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SHAPE MEMORY BEHAVIOR AND MECHANICAL
CHARACTERISTICS OF SOME COPPER ZINC ALLOYS

Terrence Alan Schroeder

Department of Metallurgy and Mining Engineering
and Materials Research Laboratory
University of Illinois, Urbana, Illinois 61801

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I. INTRODUCTION

The formation of martensite in Cu-Zn alloys was initially observed by Greninger and Mooradian (1) while deforming a Cu-40wt%Zn alloy at room temperature and on cooling the same below a critical subzero temperature, M_s (martensite start). Since then martensitic transformations in this and other similar nonferrous alloys have been studied extensively. The martensite formed on cooling was found to disappear completely on reheating (thermoelastic martensite) (2), and some or all of the stress induced martensite (SIM) disappears on relaxing the stress, depending on the temperature (3) and magnitude of the stress (4).

The formation of martensite on cooling a stress free specimen, which initiated at M_s , continues until a temperature is obtained which corresponds to 100% martensite (M_f). Similarly, two temperatures can be identified which correspond to the start and finish of the reverse transformation; these temperatures are called A_s and A_f respectively. The M_s , M_f , A_s , and A_f temperatures are usually determined by electrical resistivity or dilatation measurements (5) and are in general sensitive to changes in composition. The variation of M_s with composition for binary Cu-Zn alloys is shown in Figure 1. A small or local composition change, such as around diffusion controlled precipitates, may

substantially influence M_s . The M_s is also dependent on the deformation history (6) and the existing strain state of the specimen (7).

Martensite formed on cooling in Cu-Zn alloys has been classified as either thermoelastic or burst (8). Thermoelastic martensite is formed when the temperature is lowered just below the M_s of the alloy and is reversible at an appropriate temperature above A_s . Burst martensite (8) is associated by some with the widening of the thin plates of the initially formed thermoelastic martensite upon further cooling. The nomenclature "burst" is used because of the apparent rapid thickening of the plates. Some refer to the reversible martensite that is formed under stress above M_s as thermoelastic (3) and assume that M_s is raised above the test temperature by stressing the specimen.

Stress induced martensite (SIM) can be formed below a certain temperature M_d . Deforming specimens above this temperature limit, M_d , only results in slip and not in the formation of martensite. The amount of stress required to induce martensite decreases as the temperature is lowered below M_d . SIM may decrease M_s provided that plastic deformation has been induced (6).

Recently, an interesting phenomenon has been reported in Cu-Zn (9) and other thermoelastic alloys: an apparent plastic deformation of the specimen, while in the fully martensitic state, recovers completely on heating the specimen above A_f (i.e., the specimen regains its original shape). This effect, first

discovered for Ni-Ti alloys (Nitinol) (10), is known as the shape memory effect. It is also referred to as the strain memory effect (11) and as the marmem effect (martensite memory) (12). This effect, which is related to the thermoelastic nature of the martensite, has since been reported in Au-Cd (13), Cu-Al (14), Cu-Al-Ni (15), Cu-Zn (9), Cu-Zn-Sn (11), Cu-Zn-Si (16), In-Tl (17) and Fe-Pt (9). Prerequisites for the shape memory behavior (12) have been suggested to be 1) that the martensitic transformation is thermoelastic, 2) that the parent and martensitic phases are ordered, and 3) that the martensite is internally twinned. The shape memory is a well defined phenomenon in that recovery occurs on heating from A_s to A_f .

The formation of SIM causes a departure from linearity in the stress-strain curve, and the reversion of SIM to the parent phase likewise causes a departure from linearity in the unloading curve; this behavior has been referred to as superelasticity (18) and is one type of pseudoelasticity. Pseudoelasticity is defined as any nonlinear elastic behavior. For some marmem alloys, Au-Cd (19) and Ag-Cd (20), two types of pseudoelastic behavior are known, superelastic and ferroelastic (shown in Figure 2). Superelastic (or rubberlike) behavior is the immediate recovery of nonlinear elastic strain on releasing the stress and is thought to involve the reversion of SIM (3) or the movement of twin boundaries (21).

Ferroelastic* behavior is seen when a specimen retains SIM after releasing the stress, but which reverts to the parent phase when an inverse stress is applied. Superelasticity is observed when the test temperature is above a ferroelastic limit (19). For the Cu-Zn and Cu-Zn-X martensitic alloys, the terms superelasticity and pseudoelasticity have been used synonymously (3).

In studying the shape memory it is necessary to perform experiments at temperatures below M_f , but the range of composition of the Cu-Zn alloys that have a practical M_f temperature (i.e., above liquid nitrogen temperature) is limited due to the sharp drop in M_s with increasing Zn content (shown in Figure 1). By maintaining a constant e/a ratio and alloying a small percent of third element to the Cu-Zn base, the M_s tends to rise for elements such as Ge, Sb, Sn, and Si (23). Alloying with a third element has a twofold effect: the yield strength of the parent phase increases, and the M_s temperature rises towards room temperature with additional solute. Alloying with Si and Sn as the third component causes the pseudoelastic effect to be very pronounced at low stresses and room temperature, which is just above M_s for the appropriate composition. The initiation of yield due to the formation of SIM was identified (3) in the stress strain curve by serrated yielding (Figure 3). The pseudoelastic yield stress decreases to zero by lowering the test temperature

*Nakanishi, et al (22), referred to materials as ferroelastic which have stress-strain curves with constant hysteresis on cycling, in analogy to ferroelectric and ferromagnetic materials.

to M_s (shown in Figure 4). The graphical display for yielding of single crystals (Figure 5) illustrates the competition between pseudoelastic yield and plastic flow yield as a function of temperature (3). The pseudoelastic effect and the shape memory behavior are shown to be complementary in Figure 6 and over a large temperature range exhibit a combined recovery of over 96% after strains of 4.5%.

Single crystal work shows large amounts of STRIPE (stress induced pseudoelasticity) in Cu-Zn (24) and Cu-Zn-X (3) alloys when no grain boundaries are present to increase local stress concentrations above the plastic flow stress. In polycrystalline material of Cu-Zn-Si alloys, plastic flow, which resembles slip deformation, has been seen at strains greater than 3%, and this deformation reduced the amount of STRIPE (e.g., only 80% of 4% strain is recovered). Polycrystalline specimens of Cu-Al-Ni also have reduced total STRIPE and shape memory behavior when compared to single crystal specimens (25). Oishi and Brown have shown that single crystal specimens recover more than 5.8% strain and that polycrystalline specimens recover less than 1.5% strain when the grain size is very small (about one third the specimen thickness).

Other conclusions by Oishi and Brown about Cu-Al-Ni (25), by Pops about Cu-Zn-Sn and Cu-Zn-Si (3), and by Eisenwasser and Brown about Cu-Zn-Sn (11) are similar. Pseudoelasticity is caused by stress inducing the nucleation of martensite which is reversible

upon removal of stress. The largest amount of pseudoelasticity is seen at test temperatures close to A_f for both single and polycrystals, and increasing the test temperature above A_f causes the single crystals to lose some degree of the STRIPE effect, but not nearly the amount found for polycrystals. At higher temperatures above A_f , the stress concentrations in polycrystalline specimens causes plastic flow to occur at lower stresses.

The change in yield stress, work hardening slope, and ultimate strength for single vs. polycrystals has been discussed for f.c.c. metals by Kocks (26), where plastic deformation by slip has been considered. Many martensitic alloys tested were in the form of single crystal specimens or flat specimens which had grains that continued across the thickness; the average grain size of the latter was from 1 to 5 times the specimen thickness and were called polycrystalline. The results (3, 11) from these coarse-grained specimens show an expected increase in yield stress and work hardening slope compared to single crystal specimens. The change in mechanical properties explains the reduction of the STRIPE effect in single vs. polycrystals; however, at strains of 4.5% the shape memory behavior has been reported to be unaffected by changing from single to polycrystals (11).

II. PURPOSE

As indicated in the introduction, most mechanical testing of the marmem alloys has been concentrated on the pseudoelasticity which is related to SIM. Little attention has been given to the deformation of the martensite formed on cooling below M_f , which is directly related to the shape memory effect.

The present work deals with the mechanical characteristics of the martensite formed below M_f in Cu-Zn alloys, as well as the recoverability of the apparent plastic deformation at low temperatures. A qualitative analysis of low temperature stress-strain curves and quantitative results of the shape memory effect are presented. Special attention has been given to the yield point, work hardening rate, strain to fracture, fracture surface and shape memory. Comparisons have been made between data obtained from test runs at room temperature and 77°K. The influence of grain size and specimen geometry on the mechanical characteristics has also been investigated.

III. EXPERIMENTAL

Early attempts to establish a small grain size in round tensile specimens showed that little control of grain size could be obtained by varying the annealing time, due to the extremely rapid grain growth that occurs at temperatures over 800°C. The round specimens could only be obtained with large or small grain size. The large grained round "bamboo" specimens (RB series), so named because of the appearance of the grain structure, had grains that extended across the diameter of the specimen. The round polycrystalline specimens (RPC series) had an average of 10 grains across the diameter. The specimens of both series were machined using the design shown in Figure 7a. The specimens have a cylindrical test area of 0.130" in diameter and a gage length that varied from 0.400" to 0.750". In order to obtain a different grain size a change was made in specimen design to flat specimens with cross-sections of 0.300" by 0.040" and a gage length of 0.750". This resulted in two more series of specimens: those with large grains that continued across the thickness of the specimen (FLG series) and those with smaller grains that continued only about halfway across the thickness (FSG series). The two flat specimen series were machined as shown

in Figure 7b so that the cross-sectional areas of the round and flat specimens were nearly the same.

The round specimens were made from two alloy compositions: 38.6 and 39.8 wt%Zn. The first alloy containing 38.6%Zn was melted in sealed quartz tubes using a commercially produced base metal of composition Cu-38.1%Zn and additional Zn to bring the composition to 38.6%Zn. The second alloy was obtained from a commercial source and had an analyzed composition of Cu-39.8%Zn with 0.05% impurities. Both the 38.1%Zn and 39.8%Zn alloys were air-melted in an induction furnace and poured into ingots weighing about 25 pounds.

The Cu-38.6%Zn alloy was swaged to 0.280" rods and machined into round tensile specimens. These specimens were then solution treated in the β -phase at 850°C for various times in order to obtain the proper grain size and then quenched in order to retain 100% metastable β' brass. The second alloy containing 39.8%Zn was cut from ingots into long rods 0.375" diameter, swaged to 0.250", machined, and solution treated as above. The flat specimens were all made of the second alloy and were hot rolled at 750°C to 0.045" thickness. After machining, the flat specimens were also solution treated and quenched. All alloys were chemically analyzed by the electrodeposition method.

The RB series was treated in sealed vycor tubing with 1 atm of argon for 18 hours at 850°C and water quenched to obtain single grains across the diameter. The "bamboo effect" limited the

grain size, but in all cases at least one grain continued across the test cross-section; however, all RB specimens also had regions with grain boundary triple points in the gage length. When the round specimens were graphite coated, treated in a lead bath at 870°C for 7 seconds, and quenched in iced-brine, the result was a polycrystalline specimen with 8 - 12 grains across the diameter.

The flat specimens were treated in a vertical tube furnace at 870°C for 1 minute and free-fall quenched into iced-brine to give the FSG series of specimens a 100% β' structure. For the FLG series the specimens were treated for 10 minutes as above, giving large grains that continued across the specimen thickness. The average grain size of the specimens was as follows:
RB, 3.0 mm; FLG, 1.2 mm; FSG, 0.7 mm; RPC, 0.25 mm.

The grain size was measured with a stereo-microscope after chemical etching in a 50/50 solution of hydrogen peroxide and phosphoric acid at 40-50°C. The chemical etch, which removed about 0.003" from the surface of the specimen, also eliminated surface α that frequently was seen after quenching because of dezincification at high temperatures. The specimens were then mechanically polished to eliminate any grooves at the grain boundaries caused by etching.

All mechanical testing was conducted on a Materials Testing System Model 810. This system uses an overhead hydraulic-driven piston whose position is controlled by either the load, strain, or stroke of the piston. An assembly consisting of load cell,

pull rod, upper grip, specimen, and lower grip was attached to the overhead drive piston. An electrically heatable pot of Woods metal was suspended below the fixed base plate of the MTS. The pullrod assembly was then made to pass through the base plate, and the lower grip was placed in the Woods metal pot to insure proper axial alignment of the specimen. A Dewar filled with liquid nitrogen, which completely enveloped the specimen and strain gage, insured a uniform temperature during the test.

An Instron Extensometer, G-51-17A, was used for strain measurements at both liquid nitrogen and room temperatures. Tests were run on flat specimens and some round specimens with the extensometer knife edges placed within the test section. Some round specimens, including all the RB series, had the knife edges placed outside the test section; for these the gage length of the specimen was used for calibrating the strain.

The test system was operated with load control so that the specimen was assured of no load when cooled to low temperature and constant load while warming to room temperature. Strain control was used for the tests at room temperature, and a tele-microscope was employed to observe specimens during deformation.

An FLG specimen was made from Cu-44wt%Zn alloy and annealed to obtain a 0.9 mm grain size. A specimen of this composition does not undergo a martensitic transformation when cooled to liquid nitrogen temperature and served as a standard. In

subsequent discussion the Cu-44wt%Zn alloy will be referred to as the stable β' specimen.

IV. RESULTS AND OBSERVATIONS

A. Stress-strain Curves below M_f

Specimens were tested at liquid nitrogen temperature, which was below M_f in all cases, and trends in yield point, work hardening slopes and strains to fracture are compared for different series of specimens. Specimen series (RB, FLG, FSG, and RPC) were differentiated on the basis of specimen geometry and grain size.

1. RB series

These specimens, having at least one grain continuing across the cross-section of the specimen, are referred to as bamboo type, due to the appearance of the grain structure, which resulted from the dimensions of the specimen limiting grain growth.

The stress-strain curves of the RB specimens tested below M_f (Figures 8 and 9) show much lower yields than the curve for stable β' brass (Figure 10). The stress-strain curves for the 38.6wt%Zn and 39.8%Zn alloys show no apparent differences in mechanical behavior even though the M_s varies by 80°C (-20 and -100°C). Yielding occurs almost immediately after loading (5-10 ksi) and work hardening portions of the curves for the

RB specimens are concave upward with an inflection point varying from specimen to specimen. Strains to fracture were larger than 15%.

2. FLG series

This series of specimens had grains which continued across the thickness so each grain had two free surfaces.

The stress-strain curves for the flat specimens (as shown in Figures 11 and 12) show a small elastic range with yielding beginning at 7-10 ksi. Work hardening regions also show a linear portion not observed in the RB series. This linear portion continues for 3-5 percent strain, and then the curve begins a concave upward trend until grain boundary separation is evident. The strains for grain boundary separation, whether local or catastrophic, are much lower than for the RB type. The FLG strains to fracture were 6-9% at low temperature.

The stress-strain curve of stable β' brass shows a high yield stress of 35 ksi and a work hardening region similar in shape to a typical curve from a polycrystalline tensile test. Strain to fracture was 16%.

3. FSG series

The specimens in this series had smaller grains, which on the average continued halfway across the thickness of the specimen. The smaller grain size resulted in the loss of a free surface compared to the FLG specimens. This corresponds

essentially to the loss of another degree of freedom for the grains.

The specimens exhibited stress-strain curves (Figure 13) with similar elastic proportionality, but higher yield stress than the FLG series. The work hardening region is more linear and even slightly concave downward. A lower strain to fracture (5%) was found for the smaller grained specimen, compared with the FLG series. The higher yield of the FSG series and more linear work hardening slope make for a larger stress to fracture than for the FLG series even though FLG specimens usually had larger strains to fracture.

4. RPC series

This series was machined similarly to the RB series, but was given a shorter anneal to allow a smaller grain size. The RPC series was tested with the extensometer grips placed within the gage length, thus measuring local strain rather than the total strain over the gage length. These stress-strain curves showed some scatter (Figure 14) which could be reduced by placing the extensometer knife edges outside the gage length of the specimen (Figure 15). Some difference is seen in specimens RPC 1, 2, and 3 and RPC 4 and 5 (shown in Figure 16). Specimens 1, 2, and 3 had a 0.400" gage length with a collar and a Zn content of 38.6%, while the others (4 and 5) had a 0.750" gage length with no collar and 39.8%Zn.

All the RPC specimens had similar stress-strain characteristics, even though slight variations in the curves were seen due to the extensometer length and specimen design. The yield stress was the highest of all the series, 15-20 ksi, and the work hardening regions were the most linear of all the series, showing a definite concave downward trend. The work hardening slopes were also the steepest of any of the series, and strains to fracture were around 7%.

When polycrystalline specimens were held at a constant stress at low temperature, the observed strain was found to increase with time and asymptotically approached a certain value. The extra strain at constant stress is referred to as "creep" and is shown in the curves of Figure 17.

B. Stress-Strain Curves above M_s

These test runs were carried out at room temperature which was from 50°C to 110°C above the M_s of either alloy used. The specimens had the same design as described in Section I and were categorized in the same manner, i.e., RB, FLG, FSG, and RPC. The yield stress, general curvature of the work hardening region, and general shape of the unloading curve will be compared.

1. RB series

At room temperature the extensometer knife edges were placed outside the specimen gage length, which was used to calibrate the strain measurement. Making the runs in strain control allowed

the stress to decrease, unlike load control where strain increased rapidly in order to compensate for the stress relaxation that accompanied phase transformation.

Three general segments can be distinguished in the stress-strain curves of the RB series (Figure 19): 1) linear elastic, 2) plastic, and 3) pseudoelastic. The elastic limit (25-35 ksi) is generally lower than the normal yield point of stable β' brass (35 ksi). The elastic modulus is about 6×10^6 psi for all the specimens tested at room temperature.

The plastic region that follows the elastic region is mostly linear and extends to a strain which depends on the specimen being tested. Within the plastic region the slopes remain constant until serration occurs, the latter causing a decrease in load and sometimes even a negative slope. Upon increasing the strain the stress-strain curve concaves upward until a positive slope is reached and then a typical stress-strain curve is produced. Various strains to fracture (9-14%) were obtained from the RB series at room temperature. The unloading curves show large amounts of the pseudoelastic effect with strain recovery beginning after unloading 5-15 ksi (reverse yield). This "reverse yield" occurs at a stress where the unloading curve changes to a low slope and allows a larger amount of strain to be recovered than if the unloading curve returned to zero load in a typical elastic manner.

2. FLG series

As with the low temperature runs this series of specimens had grains which continued across the thickness of the specimen. The stress-strain curves at room temperature (Figure 20) show larger yield stresses and more of an elastic range than the RB series. A dip in the stress-strain curve is usually seen after the yield at room temperature, after which the work hardening region becomes similar to the curve for stable β' brass. Strains increase until grain boundaries separate and result in failure at about 10%.

The recoverable strain due to pseudoelastic martensite (STRIPE effect) of the FLG series produced a markedly larger "reverse yield" than that of the RB series (Figure 19), being 20-25 ksi and 5-15 ksi respectively.

3. FSG series

These specimens, with the average grain continuing only halfway through the thickness, have stress-strain curves at room temperature (shown in Figure 21) with yield stresses slightly lower than that of the stable β' brass. No "dipping" of the stress-strain curve is observed immediately after yield as with the RB and RLG series. The slopes continually decrease, resulting in a concave downward curve. The strain to fracture (up to 16%) for the FSG series at room temperature was larger than that for the FLG series (6-12%), and the stress to fracture was also higher -- 30 ksi for FLG and 45 ksi for FSG.

The RB and FLG specimens show pseudoelastic strain recovery marked by a "reverse yield", but the FSG specimens display an unloading curve with a concave upward shape and no "reverse yield." By reapplying stress, the loading curve, which is concave downward, completes a little hysteresis loop and has an average slope close to the elastic modulus of the specimen (as shown in Figure 21).

4. RPC series

The round polycrystalline specimens (0.750" gage length) were tested under strain control with an extensometer length of 0.400". The curves were very similar to those of the stable β' brass when tested at low temperature, with respect to elastic limit, work hardening slope and ultimate tensile strength, even though yields were slightly lower than those found for the stable β' brass. The RPC series had strains to fracture in excess of 20% (Figure 22).

The RPC series, compared to the RB and FLG series, shows stress-strain curves with larger yields and more linear work hardening regions. In comparison to the FSG series, the RPC series has a similar yield stress and work hardening slope, with larger strain to fracture at room temperature. The unloading curves for both the RPC and FSG series have no "reverse yield" but do have a small hysteresis that indicates that some of the SIM reverts to the parent phase.

C. Fracture Characteristics at Low Temperature

The fracture surfaces were examined after failure, and the series were compared with reference to the amount of ductile area on the fracture surface, where the failure began, and general appearance of the surface. Gage length regions near the fracture surface were also examined for evidence of deformation.

1. RB series

The failure of the RB series at liquid nitrogen temperature was catastrophic as soon as the first grain boundary initiated a crack. The failure always left a grain boundary separation that extended sharply across the specimen. Very little ductile area was observed, and some grain boundary irregularities resembling small precipitates were observed on the fracture surface by optical microscopy.

2. FLG series

At low temperature the FLG series showed cracks in specimens at grain boundaries before catastrophic failure occurred. Usually the separation of one boundary led to local ductile tearing at triple points and separation of other nearby grain boundaries. Still, the mode of failure was brittle.

The stable β' brass specimen that had the same dimensions and grain size as the series was pulled to fracture at liquid nitrogen temperature. This alloy exhibited a transgranular

ductile failure at a 45° angle and 16% strain. The very adherent grain boundaries showed little separation.

3. FSG series

At liquid nitrogen temperature, these specimens separated at grain boundaries, which initiated a crack. Then as the crack proceeded the fracture became more ductile, more so than the FLG specimens. "Brittle" grain boundary separation usually initiated in the middle of the test area; as the separation of the specimen continued, the failure became more ductile towards the edge.

4. RPC series

The round polycrystalline specimens exhibited brittle failure with grain boundary separation at the center of the cross-section and also showed some ductile tearing on grains toward the periphery of the cross-section. The grains toward the periphery that showed some ductile failure were sheared at 45° to the tensile axis. The clear majority of the fracture was brittle, but of all series tested at low temperature the RPC specimens had the most prevalent amount of ductile failure.

D. Fracture Characteristics at Room Temperature

For both alloys room temperature was above M_s and below M_d , and the runs at room temperature could be made under strain control with the extensometer measuring the strain. Specimens are compared as in previous section.

1. RB series

The large grains that separated at the boundaries at room temperature showed a brittle fracture with evidence of slip on the fracture surface. The latter was not observed on the fracture surfaces at 77°K.

2. FLG series

The failure of the large-grained flat specimens was generally brittle, as in the low temperature case, but the amount of ductile failure at room temperature was larger than that at low temperature and was more than that found for the RB series at room temperature (which was practically none). The ductile area observed in the FLG series at room temperature was near the edges of the specimen, and the brittle grain boundary separation appeared in the middle region.

3. FSG series

At room temperature there was a drastic change in failure mode comparing the FLG and FSG series. The FSG series showed a high percentage of ductile failure which fractured at a 45° angle across the width of the specimen. In comparing the fracture surfaces of the two flat specimens at room temperature, the FSG series had nearly as much ductile area as the FLG series had brittle area.

The failure mode of the FSG series at room temperature was definitely more ductile than that of the FSG series at low

temperature. Except for a few separated grain boundaries, the ductile failure of the FSG specimens resembles the failure of the stable β' brass specimen at low temperature.

After fracture is initiated, the crack propagates to a triple point where it either continues between two grains as a boundary separation or stops at triple point and causes plastic deformation. It was noted that the deformation at triple points is much greater in the FLG series compared to the FSG series.

4. RPC series

The failure of the RPC series at room temperature resulted in a ductile fracture with brittle initiation by grain boundary separation in the center. The ductile area is in the majority, but the brittle region in the center is easily identified. By comparison to the FSG series at room temperature the RPC series is less ductile, but in comparison to the RPC series at low temperature the room temperature failure of the RPC series is more ductile.

The small-grained specimens of the RPC series when fractured at room temperature showed a ductile failure that resembled the well known cup-cone fracture, where a brittle portion is surrounded by a ductile perimeter sheared at a 45° angle. Some grains at the outer edge of the cross-section did however show brittle fracture.

E. Shape Memory Recovery

The MTS system, operating in load control, insures that there is no stress on the specimen while cooling or heating and, in some cases, maintains a constant stress on the specimen during heating. At low temperature, below M_f , the specimens were stressed to give some deformation, then unloaded and heated for observation of the shape recovery at zero load.

The percentage of recovery of the specimen due to the shape memory is calculated as the difference between 100 and the percentage of apparent plastic strain not recovered on heating above A_f . This percentage of irrecoverable strain is found as the remaining strain at zero load, divided by the total strain at zero load and low temperature, times 100. To account for the difference in gage factors of the extensometer at the two temperatures, a correction term of 0.95 is applied to the percentage of irrecoverable strain.

1. RB series

The recoverability of the RB series was extremely high. 100% recovery was observed at 7.2% strain and 23 ksi stress. Since the stress-strain curve was different depending on the particular specimen being tested, some curves showed a higher stress than others for the same strain. At a plastic strain of 8% and maximum stress of 52 ksi, the recovery of specimen RB-4 was only 92%, while at a plastic strain of 10.9% and maximum stress of 38 ksi the recovery was 94% for RB-1 (Figure 8).

In the RB specimens local areas in the test section showed grains that did not continue across the diameter, resulting in triple points which showed grain boundary deformation after strains of less than 1%. Slight amounts of grain boundary deformation were observed even in cases of 100% recovery. More grain boundary deformation was observed in specimens showing less recovery, and plastic deformation was observed in specimens showing less than 95% recovery (which occurred when stresses were higher than about 35 ksi).

2. FLG series

The recoverability of the FLG series was markedly lower than that of the RB series (Table 1). The recovery ranged from 99.6% at 2.5% plastic strain and 18 ksi to 96.2% at 5.3% plastic strain and 23 ksi.

Grain boundary deformation was always observed after recovery by shape memory, with more being found after lower percentages of recovery, and with less recovery, plastic deformation resembling slip was seen at triple points.

Straining of the flat specimens at low temperature caused grain boundaries to separate without necessarily causing catastrophic failure. This reduced the amount of recoverable strain considerably and was observed in cases of low recovery.

A specimen was loaded at different rates (Figure 23) to see if the unloading-loading hysteresis would change and to see if the average slope of the unloading curve would vary with the

loading rate. The change in rate from 10 ksi/min to 100 ksi/min had no effect on either the hysteresis or the average slope.

3. FSG series

The FSG specimen showed a recovery that was similar to the FLG series. The recovery (Table 1) is similar to that found for the FLG specimens that showed higher yields. In the specimens that showed low recoverability, grain boundary deformation was observed and triple points showed areas of plastic deformation.

4. RPC series

The 38.6 and 39.8wt%Zn alloys were used to make different specimens in the RPC series, and the stress-strain curves (Figure 16) indicated that the difference did not necessitate recategorization of specimens. The gage length and collar design also varied, and the effects on the stress-strain curves were negligible. The recovery of the specimens showed only slight variations after comparing the different types of RPC specimens.

The recoverability of the RPC series was the lowest of all the series tested. Recoveries ranged from 100% at 2.6% plastic strain and 39 ksi to 72% at 4.2% plastic strain and 56 ksi (Table 1).

F. Pseudoelastic Strains (STRIPE) above M_s

At room temperature (above M_s and A_f) the specimens were strained and then unloaded. The SIM that is formed above M_s by stressing is at least partially recovered by releasing the stress. This pseudoelasticity, so called because of the nonlinearity of the stress-strain curve, is measured as the strain recovered from the maximum applied stress to zero stress.

1. RB series

The series was tested in strain control and gave a scattered set of curves, as previously noted. The specimens were each unique in grain configuration, so some variation was expected. After a drop of more than 5 ksi the strain decreased quickly for a few percent. After this the slope returned to a larger elastic modulus value as the load went to zero (Figure 19).

The percentage of pseudoelastic strain was much lower than that of the shape memory. The first loading cycle of the room temperature runs gave 30-50% pseudoelasticity, but the second cycle showed serrated yielding much earlier and exhibited up to 80% pseudoelasticity (Figure 24).

2. FLG series

The room temperature runs on the FLG series were in load control and showed small amounts of pseudoelastic strain. The pseudoelastic strain recovered sometimes showed an amount of quick strain recovery and at other times just a change in the slope of

the unloading curve from that of the loading curve (Figure 20).

The second cycle of the FLG series at room temperature (Figure 25) showed much more pseudoelastic strain, as did the second cycle in the RB series tested at room temperature.

3. FSG series and RPC series

These series showed no drop in the stress-strain curve after yield and no "reverse yield" in the unloading curve at room temperature. Both series had unloading curves which were uniform, with large slopes, and appeared similar to the unloading curves of the martensitic specimens.

G. Effect of Placing RB Series in Compression

At liquid nitrogen temperature an RB specimen was deformed to a large strain (13%) then unloaded. The unloading curve was typically without "reverse yield", with a large curvature and some strain recovery when unloaded the last 5 ksi. The specimen was then put into 8 ksi compression and recovered nearly all the deformation strain. The curve in compression was extremely smooth, with a low work hardening slope (Figure 26).

One RB specimen was strained to 9% at room temperature and unloaded. The unloading curve was smooth and did not show the expected "reverse yield", but had a low average slope indicating pseudoelastic recovery. When put into a compression of only 5 ksi, a quick strain recovery occurred for about 2.5% strain

(Figure 27). This recovery produced a very low slope which was a continuation of the unloading curve. Thus the stress-strain curve became similar to the unloading curves of the other RB specimens at room temperature, but the "reverse yield" appears to have increased by 30 ksi. This caused the unloading curve to be shifted lower and the quick recovery portion to be in a compression state (while the other unloading curves had the quick recovery portion in the tension state).

V. DISCUSSION

A. Mechanical Characteristics of Cu-39.8Zn

The thermoelastic martensite that forms in β' brass has mechanical properties that are different from untransformed β' brass above M_s . At room temperature, which is above M_s and below M_d , plastic-flow and the formation of SIM are two mechanisms of deformation that describe the mechanical properties of β' specimens. However, below M_f the exact mechanisms of deformation of the martensite are not known and can only be speculated on at present. Nevertheless the mechanical properties of the martensite can be characterized.

At low temperature, below M_f , a deformed martensitic specimen either has total shape memory on heating, or residual permanent deformation. There are generally two forms of the permanent deformation. Grain boundary deformation appears as a narrow region of rotated surface that connects two grains, while plastic-flow, which first forms at triple points, resembles cross-slip in that irreversible wavy lines appear in parallel fashion. Grain boundary deformation is formed at very low stresses and in some cases small amounts are even seen in specimens that have 100% recovery. There is a possible connection between the massive α observed at grain boundaries (27) and

the grain boundary deformation since precipitates add to the incoherency of boundaries.

The martensite below M_f that undergoes an apparent-plastic deformation (apparent because the deformation is reversible on heating to A_f) is directly related to the shape memory. This type of deformation is suggested to be due to the growth of favorably oriented martensite plates at the expense of other martensite variants present (6, 11).

The mechanical properties of the martensitic RB specimens at 77°K show that no segment of the stress-strain curve obeys Hooke's Law, but there is an amount of nonlinear elasticity before yielding. The yield stress is not well defined in any of the martensitic specimens tested. In Figure 9 an RB specimen is loaded, unloaded, and loaded again, and shows hysteresis and a somewhat well-defined yield stress. Yield stresses can be estimated by approximating where the loading curve does not return to zero strain on unloading. It is noted that the nonlinear elasticity is not related to the strain observed at constant stress ("creep"), since the unloading curves have a constant average slope when the stress rate is varied (Figure 23).

The martensitic RB specimens have properties that are similar to those reported for single crystals of Cu-Zn-Sn (11) below M_f , in that no definite yield stresses are seen, and nonlinear elastic portions of the stress-strain curves are

found (Figure 28). When deformation begins in grains that do not continue across the diameter of the specimen, the result is a non-typical stress-strain curve (e.g., RB-4 shown in Figure 8). This indicates that the specimens continue apparent-plastic deformation until the deforming grain reaches its limit, which is marked by an increase in the work hardening slope. The amount of strain that accompanies the complete apparent-plastic deformation depends on the orientation and size of the grain undergoing deformation.

As the grain size decreases and the number of grains in the cross-section increases, the mechanical properties change in a typical manner (shown in Figure 29). The yield stress of the martensitic alloy increases since the strain in the first deforming grain must be compatible with the neighboring grains of different orientation and thus a larger stress is required for deformation. With smaller grains deforming, the slope of the work hardening region rises due to the average grain experiencing less strain at a given stress. The work hardening region also becomes more linear since the average stress of several grains is required for strain, and this does not fluctuate as much as that for single crystals. The polycrystalline specimens (RPC) show the ultimate effect of neighboring grains and have stress-strain curves comparable to polycrystalline specimens of Cu-Al-Ni below M_f (Figure 30), in that yield

stress is much higher than for single crystals and the work hardening region is steeper.

The geometry of the round and flat specimens contributes to a difference in stress gradients in the cross-sections; however, this does not affect the results to any unexpected degree since the flat specimens, with grain sizes between those of the RB and RPC specimens, have stress-strain curves that lie between the sets of curves for the two round specimen series.

The strains to fracture at low temperature are always limited by grain boundary separation, which initiates failure. The condition for complete failure to occur in a specimen is not simply the strain or stress state, but a combination of the two. Below M_f , the RB series, due to the freedom of the deforming grains, can strain to large values at low stresses; however, when grain boundary separation begins failure is catastrophic. In martensitic FLG and FSG specimens the stress concentration centers at triple points cause early grain boundary separation and lower strains to fracture. Additional strain occurs in polycrystals after the initial separation due to the crack being arrested at triple points by plastic flow blunting the crack tip. Due to the higher work hardening slopes and yield points, the RPC series at 77°K attains high ultimate tensile strength, but the strains to fracture are usually less than those for the other series with lower work hardening slopes. So below M_f , as

the grain size of the specimens decreases (and the number of triple points increases), the strain to fracture decreases.

The mechanical properties of the β' specimens at room temperature show higher yields than those of martensitic specimens tested below M_f , and the work hardening regions in the stress-strain curves are similar to those of the stable β' brass. The elastic moduli of all the specimens at room temperature are seen to be slightly variable with an average value of 6×10^6 psi (Figure 20). Rinehart (28) found the elastic moduli of β' brass single crystals to be highly anisotropic, with values ranging from 3.68×10^6 to 30.8×10^6 psi. The variation of slope from specimen to specimen can be explained by the difference in average orientation of the grains in the specimen.

The yield stress also depends on the orientation of the grains, in that the stress needed to form SIM is directionally dependent. This explains the difference in yield stresses of the RB specimens at room temperature (Figure 19). The RB specimens show three general regions in the stress-strain curves: elastic, plastic, and pseudoelastic. The pseudoelastic region is usually marked by serrated yielding, which is not seen in the stress-strain curves of the polycrystals.

The strains to fracture of the specimens tested at room temperature are all larger than those of the same series deformed at 77°K, and the fracture surfaces show more transgranular ductility. Even though β' brass has a higher yield stress than

the thermoelastic martensite, the grain boundaries do not separate until a large amount of SIM is formed at higher strains. (Some grains appear completely martensitic.) When grain boundaries separate the stress is so high that plastic flow blunts the crack tip immediately, which allows the specimen to continue straining without complete failure. In general, at room temperature, the smaller the grain size of a specimen, the larger the strain to fracture. (The RPC series has over 20% strain to fracture.)

B. Shape Memory Recovery and Pseudoelastic Strains

The shape memory effect is most apparent in the RB series. A large shape recovery is possible when the transformation of the martensite to the parent phase is not inhibited by permanent deformation. This is illustrated by the recoverability of the RB series which is not only dependent on the magnitude of the strain but also on the stress level attained. When a deforming grain completes apparent-plastic deformation at low temperature, the stress begins to increase and plastic-flow is initiated which inhibits full recovery. Thus some RB specimens have larger strains of 100% recovery due to low work hardening slopes allowing larger strains that remain below the plastic-flow stress. In Figure 31 it is seen that the percentage of strain recovered decreases with increasing stress. Absolute values of recovered strain range up to 10.85%. Larger values of strain are expected to be recovered due to shape memory from specimens that are single crystals since the apparent-plastic deformation will continue

the entire gage length. The amount of permanent deformation observed at triple points and grain boundaries after heating some specimens increases with less recovery and larger applied stresses.

As the grain size decreases, the amount of permanent deformation after heating increases for the same amount of shape memory. The recovery of the FLG specimens decreases from 99.6% at 18 ksi and 2.5% strain to about 90% at strain levels near fracture (6.0% strain). Some FLG specimens have larger yields and thus higher stresses for comparable strains, causing lower recoveries (e.g., 86.0% at 32.5 ksi stress and 4.65% strain). Similar observations are seen for the FSG specimen which has a higher yield and work hardening slope and a recovery of 94.1% at 31 ksi and 4.3% strain. Due to the higher yields, more triple points, and less strain to fracture, relative to the RB series the largest strain recovered in the flat series was 5.1%.

The influence of the yield stress and work hardening slope on the recoverability of the RPC series is seen to be similar to that for the flat specimens. In general, the RPC series shows that in a completely polycrystalline specimen, the more area under the stress-strain curve (i.e., the more energy put into deformation) the smaller the recovery on heating. The families of curves in Figures 14, 15, and 16 show decreasing recovery with larger energies of deformation. This accompanies an increase

in permanent deformation and areas of residual martensite. The largest strain recovered in a polycrystalline RPC specimen was 4.11% strain.

Recently Cu-39.4 and 39.6at%Zn single crystals have been shown to exhibit large amounts of pseudoelasticity above M_s (24). These studies used single crystals, which have no grain boundaries or triple points to cause permanent deformations which inhibit the reversion of SIM. The RB series, which have single grains in the test section but also grain boundaries, shows the elastic and pseudoelastic regions in the stress-strain curves as do the curves of the single crystals (see Figure 32). However, stress-strain curves of the RB series also show a plastic region (Figure 19) in the loading portion that equals the amount of residual strain after unloading. Even though the single crystals were tested closer to M_s than the RB specimens, it is evident that the presence of grain boundaries causes the plastic region in the stress-strain curves.

The beginning of the pseudoelastic region in the stress-strain curve is usually identified by serrated yielding, and also indicates the end of the plastic region, as seen in the curves of the second loading cycles, where serration occurs at low strains and the amount of plastic strain is small. If the serrated yield stress that marks the end of the plastic region in the first loading cycle is not exceeded, then the RB specimens exhibit high pseudoelasticity in the second loading cycle (Figure 24).

The residual martensite seen in the RB specimens after pseudoelastic recovery indicates that martensite forms before serration occurs. This shows that in the initial loading cycle plastic-flow and SIM form simultaneously until a stress is reached that allows SIM to continue to form pseudoelastically (with the ability to revert to the parent phase on unloading).

Since the polycrystalline specimens tested at room temperature have smaller grain size, the pseudoelasticity decreases to a state where the work hardening regions of the stress-strain curves show no indication of pseudoelastic yielding (no "dipping" or serration) and the unloading curves show no "reverse yield." At room temperature a specimen will show a linear unloading curve provided the applied stress is below that needed for the formation of SIM.

Even though the formation of SIM is abundant in all specimens tested at room temperature, the SIM remains to a larger extent in the smaller-grained specimens on unloading. The growth and disappearance of the martensite was observed through a tele-microscope at room temperature upon loading and unloading the specimen. When an RB specimen was strained past the elastic and plastic limits and into the pseudoelastic region, the martensite, which is formed as early as 1.5% strain, was seen to shrink on unloading. By applying a slight compressive stress, the martensite plates are seen to completely disappear in some cases. When the specimen is again loaded to the serrated yield stress the same

martensite plates appear to form again in the same locations as in the first tensile loading. Thus it is reasonable that compressive stresses at room temperature enhance the reversion of SIM.

Similarly, at low temperature, where the mechanism for the deformation of the martensite has been suggested as the growth of thermoelastic martensite with preferred orientations at the expense of other variants, a small compressive stress on a martensitic RB specimen caused the removal of some of the apparent-plastic deformation. (The remainder however is recovered upon heating.) This indicates that only a small energy barrier has to be exceeded for the stress assisted removal of the deformation and the recovery of strains below M_f . So the martensite that undergoes apparent-plastic deformation below M_f can either revert to the β' parent phase on heating to A_f causing shape memory or can have a stress of an opposite sign applied which will recover strain while remaining below M_f . Also, a specimen that undergoes apparent-plastic deformation below M_f can be maintained under stress while heating above A_f , which preserves the martensitic structure, and upon unloading the martensite reverts to the parent phase in a pseudoelastic manner.

VI. CONCLUSIONS

(1) The Cu-39.8wt%Zn martensite alloy deformed at 77°K shows a nonlinear stress-strain relationship on loading, and yields almost immediately (10 ksi) in specimens with grains that continue through the cross-section.

(2) As the grain size of the martensitic specimens becomes smaller the yield stresses and work hardening slopes increase while the strains to fracture decrease.

(3) Two different types of permanent deformation remain after incomplete shape recovery on heating. Grain boundary deformation, confined to local areas between grains, begins soon after yielding, and small amounts are observed even in specimens that show 100% recovery. Plastic deformation, which resembles cross-slip, begins at higher stresses than those for grain boundary deformation and usually is initiated at triple points.

(4) At low temperature grain boundary separation causes fracture at low strains and thus limits the extent of shape recovery on heating.

(5) Specimens with grains that continue through the cross-section (RB series) showed strains to fracture over 15% and a total shape recovery up to 10.85% strain. In contrast,

polycrystalline specimens (RPC series) exhibit strains to fracture of about 6% and show a maximum recovery of 4.8% strain on heating. (The magnitude of recoverable strain in the RB specimens is limited by the grain shape and orientation. Since the grains in the specimens tested did not extend the length of the test section, larger amounts of recoverable strain due to shape memory are expected from true single-crystals.)

(6) Because of high yield stresses and work hardening slopes, the polycrystalline specimens deformed below M_f have large energies of deformation (area under the stress-strain curve). The permanent deformation is more abundant in specimens showing less recovery and inhibits complete reversion of the martensite. Thus below M_f the simultaneous formation of permanent and apparent-plastic deformation causes the percentage of shape recovery to be reduced with increasing energies of deformation.

(7) The apparent-plastic deformation of martensite below M_f can be restored by transforming back to the original β' parent phase on heating to A_f (shape memory) or can be recovered below M_f by applying a small stress of opposite sign. Also the deformed martensite, with a superimposed stress maintained while heating, persists above A_f ; then upon unloading the martensite reverts to the β' phase in a pseudoelastic manner.

(8) After a stress has been applied to a polycrystalline specimen, below M_f , and held constant, a certain amount of

additional strain, "creep," is observed. After specimens are stressed for 2 minutes, amounts of creep range from 0.250% to 0.500% strain.

(9) At room temperature (above M_s and below M_d) the yield stress is higher in the polycrystals relative to the single crystals, and accordingly the former do not show the pseudo-elastic effect because plastic-flow inhibits the reversion of SIM.

(10) In the RB series the SIM shrinks and disappears to some extent on unloading (pseudoelasticity). Additional SIM can be made to disappear by the application of a stress of opposite sign.

Table 1. Percentage of recovery, apparent plastic strain, maximum applied stress and magnitude of shape recovery for Cu-39.8Zn specimens undergoing shape memory.

SPECIMEN NO.	PERCENTAGE OF RECOVERY (%)	APPARENT STRAIN (%)	MAXIMUM STRESS (ksi)	MAGNITUDE OF SHAPE RECOVERY (%)
RB-01	94.0	10.85	38.0	10.2
-02	100	7.20	23.0	7.2
-03	100	5.05	22.0	5.05
-04	92.0	8.05	52.0	7.4
FLG-01	99.6	2.50	18.0	2.49
-02	97.2	3.35	19.0	3.25
-03	96.2	5.25	24.0	5.05
-10	92.1	4.75	24.0	4.38
-11	90.0	5.25	30.0	4.72
-12	86.0	4.65	32.5	4.00
FSG-04	94.1	4.25	31.0	4.00
RPC-01	96.5	2.00	37.5	1.93
-02	90.1	2.75	45.0	2.48
-03	88.0	3.55	47.5	3.12
-21	100	2.65	24.0	2.65
-22	96.8	4.25	35.0	4.11
-23	94.4	2.65	38.5	2.50

Figure 1. Temperature vs. weight percent zinc in copper showing stress free M_s temperature. Data are from Titchener and Bever (2).

(a)

(b)

Figure 2. Stress-strain curves showing (a) ferroelasticity in β' -Au-30Cu-44Zn single crystals and (b) superelasticity in β -Au-47.5Cd single crystals (22) (compositions in at%).

Figure 3. Stress-strain curve for Cu-Zn-Sn alloy deformed above M_s showing pseudoelastic strain and serrated yielding (3).

Figure 4. Effect of temperature on tensile behavior of Cu-34.1Zn-1.8Si alloy (at%). (a) Stress-strain curves at different temperatures. (b) Variation of stress required to produce martensite with test temperature (3).

Figure 5. Schematic illustration showing ranges of stress and temperature for pseudoelasticity (shaded region) (3).

Figure 6. Effect of deformation temperature on the pseudo-elastic strain and shape memory in single crystals of Cu-34.7Zn-3.0Sn (11).

Figure 7. Design of (a) round tensile specimens with various gage lengths from 0.400" to 0.750" and (b) flat tensile specimens. All dimensions are in inches.

Figure 8. Stress-strain curves of RB specimens tested below M_f . Shape recoveries are indicated.

Figure 9. Stress-strain curve of an RB specimen tested below M_F , which is cycled to zero stress, showing a hysteresis loop and a more well defined yield stress. Shape recovery is indicated.

Figure 10. Stress-strain curve of stable β' brass specimen tested at 77°K in load control.

Figure 11. Stress-strain curves of FLG specimens tested below M_f . Shape recoveries are indicated.

Figure 12. Stress-strain curves of FLG specimens tested below M_f . Shape recoveries are indicated.

Figure 13. Stress-strain curves of FSG specimens tested below M_f . Shape recoveries are indicated.

Figure 14. Stress-strain curves of RPC specimens tested below M_f with extensometer knife edges placed within the gauge length of the specimen. Shape recoveries are indicated.

Figure 15. Stress-strain curves of RPC specimens tested below M_f with extensometer knife edges placed outside the gage length of the specimen. Note the reduction of shape recovery with larger energies of deformation.

Figure 16. Stress-strain curves of RPC specimens tested below M_f . Specimens 1, 2 and 3 have a 0.400" gage length, while the others (4 and 5) have a 0.750" gage length. Shape recoveries are indicated.

Figure 17. Stress-strain curves of RPC specimens tested below M_f showing additional strain, called "creep," that results from maintaining a stress for two minutes. Shape recoveries are indicated.

Figure 18. Stress-strain curve of an RPC specimen tested below M_f and then maintained under stress while heating above A_f . On loading at 296°K the martensite reverted in a pseudo-elastic manner.

Figure 19. Stress-strain curves of RB specimens tested above M_s showing three general regions: elastic, plastic, and pseudoelastic. (These are set apart by the points A and B.)

Figure 20. Stress-strain curves of FLG specimens tested above M_s showing "dip" after yield and limited pseudoelastic strains. Specimen FLG-32 fractured at 12% strain.

Figure 21. Stress-strain curves of FSG specimens tested above M_s showing no "reverse yield." The average slope of the unloading curve is similar to the elastic modulus slope.

Figure 22. Stress-strain curve of an RPC specimen tested above M_s having fracture at 21% strain.

Figure 23. Stress-strain curve of an FLG specimen tested below M_f showing that changing the loading rate from 10 ksi/min (between points A and B) had little effect on the size of the hysteresis loop or the average slope of the unloading curve.

Figure 24. Stress-strain curve of an RB specimen tested above M_{ss} showing reduction of plastic strain in the second loading cycle, as compared with that in the first.

Figure 25. Stress-strain curve of an FLG specimen tested above M_s showing increased percentage of pseudoelastic strain in second loading cycle.

Figure 26. Stress-strain curve of an RB specimen tested below M_f showing the effect of compression after initial tensile deformation.

Figure 27. Stress-strain curve of an RB specimen tested above M_s showing recovery of strain after applying a slight compressive stress.

Figure 28. Stress-strain curve for Cu-34.7Zn-3.0Sn single crystal tested below M_f showing a nonlinear elastic portion and no well defined yield stress (11).

Figure 29. Characteristic stress-strain curves for Cu-39.8Zn specimens of various grain sizes.

Figure 30. Stress-strain curves of (a) Cu-14Al-3Ni (25) and (b) an RPC specimen, both tested below M_f , showing a similarity in curvature.

Figure 31. Percentage of shape recovery vs. maximum applied stress showing an inverse dependence for RB specimens.

Figure 32. Stress-strain curve for a Cu-39.6at%Zn single crystal tested above M_s showing a large amount of pseudoelastic strain (24).

LIST OF REFERENCES

1. A. B. Greninger and V. G. Mooradian, Trans. AIME, 1938, 128, 337.
2. A. L. Titchener and M. B. Bever, Trans. AIME, 1964, 200, 303.
3. H. Pops, Met. Trans., 1970, 1, 251.
4. J. E. Reynolds and M. B. Bever, Trans. AIME, 1952, 194, 1065.
5. K. Otsuka, K. Shimizu, I. Cornelis, and C. M. Wayman, Scripta Met., 1972, 6, 377.
6. R. E. Hummel, J. W. Koger, and V. Pasupathi, Trans. AIME, 1968, 242, 249.
7. E. J. Suoninen, R. M. Genevray, and M. B. Bever, Trans. AIME, J. of Metals, 1956 (Feb.), 283.
8. H. Pops and T. B. Massalski, Trans. AIME, 1964, 230, 1662.
9. C. M. Wayman, Scripta Met., 1971, 5, 489.
10. W. J. Buehler, J. V. Gilfrich, and R. C. Wiley, J. Appl. Phys., 1963, 34, 1475.
11. J. D. Eisenwasser and L. C. Brown, Met. Trans., 1972, 3, 1359.
12. C. M. Wayman and K. Shimizu, Metal Science Journal, 1972, 3, 1359.
13. L. C. Chang and T. A. Read, Trans. AIME, 1951, 191, 47.
14. H. Tas, L. Delaey, and A. Deruyttere, to be published.
15. K. Otsuka and K. Shimizu, Scripta Met., 1970, 4, 469.

16. D. V. Weld and E. Gillam, Scripta Met., 1972, 6, 1157.
17. Z. S. Basinski and J. W. Christian, Acta Met., 1954, 2, 161.
18. W. A. Rachinger, J. Aust. Inst. Metals, 1960, 5, 114.
19. N. Nakanishi, T. Mori, S. Miura, Y. Murakami, and S. Kachi, Phil. Mag., 1973, 28, 277.
20. S. Miura, T. Mori, and N. Nakanishi, Scripta Met., 1973, 7, 697.
21. H. K. Birnbaum and T. A. Read, Trans. AIME, 1960, 218, 662.
22. N. Nakanishi, Y. Murakami, S. Kachi, T. Mori, and S. Miura, Phys. Let., 1971, 37A, 61.
23. H. Pops, Trans. AIME, 1966, 236, 1532.
24. W. Arneodo and M. Ahlers, Scripta Met., 1973, 7, 1287.
25. K. Oishi and L. C. Brown, Met. Trans., 1971, 2, 1971.
26. U. F. Kocks, Acta Met., 1958, 6, 85.
27. I. Cornelis, private communication.
28. J. S. Rinehart, Phys. Rev., 1941, 59, 308.