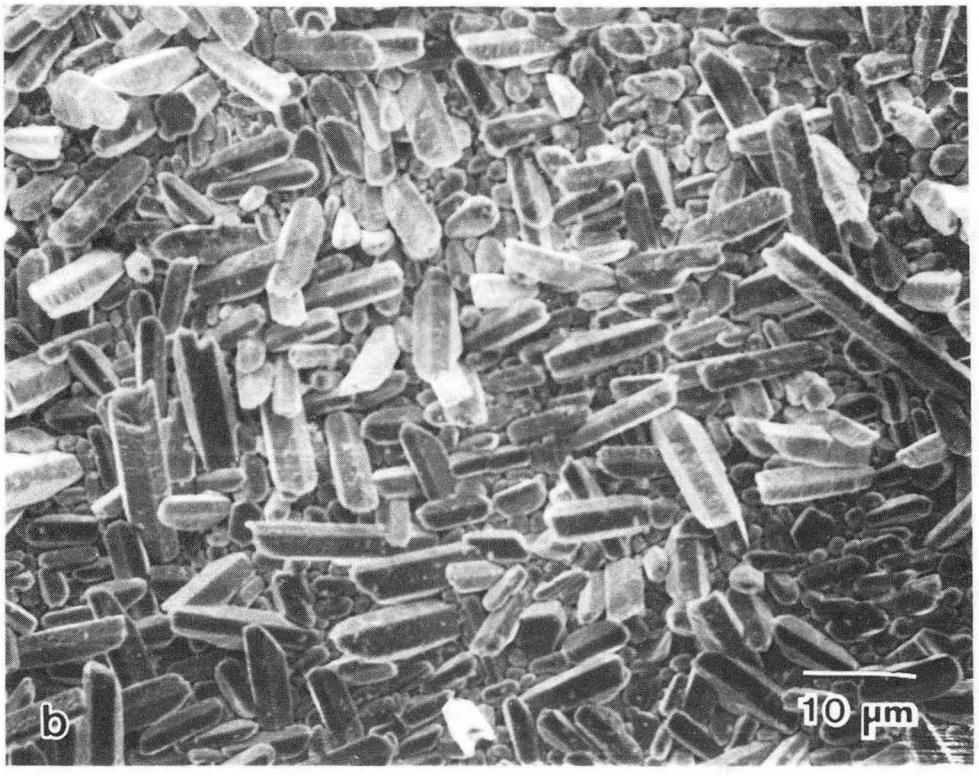
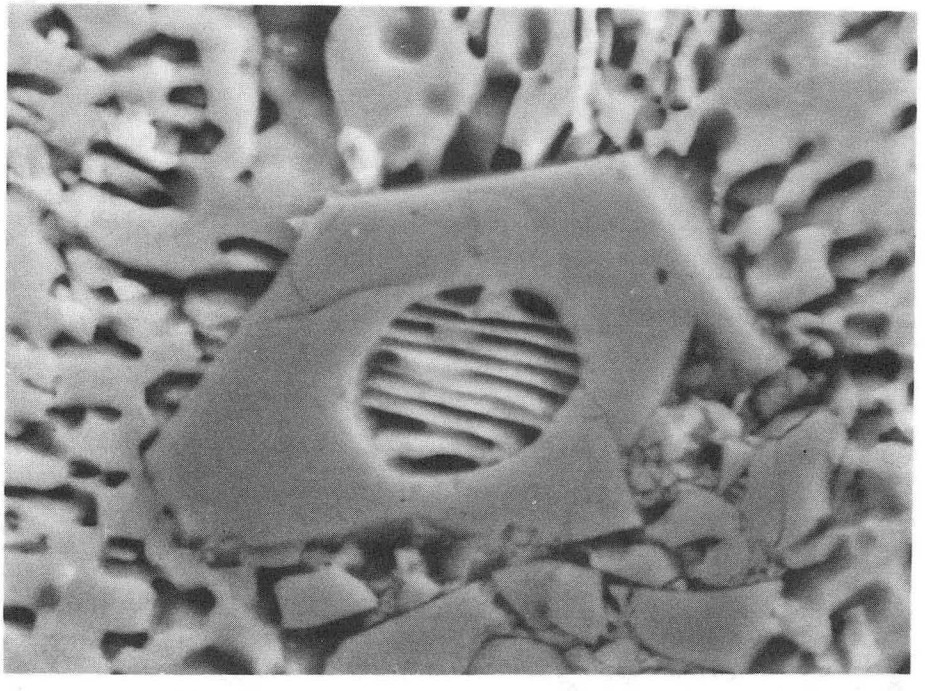


XBB 976-4404 (BI)

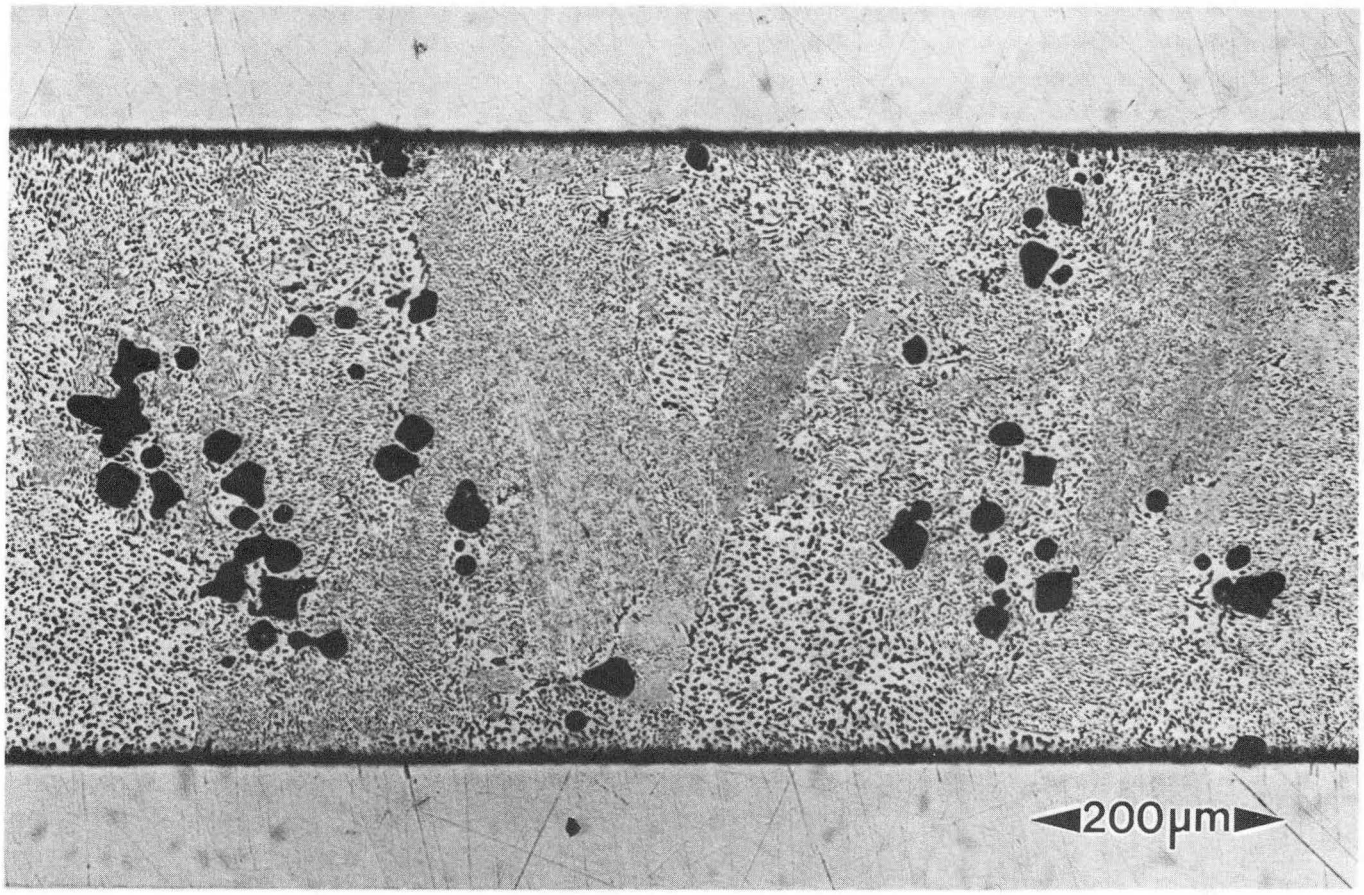
Figure 7.



2 μm

XBB 869-7448 (BI)

Figure 8.



XBB 887-7081

Figure 9.

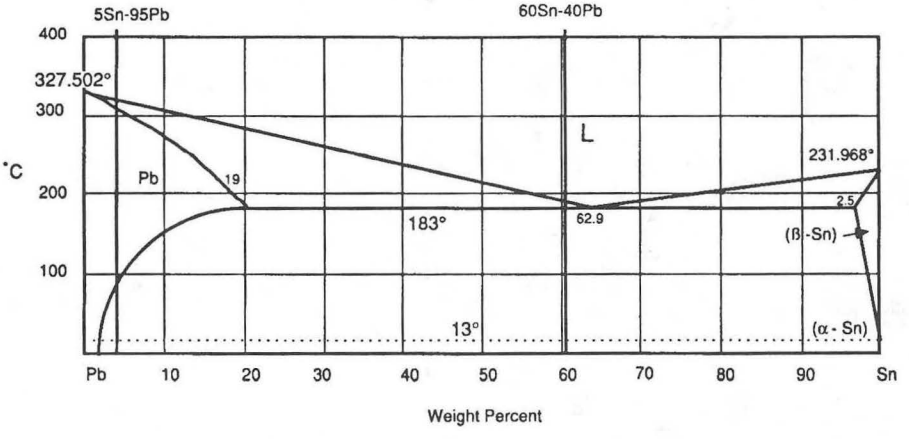
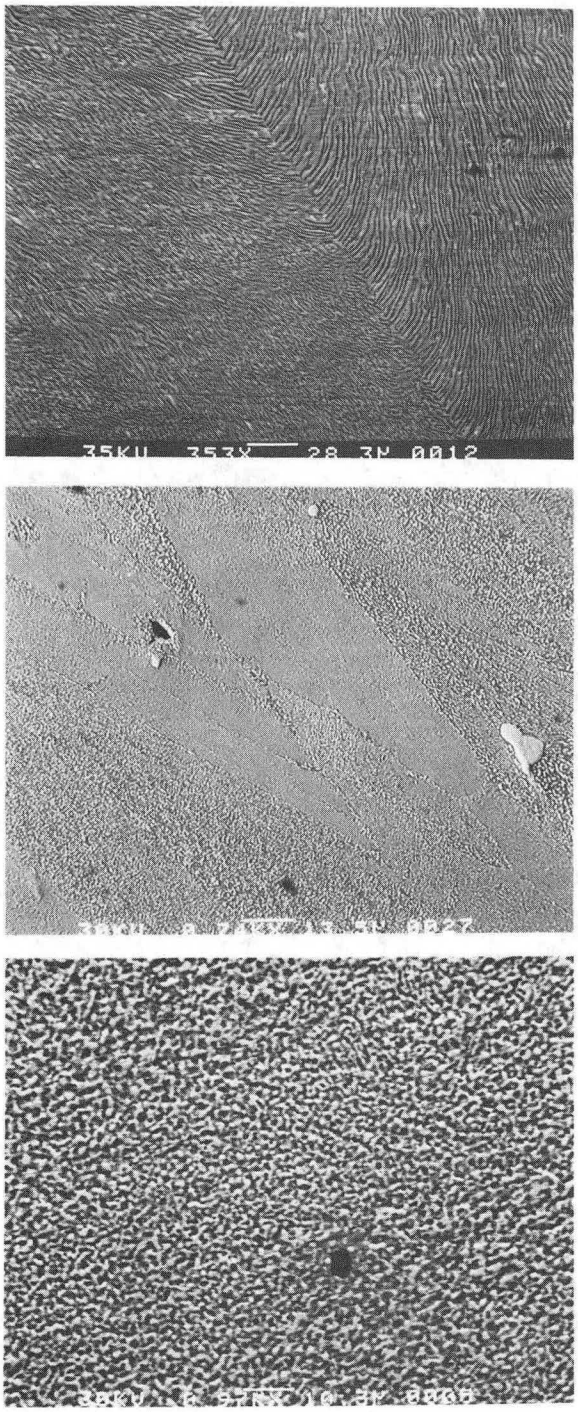
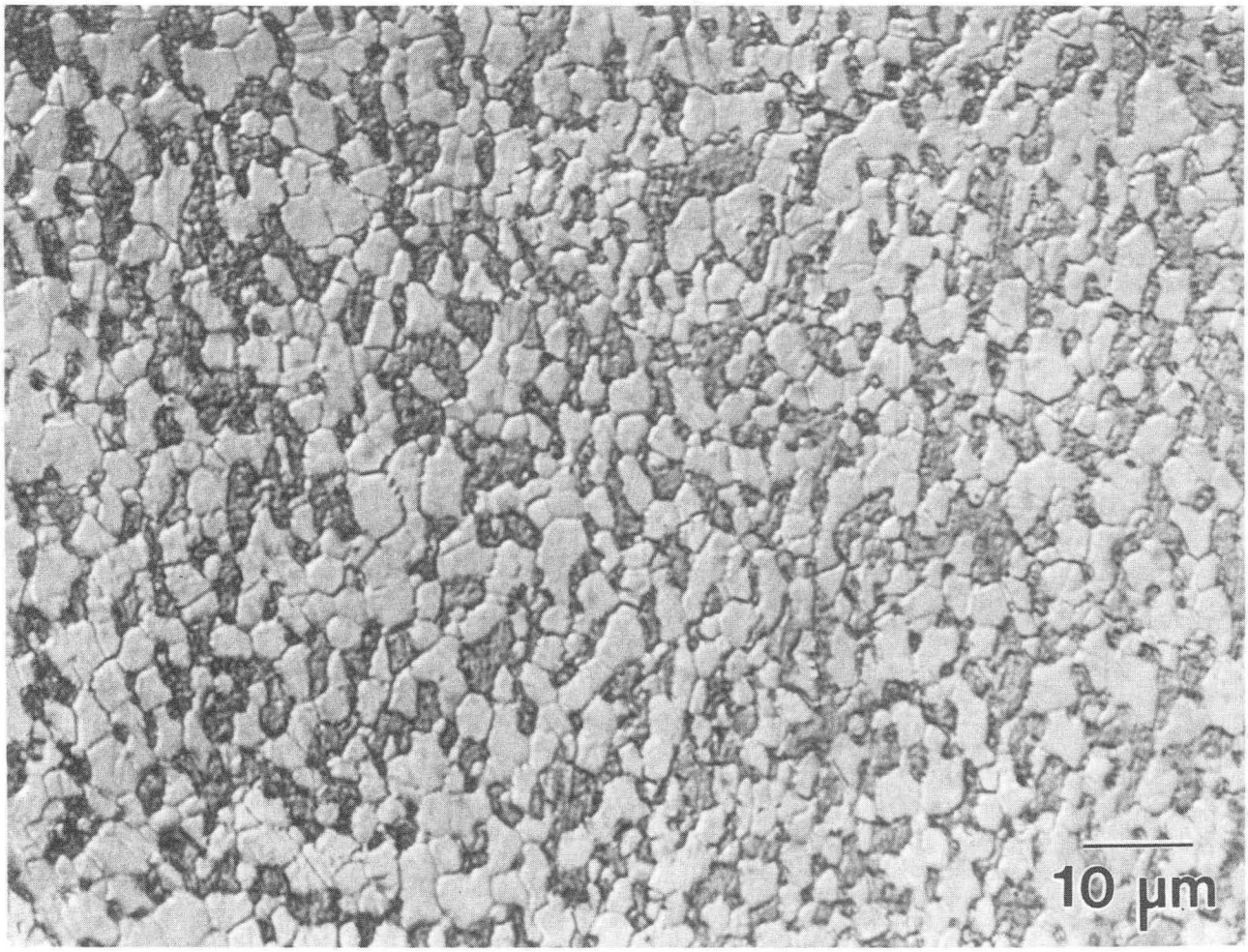

5Sn-95Pb
60Sn-40Pb
400
327.502°
300
L
231.968°
°C
Pb
19
200
2.5
183°
62.9
(ß-Sn)
100
13°
(α - Sn)
Pb
10
20
30
40
50
60
70
80
90
Sn
Weight Percent

Figure 10.



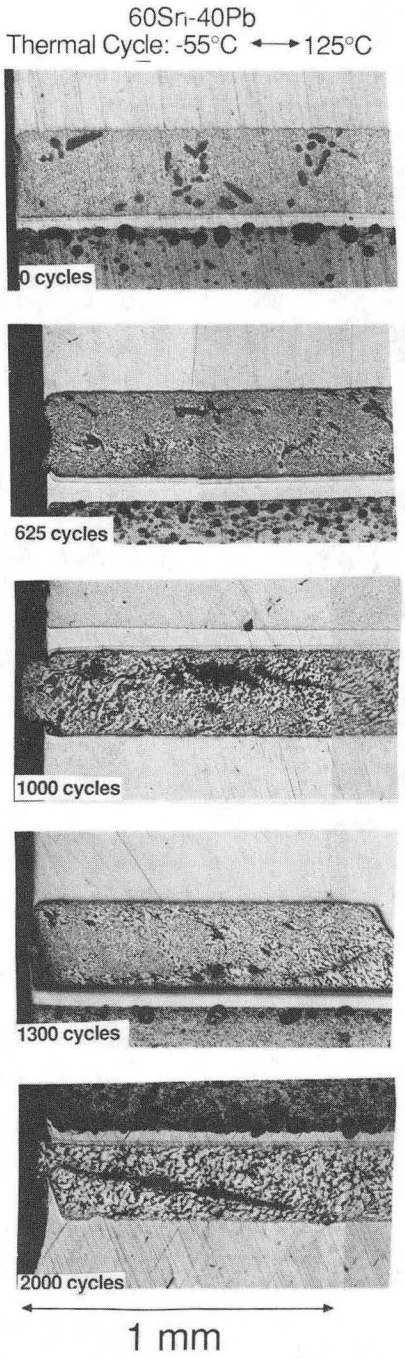
XBB 877-5538

Figure 11.



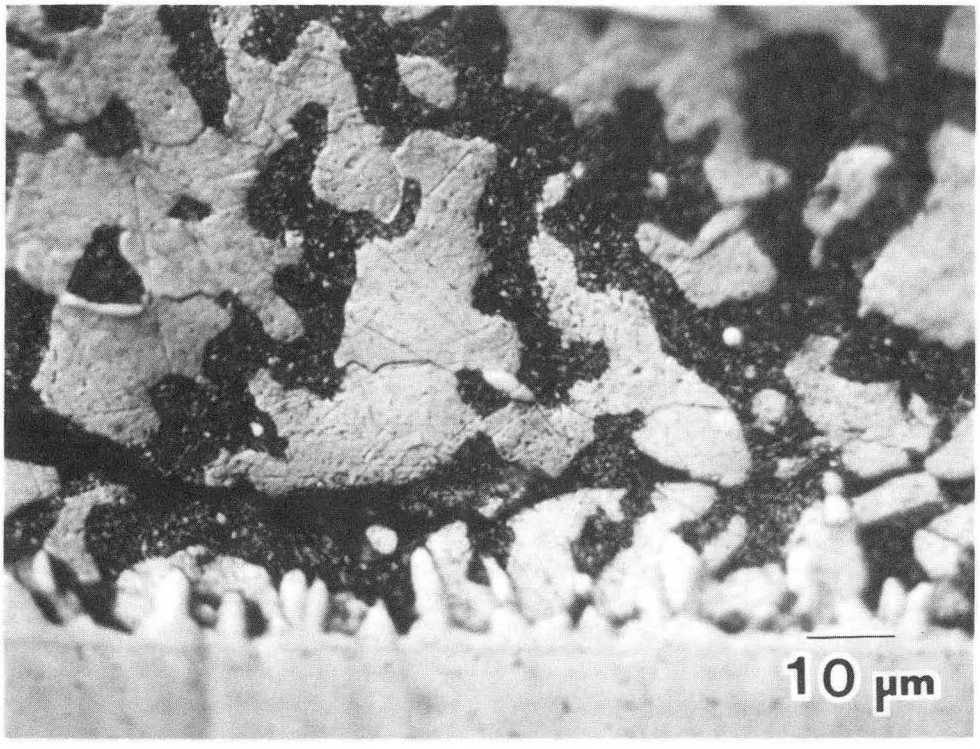
XBB 885-5304

Figure 12.



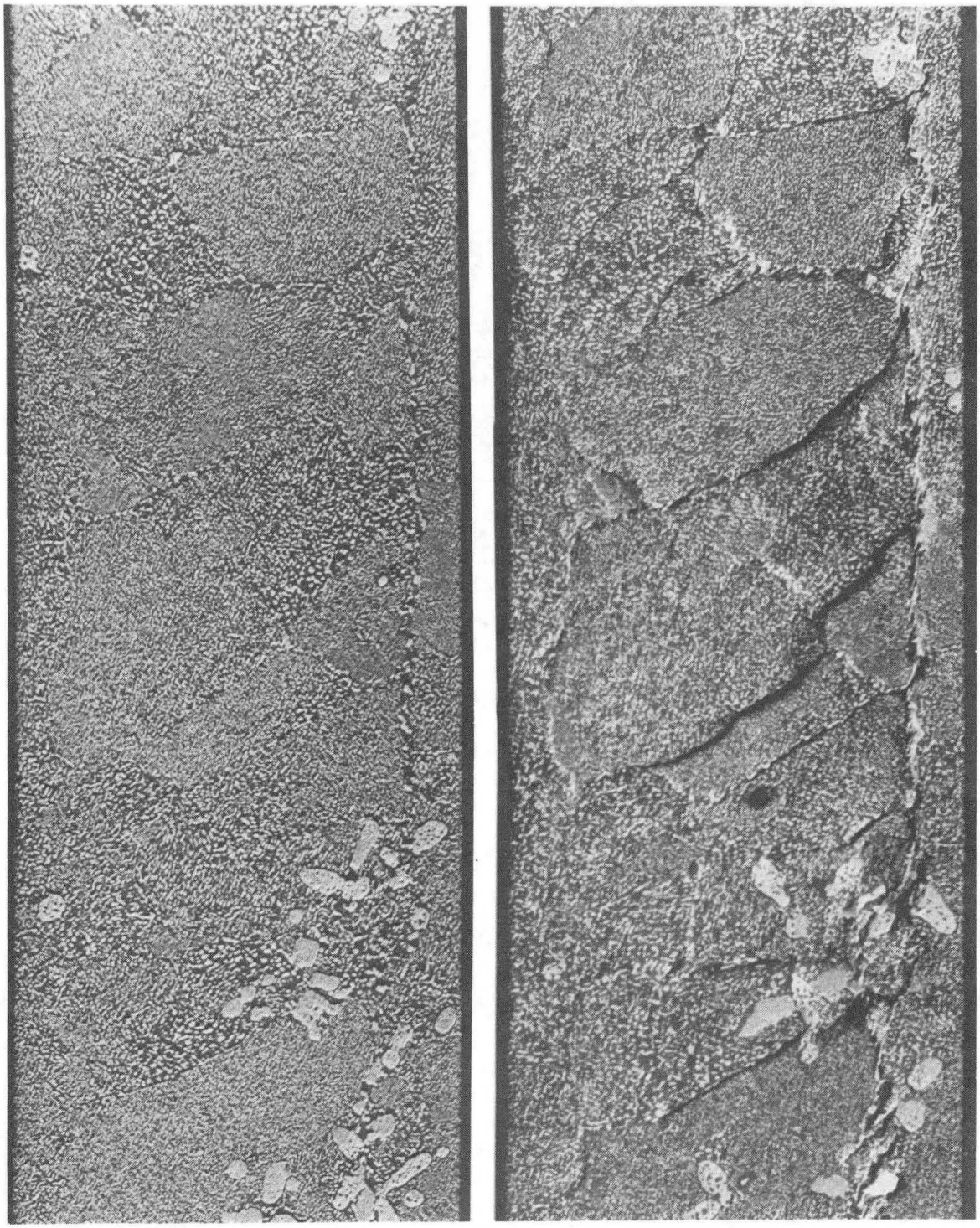
XBB 872-1250

Figure 13



XBB 860-10349

Figure 14.



◄0.02 mm►

XBB 882-1493

Figure 15

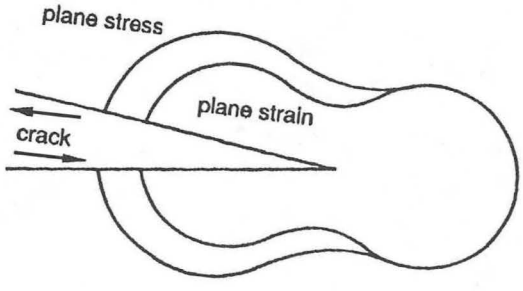
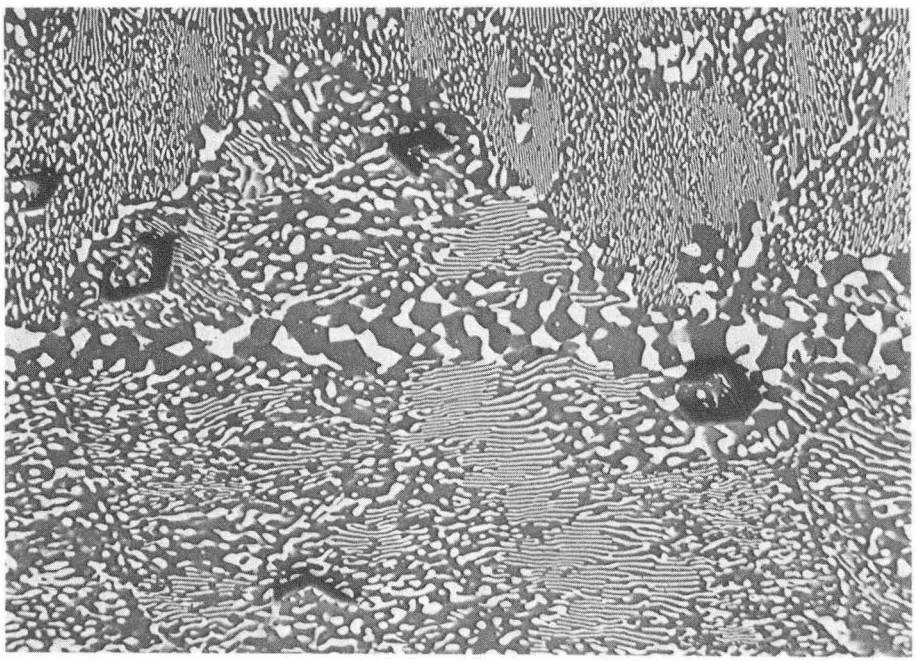


Figure 16.

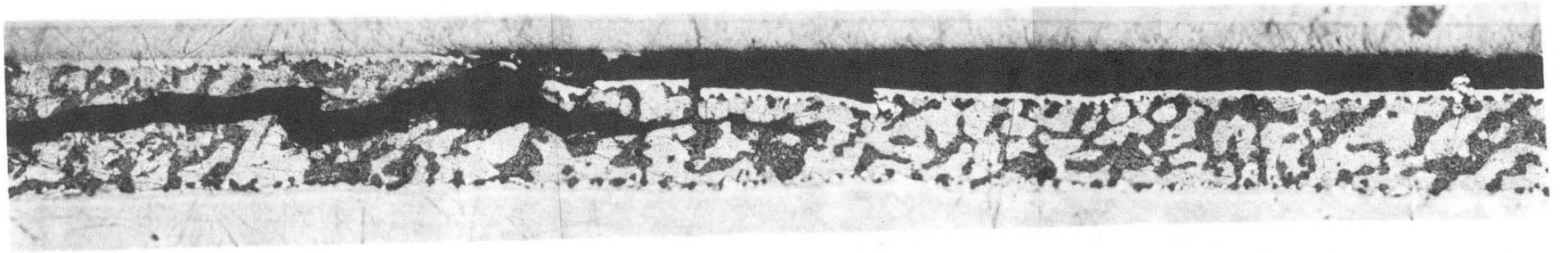


20 μm

XBB 881-277

Figure 17

60Sn - 40Pb
Thermal Cycle: -55°C ↔ 125°C



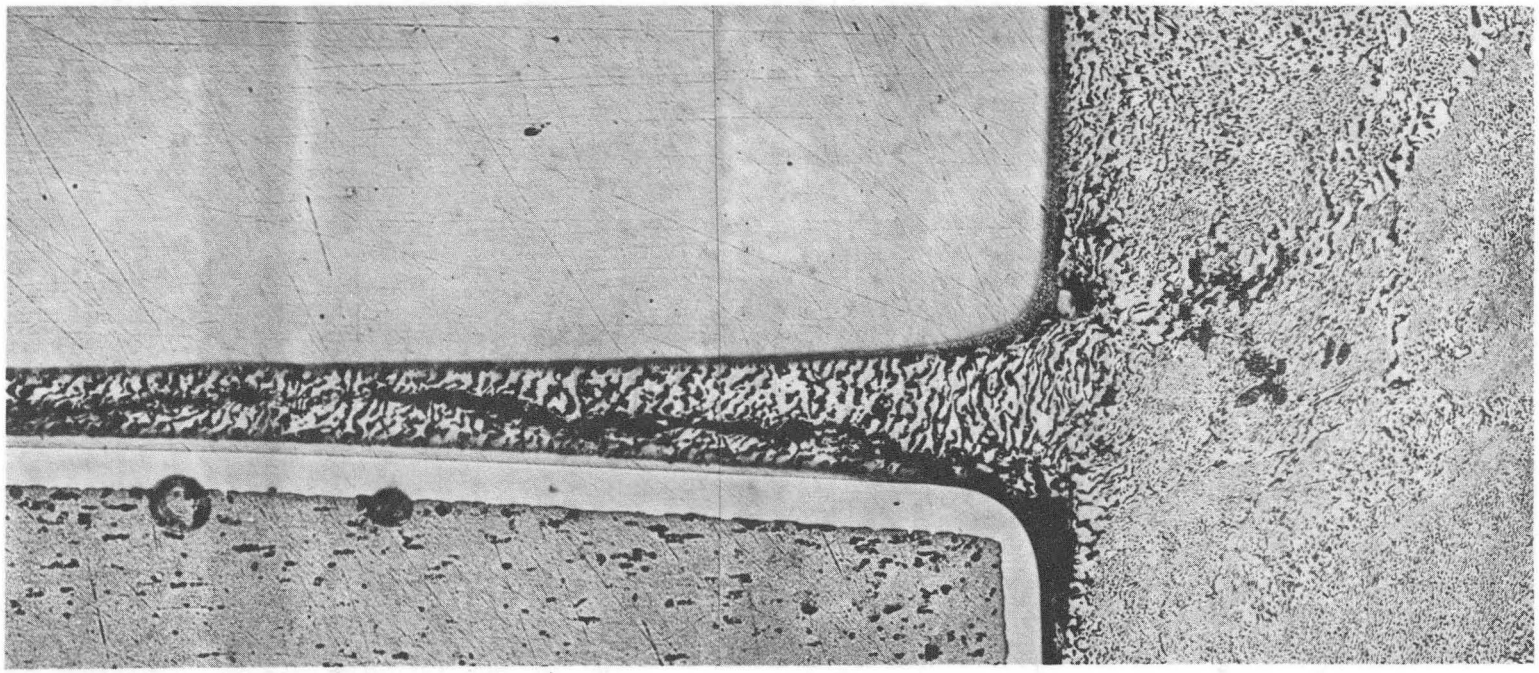
1500 Cycles

50 μm

XBB 879-7147

Figure 18.

60Sn - 40Pb
Thermal Cycle: -55°C ↔ 125°C



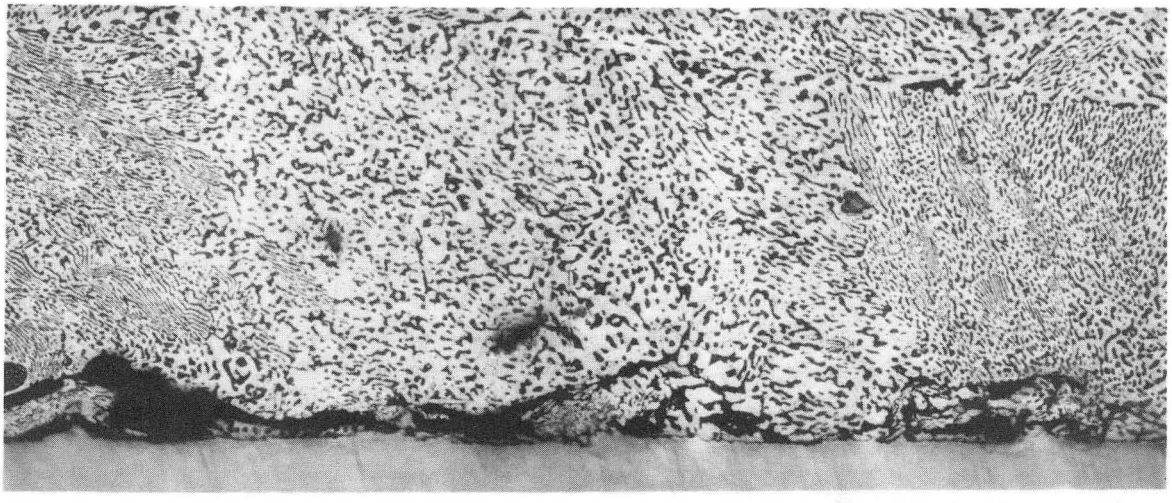
1500 Cycles

200 µm

XBB 879-7148

Figure 19.

60Sn-40Pb Tensile Orientation
Thermal Cycle: -55°C ↔ 125°C



250 Cycles

50 μm

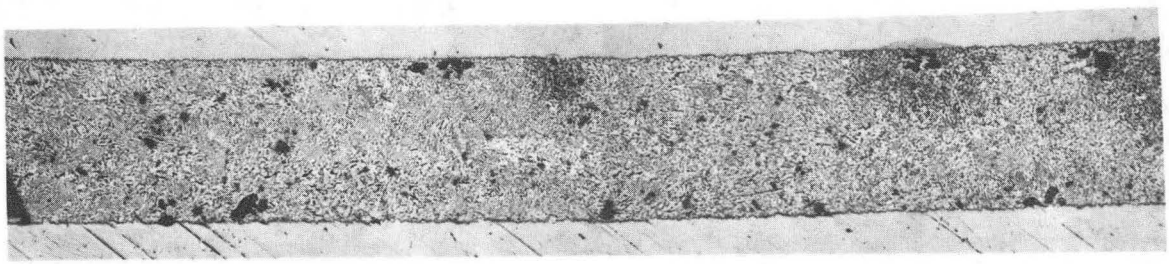
500 Cycles

20 μm

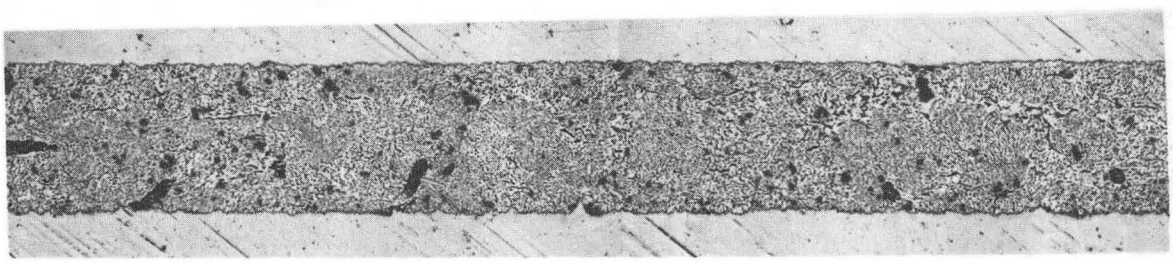
XBB 879-7150

Figure 20.

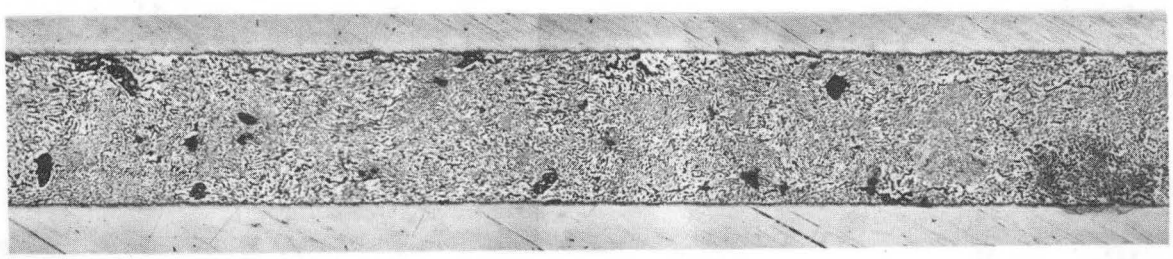
60Sn-40Pb Tensile Orientation
Thermal Cycle: -55°C ↔ 125°C



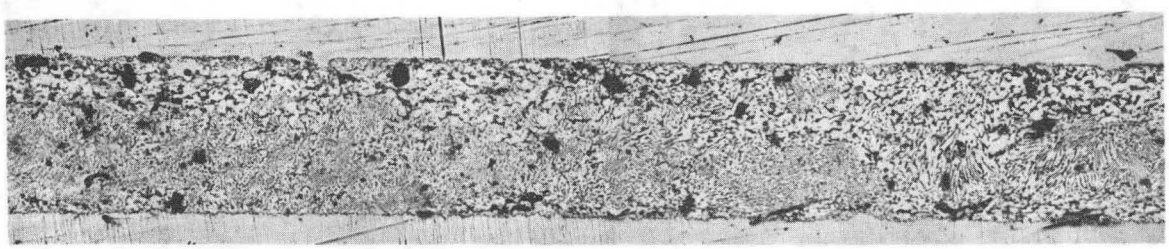
250 Cycles



500 Cycles



850 Cycles



1500 Cycles

200 μm

XBB 879-7152

Figure 21

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CENTER FOR ADVANCED MATERIALS
1 CYCLOTRON ROAD
BERKELEY, CALIFORNIA 94720