EFFECT OF TEMPERATURE ON THE DEFORMATION BEHAVIOR OF TWO-PHASE BICRYSTALS OF ALPHA-BETA BRASS

Dissertation for the Degree of Ph. D.
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This is to certify that the

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EFFECT OF TEMPERATURE ON THE DEFORMATION BEHAVIOR OF TWO-PHASE BICRYSTALS OF ALPHA-BETA BRASS

presented by

ALI ASGHAR KHEZRI YAZDAN

has been accepted towards fulfillment
of the requirements for

Ph.D. degree in Metallurgy

K. N. Subramanian
Major professor

Date January 20, 1976

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ABSTRACT

EFFECT OF TEMPERATURE ON THE DEFORMATION BEHAVIOR OF TWO-PHASE BICRYSTALS OF ALPHA-BETA BRASS

By

Ali Asghar Khezri Yazdan

Two-phase bicrystals of alpha-beta brass were deformed at various temperatures to investigate the effect of ordering and disordering in beta brass on the overall deformation behavior of this two-phase alloy. Tensile testing of these specimens was carried out at various strain-rates at 700°, 800°, 900° and 1000°F to gain understanding of the basic mechanisms involved. Single crystals of alpha and beta brass, as well as coarse-grained beta brass, were also deformed at these temperatures at various strain-rates. The deformation structures were studied by optical microscopy.

Deformation of Beta, unlike Alpha, was found to be highly sensitive to temperature and strain-rate. The predominant mode of deformation in Beta at temperatures below T_c (ordering temperature) was by grain boundary sliding and at temperatures above T_c it was by slip. Further, Beta did not exhibit any work-hardening at temperatures above T_c.

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Unlike in the room temperature deformation of these bicrystals, interaction of slip in one phase with the phase boundary was not an important factor in initiating deformation in the other phase. Overall deformation of the bicrystals at these temperatures depended on the deformation of the individual phases. Usually, the deformation of one phase did not influence the deformation of other phase. The alpha-beta phase boundary did not play any role in the deformation of these bicrystals.

It was discovered that a uniform deformation of both the phases present in alpha-beta brass could be achieved by using low and moderate strain-rates at temperatures below T_c and by using high strain-rates at temperatures above T_c. The results are explained by using simple hypothetical models.

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EFFECT OF TEMPERATURE ON THE DEFORMATION BEHAVIOR OF TWO-PHASE BICRYSTALS OF ALPHA-BETA BRASS

By

Ali Asghar Khezri Yazdan

A DISSERTATION

Submitted to
Michigan State University
in partial fulfillment of the requirements
for the degree of

DOCTOR OF PHILOSOPHY

Department of Metallurgy, Mechanics and
Materials Science

1976

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ACKNOWLEDGMENT

I would like to express my deepest appreciations and gratitudes to my major advisor Dr. K.N. Subramanian for his contributions through his guidance and encouragements. My appreciations are also extended to members of my thesis committee, Drs. G. Cloud, D. McGrady and W. Hartman (sitting in for Dr. P. Schroeder) for their valuable suggestions along with Mr. and Mrs. Siriani for helping me in editing.

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My special love and affections go to my wife Naheed and my daughter Shabnam who suffered the most along with our parents during this work.

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1.1 General

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CHAPTER I

INTRODUCTION

1.1 General

A vast number of engineering materials used in practical applications are made of two-phase or multi-phase substances. These materials, depending on their properties, are used in making various types of products. One has to investigate and understand these materials better, so that they can be used most efficiently under severe thermo-mechanical conditions of service. Examples of such two-phase materials are, steel, duplex stainless steel and two-phase alloys of titanium-Aluminum, copper-zinc (brasses), copper-tin (bronzes) and copper-aluminum. Although these two-phase materials are widely used, a thorough understanding of their mechanical behavior is far from complete.[1]

In order to understand the basic mechanisms controlling two-phase materials, the number of factors influencing their deformations should be minimized. Single crystals constitute ideal models for studying the deformation behavior of single phase materials. One cannot produce a single crystal of a two-phase material, since such a crystal can have only one phase. So, a basic unit for a two-phase mate

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two-phase material must be a two-phase bicrystal. Such a two-phase bicrystal will have a phase boundary in the fundamental unit. The deformation characteristics of the phase boundary may correspond to that of a grain boundary; as a result, two-phase bicrystals will be, in reality, close to a bicrystal of a single phase material, which has a boundary in the model system.

The fundamental model for a two-phase material should consist of a single crystal of one phase joined to a single crystal of another phase. However, it is not possible to obtain such fundamental units for most of the two-phase materials. Quite often, both the phases present in a two-phase material are metallic in nature. One such material is alpha-beta brass. In the case of alpha-beta brass, the fundamental unit used for deformation studies consists of a single crystal of alpha joined to a coarse-grained polycrystal of beta. Often, only a single grain of beta will be in contact with the alpha single crystal. Such a unit can be called a two-phase bicrystal.[2]

Hingwe and Subramanian[3] have developed a technique for growing bicrystals of alpha-beta brass. Hingwe[4] and Nilsen[5] have studied the mechanical properties of such bicrystals of alpha-beta brass at room temperature, under uniaxial tension, at various strain rates. These studies on bicrystals of alpha-beta brass were aimed at understanding the basic mechanisms involved in the deformation two-phase

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of two-phase materials. During elastic deformation, response of such a bicrystal to an applied stress depends on the ability of its constituents to transfer strain across the phase boundary. The two participant phases in a bicrystal will differ in their responses to an applied stress. This is a direct result of differences in elastic constants, crystal structures, relative crystallographic orientations, etc. The number of available slip systems, Peierls-Nabarro stresses, work-hardening in each phase, the ability for the deformation to progress through the boundary, and the relative crystallographic orientations of the phases will affect the over-all plastic deformation behavior of such bicrystals.

All of the mechanical tests performed on bicrystals of alpha-beta brass have been carried out at room temperature. Materials such as brasses are often hot-rolled and extruded during manufacturing processes. In order to understand the basic mechanisms controlling the deformation behavior of such materials at elevated temperatures, mechanical tests have to be carried out on fundamental units, such as two-phase bicrystals, at various temperatures. Beta brass undergoes a structural change at 850°F (454°C). The effect of such a structure change on the high temperature deformation behavior can also be investigated by using alpha-beta brass bicrystals.

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The purpose of this study is to grow bicrystals of alpha-beta brass and to study the effect of temperature on the mechanical behavior of such specimens at elevated temperatures.

The effect of the phase boundary orientations with respect to the tensile axis are studied at room temperature and are presented in Appendix A.

Growth and mechanical properties of Alpha, Beta as well as of two-phase bicrystals of alpha-beta brasses are reviewed in the following chapter.

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CHAPTER II

HISTORICAL REVIEW

2.1 Growth of Crystals

2.1.1 Growth of Alpha and Beta Brass Single Crystals

Single crystals of alpha and beta brass can be grown by the Bridgmen technique. In this technique, the molten brass, which is contained within a sharp-tipped graphite crucible, is lowered through a temperature gradient. Upon cooling, the nucleus is formed at the sharp tip, and the rest of the crystal grows on it. Maintaining a uniform chemical composition during growth of an alloy single crystal is extremely difficult, since segregation contributes to the non-uniformity of composition. In brasses, the parts solidifying first contain less zinc than those solidifying later. Non-uniformity in composition can be minimized by employing very slow crystal growth-rates. Furthermore, annealing the brass in a sealed tube at 1450°F (800°C) for 16 hours will completely remove any segregation.[6]

2.1.2 Growth of Bicrystals of Alpha-Beta Brass

Hingwe and Subramanian[3] developed the technique for growing bicrystals of alpha-beta brass in which beta brass (48.5 w/o Zn) was melted on a single crystal substrate

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of alpha brass (30 w/o Zn). A two-phase foil with an intermediate zinc composition of 40 w/o Zn was placed between alpha and beta stock before melting, to reduce the two-phase transition zone formed between the alpha and beta phases. Two special heat-treatments were employed to reduce the two-phase transition zone formed. In the first heat-treatment, the as-joined crystal was lowered through a temperature gradient. This operation helped to reduce the size of the transition zone, and caused the alpha and beta platelettes present in the transition zone to become continuous with alpha and beta regions respectively. In the second heat-treatment, the transition zone was heated to 1450°F (800°C) in a furnace having a very sharp temperature gradient for about one hour. The sample was then cooled to room temperature. This process was repeated several times. During the heating cycle, the two-phase transition zone becomes a single phase beta region, and diffusion of zinc from regions containing higher zinc concentration to a lower concentration is made possible. Beta grain growth also takes place during such a cyclic local annealing. Repetitive heating and cooling produces the required bicrystal. Since then, such a unit has been referred to as the basic model for fundamental studies on alpha-beta brass by various researchers.[3,4,5] The deformation behavior of such model systems will invariably be dependent on the deformation behavior of the individual phases, and also on the deformation of phase boundary.

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The deformation behavior of alpha and beta brasses are reviewed in the following section.

2.2 Deformation of Single Phase Materials

2.2.1 Deformation of Single Crystals

Plastic deformation by slip occurs in crystalline materials when the shear stress acting along a specific direction (slip direction) on a specific plane (slip plane) reaches a critical value. The value of stress at which this plastic flow occurs is defined as the critical resolved shear stress. The Schmidt Law[7] relationship for the critical resolved shear stress σ_c is:

$$\sigma_c = \sigma_o \cos \lambda \cos \phi$$

where σ_o is the yield stress in uniaxial loading, λ is the angle between the normal to the slip plane and the axis of loading, and ϕ is the angle between the slip direction and the axis of loading. In this expression $(\cos \lambda)$ $(\cos \phi)$ is termed the Schmidt factor.

Single crystals of face-centered cubic materials, [F·C·C] such as copper and alpha brass, exhibit three stages of work-hardening during deformation. During the first stage, there is a low work-hardening rate. The extent of this stage is sensitive to crystal orientation, purity and temperature of testing. In Stage II there is much higher rate of work-hardening, which is constant at moderate temperatures. This stage is independent of

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crystallographic orientation and other test variables. Single as well as polycrystals exhibit this stage. The decreasing work hardening rate in Stage III is attributed to dynamic recovery of the material due to cross-slip and conservative climb. Body centered cubic, [B·C·C], metals such disordered beta brass, display a decreasing strain-hardening rate over their entire range of deformation.

Orowan[8] has made observations on the factors influencing the strain-rate sensitivity of materials and has given the formula for strain-rate $(\dot{\varepsilon})$ as,

$$\frac{d\varepsilon}{dt} = \dot{\varepsilon} = \rho b v$$

where ρ is the mobile dislocation line length per unit volume, b is the Burgers vector of the mobile dislocations, and v is the average velocity of the dislocations. The magnitude of each of the factors b, v, and ρ can vary with respect to time.

Johnston and Gilman[9] suggested that the total strain was directly related to the density of the mobile dislocations. They also suggest that dislocations can multiply, especially by a cross-glide process. It has been observed that materials such as beta brass are highly strain-rate sensitive. With increasing strain-rate, the ease of movement of mobile dislocations decreases, because of the more frequent dislocation interactions. After the mobile dislocations are tangled in some ways, a higher stress is required

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At elevated temperatures (if the strain-rate of the tensile test is small), the specimen can deform by creep. Creep is a thermally activated process and the thermal fluctuations aid the applied stress in overcoming the resistance for plastic flow. Since the deformation of crystalline materials occurs primarily by the motion of dislocations, the deformation behavior of crystals is determined by the method by which these dislocations overcome obstacles.

The deformation behavior of crystalline metals at high temperature is usually dependent on a diffusion controlled mechanism.

Examination of polished surfaces of metallic crystals deformed at high temperatures indicates that slip occurs in the two forms described below:

(a) Coarse slip bands whose spacing and displacements are of the order of 1 to 10μ ($1\mu = 10^{-3}$mm). These slip bands in nature are steps in the crystal surface and can be seen at relatively low magnifications. The spacing between coarse slip bands depends on the stress experienced by the specimen.

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(b) Fine slip lines are those whose spacing will be of the order of several angstroms. These fine markings appear in areas between prominent slip bands.

McLean[10] observed polygonization in Aluminum single crystals during the early stages of deformation at elevated temperatures. The polygonization process requires climb of arrested dislocations in a direction normal to the slip plane. These dislocations subsequently glide and form a vertical wall which is termed a low angle grain boundary.

McLean[11] also reported that above certain temperature recovery becomes possible and offsets some of the strain-hardening.

2.2.1.a <u>Deformation Behavior of Single Crystals of Alpha and Beta Brasses</u>

The structure of alpha brass is face-centered cubic, and atoms of zinc occupy randomly some of the lattice positions of copper atoms. Alpha brass normally deforms in close packed {111} planes and along <110> directions. There are a total of twelve operable slip systems in alpha brass. Those dislocations with a Burgers vector that retain the original crystal structure in the deformed lattice, are referred to as perfect dislocations. In that case atomic translations in some specific direction do not restore the crystal to its original structure, but do leave it in a stable state, thus forming a stacking fault.

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Heiderich and Shockley[12] have suggested that a dislocation in F.C.C. crystal structure can dissociate into two partial dislocations, according to the reaction

$$\frac{a}{2}\,[110] = \frac{a}{6}\,[211] + \frac{a}{6}\,[12\bar{1}];$$

and thereby lower the energy of the system.

The F.C.C. structure of alpha brass is retained up to its melting point. The process of dislocations cutting through the forest dislocations present becomes more possible at elevated temperatures. Since blocked dislocations become liberated from their locked positions, the deformation of alpha brass becomes easy at elevated temperatures.

The ordered beta below critical temperature (T_c), has cesium chloride (Cs Cl) structure. This cesium chloride structure consists of two interpenetrating simple cubic unit cells, one of copper and the other of zinc, and is referred to as B-2 type superlattice. Figures 1 (a) and (b) illustrate such a structure in two and three-dimensional atomic arrangements. Slip systems in this structure are of the type $\{110\}\ \langle\bar{1}\bar{1}1\rangle$.

In B-2 type structures,[13] when a dislocation, whose Burgers vector is a unit lattice vector moves through an ordered lattice on a slip plane, a disturbance in the local arrangement of the atoms on this slip plane is produced. This process creates an antiphase boundary. A second dislocation of the same Burgers vector is required

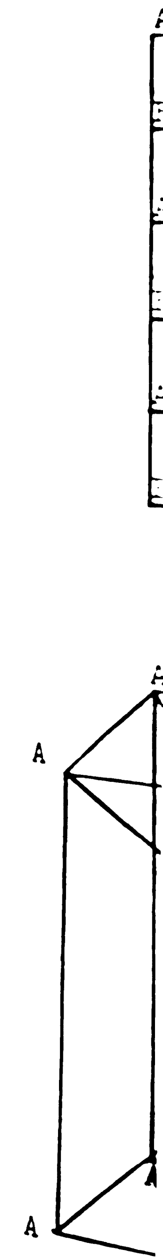
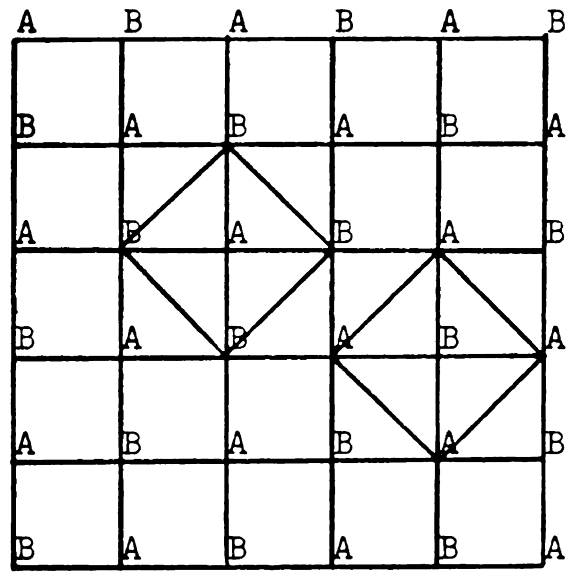


Figure 1. Cesium Chloride Structure

(a) A two-dimensional model of Cesium Chloride structure. "A" representing Cs and "B" representing Cl or vice versa.

(b) A three-dimensional model for CsCl type structure and a representation for beta brass in superlattice structure (a unit cell). (J.P. Hirth 14).



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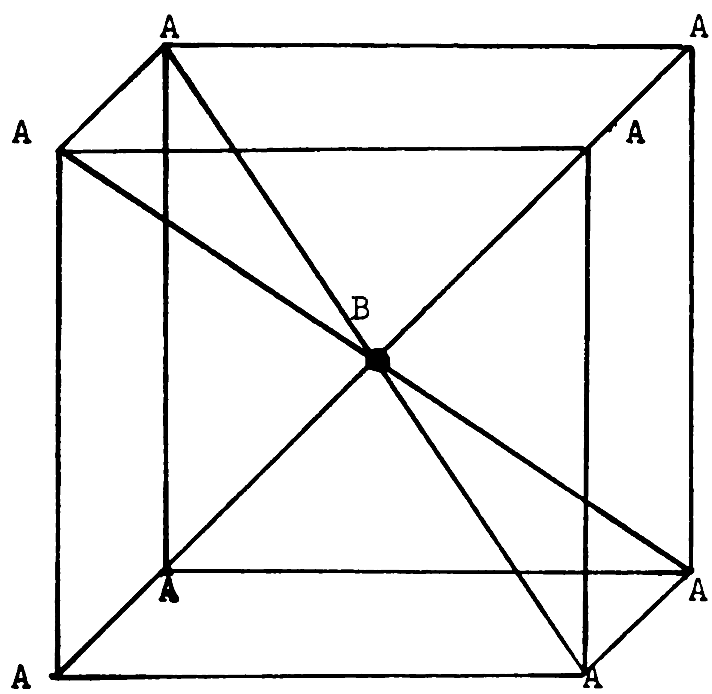

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Beta brass exists in a disordered state at temperatures above T_c [850°F or 454°C] and has a B.C.C. structure on the average. The lattice positions can be occupied by either copper or zinc. In the B.C.C. structure, the slip system is of the type $\{110\}$ $\langle\bar{1}\bar{1}1\rangle$. The dissociation of perfect dislocations in B.C.C. lattice is not common, and therefore, screw components are not limited to a particular slip plane. Since three $\{110\}$ type planes intersect in a $\langle\bar{1}\bar{1}1\rangle$ direction, screw dislocations may move in a haphazard way on those $\{110\}$ planes depending on the local stress state.[15] For this reason, the slip lines in beta brass are often wavy and ill-defined.

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J.H. Westbrook[16] studied the relation between the critical resolved shear stress and the test temperature of beta brass. His results are in Figure 2. The sudden drop of the critical resolved shear stress for beta brass at temperatures above T_c is accounted for on the basis of decrease in short-range order.

2.2.2. Deformation of Bicrystals

A bicrystal of a single phase material consists of two grains of the same phase. The grain boundary is the region of disregistry between these grains. Grain boundaries resist the motion of dislocations during plastic deformation of a bicrystal. Due to the irregularities in the structure of a grain boundary, a grain boundary is often referred to as a forest of dislocations. During the deformation of such a bicystal, it appears that some kind of a slip interaction with the boundary should take place for the propagation of slip from one grain to the other. Grain boundaries are effective barriers in blocking dislocations. Grain boundaries are found[17] to undergo structural changes sooner than the grains (by phase transformations), and in most cases exhibit a loss of strength at elevated temperatures.

2.2.2.a. Deformation Behavior of Bicrystals of Alpha and Beta Brasses

During room temperature deformation, a grain boundary in a bicrystal of Alpha acts as a barrier to the

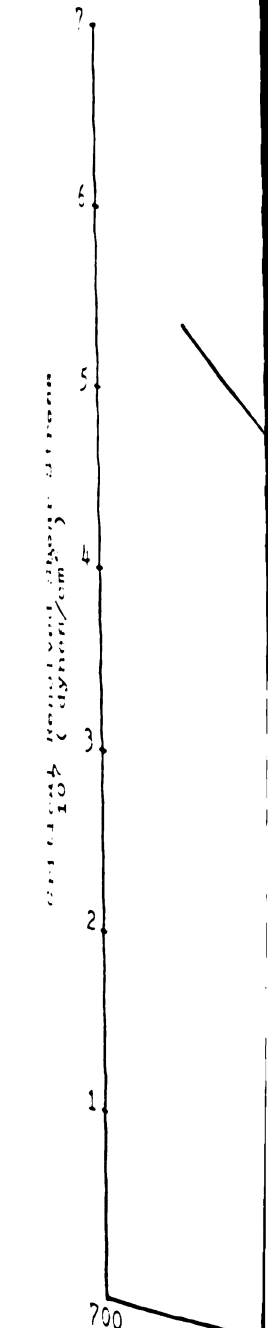
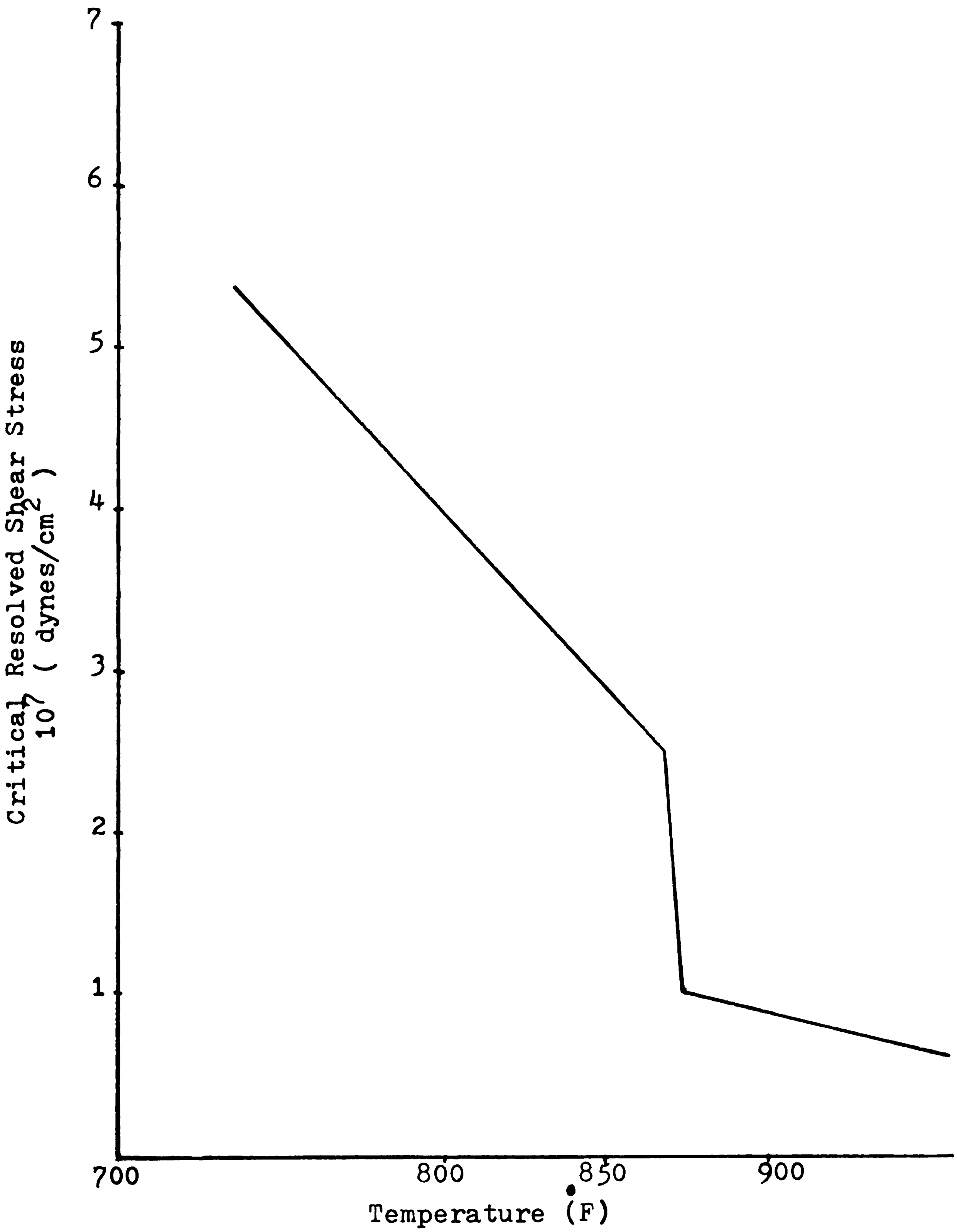


Figure 2. Critical Resolved Shear Stress for Beta Brass Single Crystals.

After, J.H. Westbrook,[16] Mechanical properties of intermetallic compounds. A review of the literature in "Mechanical Properties of Intermetallic Compounds", J. Wiley, N.Y. (1960), p. 180.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interaction dislocations. Usually a favorably oriented grain deforms first. In such a case, deformation progresses through the boundary after the deforming grain work-hardened considerably. The resolved shear stress for slip to propagate through the boundary is greatly enhanced by the structure of the boundary, which strongly depends on the temperature.

McLean[11] studied some of the mechanical behaviors of grain boundaries in Cu-Zn alloys at elevated temperatures. The grain boundaries in alpha bicrystals have been observed to slide with respect to each other. A grain boundary with an orientation of 45° with respect to the tensile axis was observed to undergo a continuous deformation and necking took place at this grain boundary.

Chuang and Margolin[18] studied the stress-strain relations of beta brass bicrystals at room temperature. The grain boundaries in these specimens were parallel to the compression axis. The isoaxial beta brass bicrystals used were elastically and plastically incompatible. This incompatibility led to the formation of a clearly defined grain boundary deformation zone for strains up to 3.2 percent. Outside of the grain boundary deformation zone, each component crystal behaved as a single crystal. For the non-isoaxial bicrystal of beta brass consisting of a hard and a soft component, it was possible to express the total applied stress in terms of the sum of the product of the average stresses and volume fraction of the hard and soft crystals respectively.

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Chuang[19] observed that grain boundaries in beta brass undergo grain boundary sliding at temperatures above 50°C (122°F). Grain boundaries in beta brass are highly sensitive to the rate of deformation. The beta brass bicrystals, which had a large grain boundary surface area, deformed to the shape of a spear at elevated temperatures.

Grain boundaries in beta brass at elevated temperatures above T_c, resisted deformation at high strain-rates. Consequently the grains themselves accommodate almost all of the strain.

2.2.3. Deformation Behavior of Polycrystals

General

The deformation behavior of a polycrystalline material is more complicated than that of a single crystal. Grain boundaries impose added constraints to the deformation of individual grains. Unlike the grains, the grain boundaries do not have any repeating patterns in their structures, and their deformation behavior is not clearly understood.

The effect of temperature on the deformation behavior of polycrystals is very significant. In most cases the result is a drop in the yield stress arising from the loss of strength in the grains and in the grain boundaries. In certain cases where the temperature and testing conditions are met, a region in a form of a narrow strip adjacent to the grain boundaries is deformed. This kind of deformation occu

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of deformation occurs at low strain rates and at high temperatures in coarse-grained polycrystals and is called microcreep. At elevated temperatures, a mechanism by which vacancies migrate from one side of a grain to the other side takes place. At low strain rates, because of the motion of vacancies, further plastic deformation proceeds. This is a stress motivated diffusion process.

2.2.3.a. Deformation Behavior of Polycrystalline Alpha and Beta Brass

Karashima[20] studied the deformation behavior of alpha brass polycrystals. He concluded that the process of deformation propagation in polycrystalline aggregates is due to stress concentrations from piling up of dislocations against the grain boundaries. There were two possibilities that could account for this process. The first one postulates that Frank-Read sources in neighboring grains begin to operate as a result of stress concentrations;[21] the second one suggests that dislocations are emitted from grain boundaries.[22]

Observations made on deformed samples of alpha brass polycrystals show that[23] dislocations piled up against grain boundaries at small strains and slip lines are continuous across the boundaries.

Greninger[24] made observations on polycrystals of beta brass deformed slightly in a vise at room temperature. Two distinct structural characteristics were observed on the polished surfaces:

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(1) Sets of parallel fine hairlike lines, referred to as slip bands, and

(2) a corrugated relief, which under oblique illumination appeared as a coarse light and dark banding effect to the surface of a grain. These were referred to as "deformation bands".

Neither slip lines nor deformation bands continued over a grain boundary without changing direction; hence, both must be related in some way to the orientation of the grains and the superlattice structure of Beta at room temperature. Slip lines in beta brass are decidedly less prominent than those in alpha brass.

Grain boundaries in beta brass polycrystals add to the strength of the aggregate structure. Beta brass yields at a fairly high stress. The difficulty involved in the deformation of beta brass is because of the motion of the superlattice dislocations. Deformation of beta brass at a high strain rate results in transcrystalline fracture.

The high temperature tensile deformation of 70-30 (Cu-Zn) polycrystals of alpha brass was studied by Asano and co-workers.[25] The yield stress of the alpha brass polycrystals was determined through a range of temperatures. There was found a marked drop of yield stress as the temperature was increased. Microcreep mechanisms are accounted for

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The deformation of beta brass is more or less accommodated in the grain boundaries by grain boundary sliding below T_c. As the temperature increases, the grains react to the supplied energy and ultimately undergo deformation. Those grain boundaries, with an orientation close to 45° with the tensile axis, undergo a continuous deformation. Depending on the strain rate,[26] the final surface of the deformed specimen becomes shiny and in the case of coarse-grained beta brass, grain boundary deformation can be observed with the naked eye.

The deformation of coarse-grained beta brass at temperatures below T_c was studied by Chunke.[27] His work also showed that the grain boundary orientation with the tensile axis was extremely important in determining the deformation behavior. Specimens having grain boundaries that were oriented at about 45° to the tensile axis slid from the start, and continued deforming uniformly; beta grains were intact and did not deform plastically through the entire deformation. Grain boundaries, with orientations perpendicular or parallel to the tensile axis, underwent very little grain boundary sliding and fractured right after the yield point.

Based on the observations made by Chung[28] on the fractured surfaces, it was suggested that the failure was ecause of a high

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Uniform deformation in beta brass polycrystals was observed at temperatures above T_c at intermediate strain-rates. Deformation by grain boundary sliding was more strain-rate sensitive than deformation of individual grains. Fine-grained specimens exhibited a higher strain-rate sensitivity at temperatures just above T_c compared to coarse-grained beta brass.

2.3. Deformation of Two-Phase Materials

General

So far, this review has dealt with deformation of single phase materials. A more complicated deformation behavior is observed in two-phase materials. Two-phase materials contain phase boundaries rather than grain boundaries. The deformation behavior of such two-phase materials that have phase boundaries is not well understood.

Marcinskowski and co-workers[29,30,31] have developed some theories with respect to the deformation behavior of internal boundaries. Grain boundary dislocation is discussed in terms of disturbances created at a boundary, when a glide dislocation crosses a grain boundary. The nature of the grain boundary dislocation can be determined from a knowledge of the orientation relationships between slip systems adjacent to the boundary. In fact,

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the Burgers vectors of all of the dislocations associated with all of the internal boundaries, can be represented by a single expression:

$$\vec{b}_I = (\vec{b}_1 - a\vec{b}_2)$$

where $\vec{b}_I$ is the effective Burgers vector in the grain boundary, $\vec{b}_1$ and $\vec{b}_2$ are the Burgers vectors of the adjacent phases, and a is an entry in the matrix of directional cosines between the $\vec{b}_1$ and $\vec{b}_2$. They also showed that the orientation and shape of the boundaries are altered as a result of a homogeneous shear. Such a shear also leaves an array of dislocations that may have long range stress fields, which could be relieved by nucleation of crystal lattice dislocations in both of the phases. The nature of the interface resultant dislocations was derived from some simple orientations as a result of passage of a dislocation through the boundary. Further, the concepts related to grain-boundary dislocations were claimed to be applicable to the deformation of internal boundaries, such as grain boundaries, twin boundaries, two-phase interfaces, etc.

2.3.1 Deformation of Bicrystals of Two-Phase Materials

The deformation behavior of duplex and bicrystals of alpha-beta brass was studied[3,4,5] at room temperature at different strain-rates. There were four types of boundaries considered, namely, oriented duplex, flat

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boundary, corrugated boundary, and equiaxed duplex. In the progression of deformation from one phase to the other, the equiaxed duplex was the most effective type of barrier, and the oriented duplex was found to be the least effective. All of the bicrystals were found to be strain-rate sensitive. Low strain-rate tests, especially on specimens with flat interfaces, exhibited extensive cross-slipping near the boundary. There was no void formation at the interfaces due to deformation. The flat boundary bicrystals did give an indication that the dislocation sources in Beta became operative, because of dislocation pile-ups formed at the boundary in the Alpha region. It was found that the slip traces found in Beta phase need not be parallel to the direction of primary slip lines in Alpha. Slip was observed in Beta away from the phase boundary on its own. Slip systems making shallower angles with the phase boundary proved to be more effective in activating dislocation sources in Beta phase than slip lines making steep angles with the boundary. At room temperature, the deformation of Alpha seemed to precede the deformation of beta phase. Beta either deformed on its own, or it deformed as a result of stress concentration near the phase boundary. Similar observations are found for specimens with boundaries inclined to the tensile axis and are presented in Appendix A.

Elevated temperature deformation behavior of two-phase bicrystals has not been studied so far. The present

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work is to fill in the understanding of the overall deformation behavior of two-phase materials.

2.3.2. Deformation of Polycrystals of Two-Phase Materials

Honeycomb and Boas[32] were the first to make detailed metallographic studies of deformed alpha-beta brass containing 40% Zn. Initial deformation occurred in alpha phase. It was only after an appreciable amount of heavy deformation in Alpha that slip was observed in beta grains. Deformed specimens of alpha-beta brass were repolished and were strained further. Slip lines again reappeared in Alpha grains. Slip traces occasionally crossed the alpha-beta phase boundaries. The grains that deformed had parallel slip traces. The orientation relationship required between the alpha and beta brass for slip propagation through the boundary to occur was $(110)_\beta \parallel (111)_\alpha$ and $\langle 111\rangle_\beta \parallel \langle 110\rangle_\alpha$. Grains satisfying these orientation conditions are called contiguous grains. More deformation was observed to occur in the vicinity of alpha-beta phase boundary, than in the interior of the beta grains.

Honeycomb and Boas[32] made an attempt to study the extent of deformation of individual alpha and beta phases by studying the recrystallization temperature of each phase through a series of annealing experiments. These studies were complicated by precipitation in the alloy, and by the order-disorder transformation in the beta phase.

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Clarebrough[33] carried out a study on silver-magnesium, soft silver grains have F.C.C. structure and hard beta phase with cesium chloride structure. These alloys were deformed at room temperature. Measurements of the recrystallization temperature of the alpha and beta phase in alloy specimens that were deformed in equal amounts indicated that, below 30 volume percent of Beta, the alpha phase deformed more than the beta phase. When the volume fraction of Beta exceeded 30 volume percent, the extent of deformation in both the phases was the same. Clarebrough and Perger[34] observed similar behavior in alloys of copper and zinc.

Suery and Baudelet[35] studied the plastic deformation and microstructure in superplastic 60% Cu-40% Zn brass at 600°C (1112°F). Specimens were extruded, and they showed fibrous structure with elongated grains. The two-phase alpha-beta brass exhibited strain-rate sensitivity. The alpha phase with fibrous shape lengthened at first parallel to the tensile axis and then developed to an approximately equiaxed structure. The shape change was attributed to what was called cell structure formation at the beginning of the deformation. The variation of the phase size and, in particular, the microstructural coarsening was aided by the deformation. The coarsening was accounted for by the possibility of combining of phases of the same nature, caused by their rotation during the deformation and by mass-diffusion along the phase interfaces.

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Baro[36] studied the superplasticity of alpha-beta brass at temperatures above T_c for Beta. From the deformation behavior of 60-40 brass, it was determined that moving dislocations also contributed to the total deformation, along with grain boundary sliding; both processes depend on one another. The essential thing is that the deformation proceeds so that no stress concentrations occur in the material. A model for superplasticity which considers grain boundary sliding, and gliding and climbing of dislocations is suggested. The model illustrates the subsequent deformation by grain boundary deformation, a shear stress is produced on a grain and the grain enclosed is deformed. The reorientation of hard phase Alpha at elevated temperature is like flotation of Alpha in soft beta phase. Continuous deformation and alignment of alpha grains along the loading axis took place. Coarsening of grains were also observed during the deformation.

2.3.3. Objectives of the Work

The deformation behavior of two-phase materials is affected by the temperature and deformation rate. So, under nearly identical deformation conditions and temperature, each phase exhibits properties of its own along with additional changes that occur during the overall deformation because of the presence of the other phase.

The primary aim of this research is to study the effect of temperature on the deformation behavior of

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two-phase bicrystals of alpha-beta brass. Beta brass experiences order-disorder transformation at elevated temperatures. The effects of disorder on the deformation behavior of such materials is also investigated. The secondary objective is to study the effect of the strain-rate on the mechanical behavior of bicrystals of alpha-beta brass at elevated temperatures and to find suitable conditions for obtaining a uniform deformation. Meanwhile, the effect of shear stresses imposed on a phase boundary could be studied on specimens having their phase boundaries oriented at different angles with respect to the tensile axis. (Observations will be presented in Appendix A of this thesis.)

It is hoped that a better understanding of the deformation behavior of two-phase materials at elevated temperatures may be helpful in developing suitable conditions for hot rolling and extruding two-phase materials.

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CHAPTER III

EXPERIMENTAL PROCEDURE

3.1. Preparation of Alpha and Beta Stock

Alpha brass stock that was used for preparation of single crystals was of commercial purity and had a nominal composition of 70% Cu and 30% Zn. Pieces, 6 cm long and 0.6 cm in diameter were cut off and machined from this stock. Beta brass stock was produced by adding zinc of 99.99^{+} purity to Muntz metal (60% Cu and 40% Zn). The nominal composition of beta brass produced was 51.5% Cu and 48.5% Zn. This alloy was produced by melting measured quantities of Muntz metal and zinc in a sealed quartz tube at 950°C. The quartz tubes were flushed with Argon before sealing to prevent any oxidation of the metal. After solidifying the melt in the quartz tubes, the stock for growing single crystals, 6 cm long and 0.6 cm in diameter, were produced by machining.

3.2. Growth of Single Crystals of Alpha and Beta Brass

Single crystals of alpha and beta brass were produced by using the Bridgeman technique. For growing single crystals, the stock material was kept in sharp-tipped graphite crucibles. These crucibles containing the stock material were kept in sealed quartz tubes to prevent zinc loss. The c

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zinc loss. The quartz tubes were kept inside mullite tubes. The volume between the quartz tube and the mullite tube was filled with refractory cement to minimize the radial temperature gradient. Such an assembly is illustrated schematically in Figure 3.

This assembly was lowered through a tube furnace to produce single crystals. The set-up used and the temperature profile of the furnace are given in Figure 4. The constant temperature zone was maintained at 1860°F for growing alpha brass crystals and 1820°F for beta brass crystals. The crystals were grown at a growth rate of 1 cm/hr. These single crystals were annealed at 1480°F for sixteen hours to remove micro-segregation, as suggested by Maddin.[6]

3.3. Growth of Two-Phase Bicrystals of Alpha-Beta Brass

To obtain the two-phase bicrystals used in this work, beta brass was melted on a substrate of a single crystal of alpha brass in a quartz tube. To minimize the transition zone that developed in this joining operation, a 60% Copper-40% Zinc brass foil, 0.2 mm thick, was placed between the alpha single crystal and the beta stock before melting. The alpha brass single crystal and the beta stock were chemically polished with 70 volume percent nitric acid - 30 volume percent water to minimize any contamination. The melting of the beta brass was achieved by heating the assembly with a Nichrome heating element.

Figure 3. Schematic Diagram of the Crucible Assembly for Growing Single Crystals of Alpha and Beta Brass.

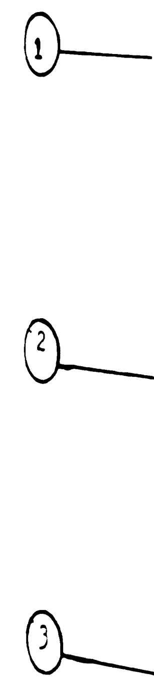
1 - Chromel wire

2 - Mullite tube

3 - Fireclay (Al_2O_3)

4 - Sealed quartz tube

5 - Alpha brass stock



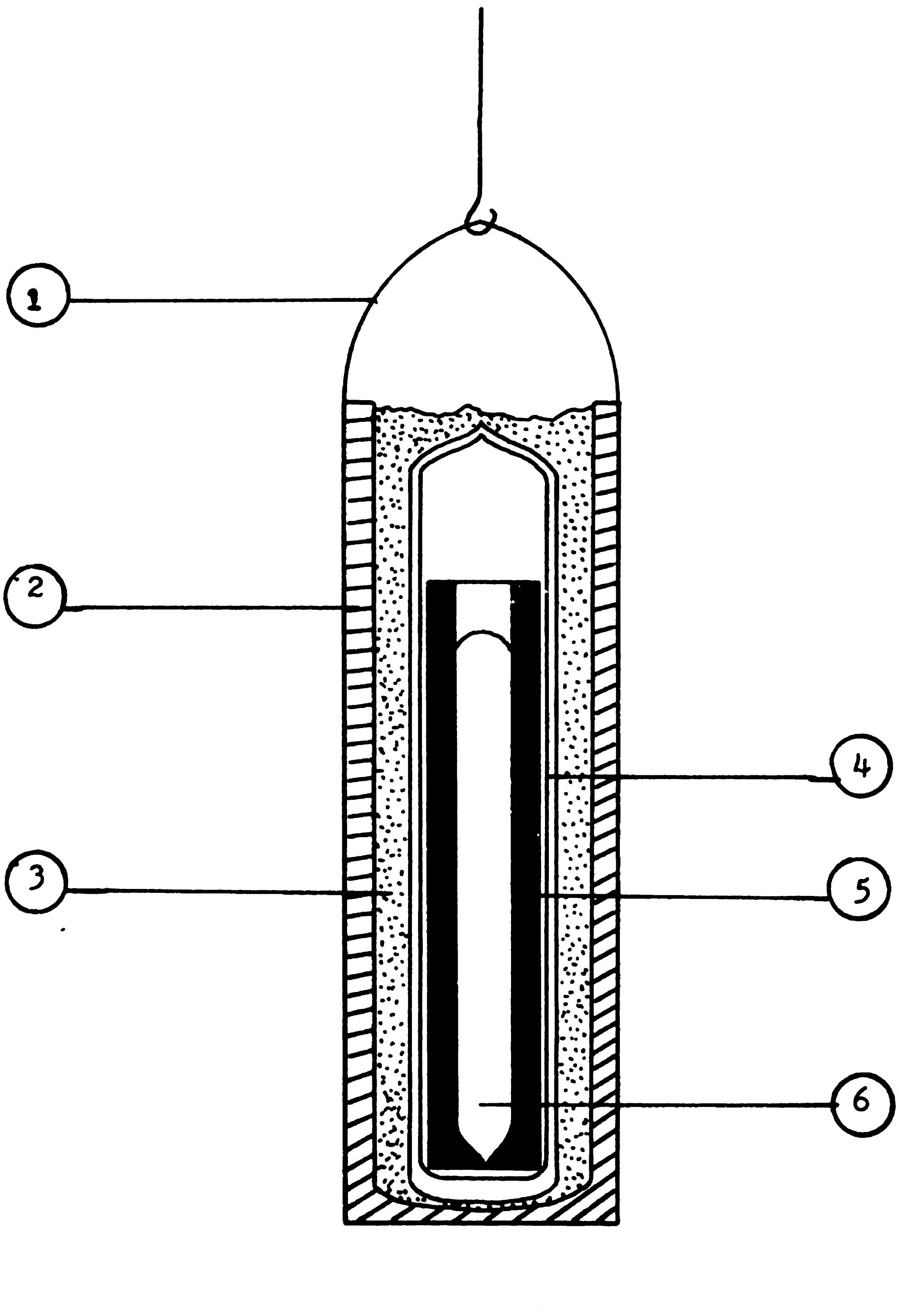


Figure 3

Figure 4. Schematic Diagram of the Furance for Growing Single Crystals of Alpha and Beta Brass.

1 - monofilament line

2 - hooks

3 - furnace insulator

4 - mullite crucible

5 - quartz tube

6 - heating element

7 - one R.P.H. motor and pulley

8 - stabilizing weight

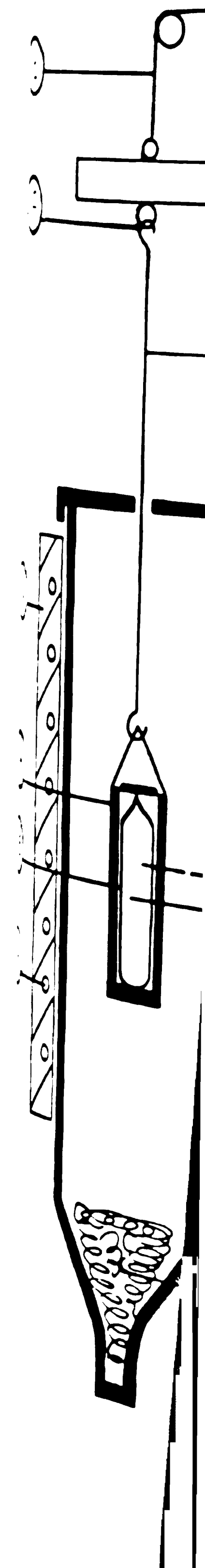
9 - wire suspension rod

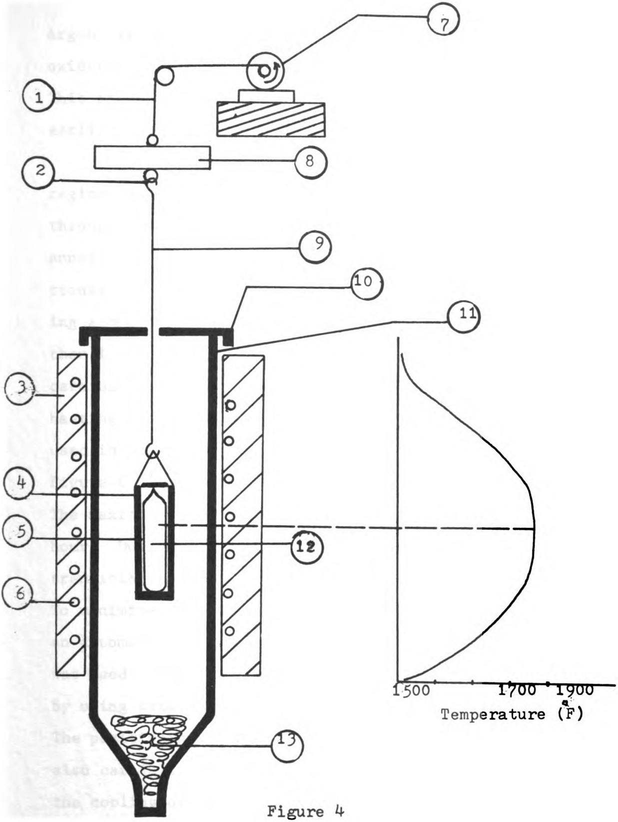
10 - fire brick top

11 - furnace tube

12 - crucible assembly

13 - glass wool





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Argon was passed through the assembly to prevent any oxidation. The set-up used is illustrated in Figure 5. This procedure is exactly the same as those used in earlier works.[3,4,5]

The transition zone formed between alpha and beta region can be eliminated by a combination of lowering through a temperature gradient and by cyclic local annealing.[3,4,5] However, it seems possible to reduce this transition zone to a flat boundary by cyclic local annealing alone provided that the transition zone is smaller than 1 mm. This successful joining can be achieved by carefully controlling the temperature to reduce superheating and by some experimentation. All the specimens used in this work were obtained by cyclic local annealing. Figure 6 illustrates the local annealing set-up used. The maximum temperature was maintained at 1450°F for one hour. During the cooling cycle the temperature of the transition zone must be brought back to room temperature. To minimize handling and to maximize the production rate, an automated cyclic annealing set up consisting of ten units was used. The temperature of each furnace was maintained by using carefully measured lengths of the heating elements. The positioning of the transition zone in the hot zone was also carefully arranged by vertical moving screws. Further, the cooling of the specimens after each cycle of heating was achieved by shutting off the power for the heating elements and blowing cool air through the quartz tubes

Figure 5. Schematic Diagram of Apparatus for Joining Beta Brass to the Single Crystal of Alpha Brass.

1 - power supply

2 - single crystal alpha

3 - refractory brick

4 - aluminum block used for heat sink

5 - argon gas inlet

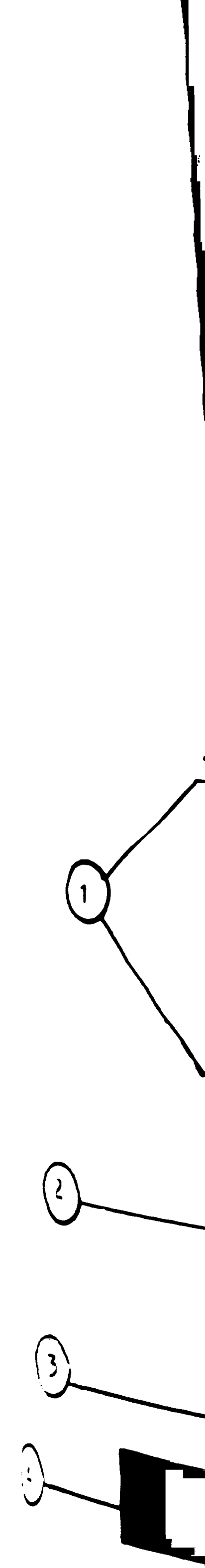
6 - beta brass stock

7 - heating element

8 - alpha - beta foil

9 - quartz tube

10 - refractory cement



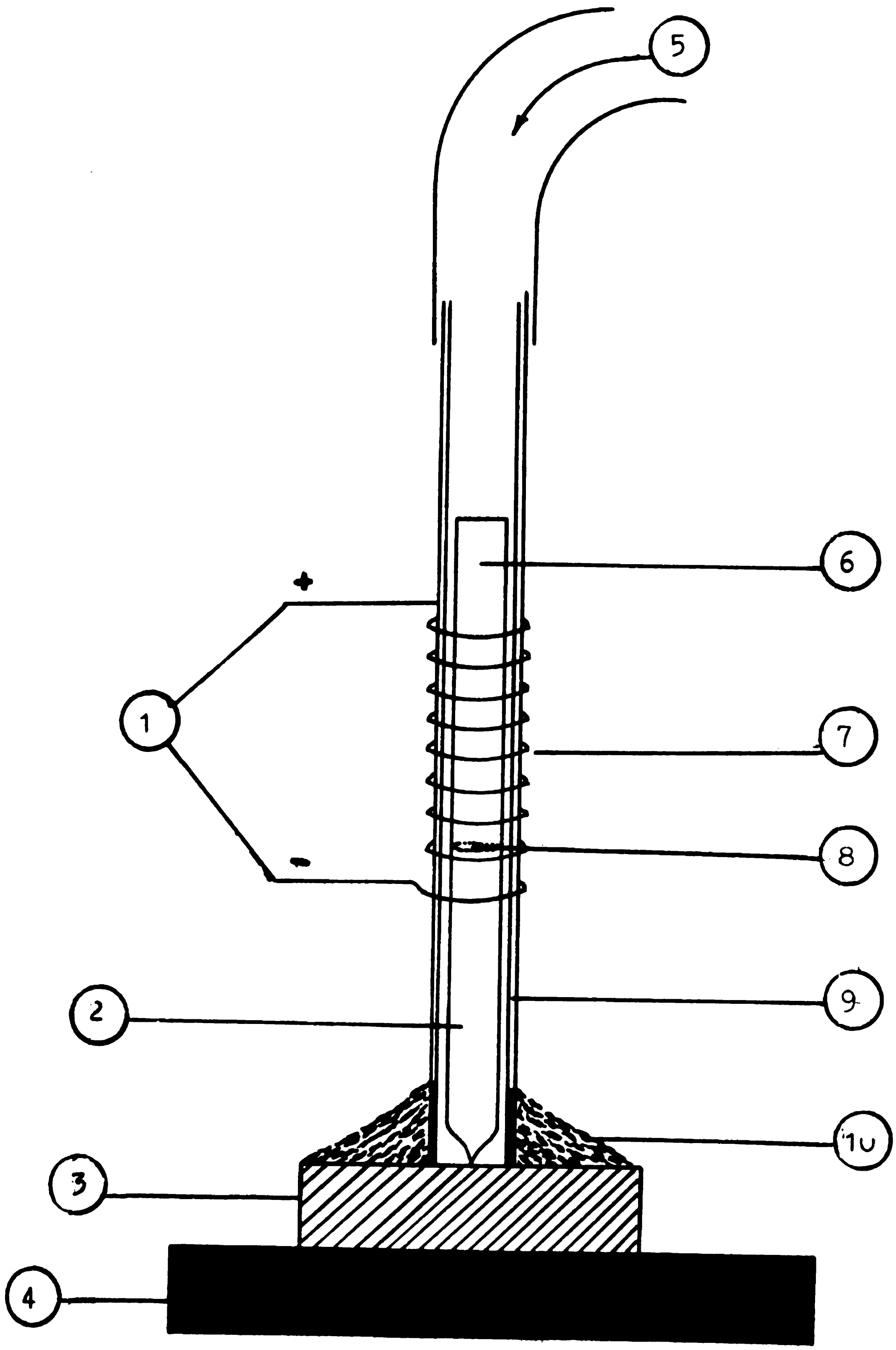


Figure 5:

Figure 6. Schematic Diagram of the Apparatus for Local Annealing.

1 - air inlet

2 - bicrystal alpha-beta brass

3 - transition zone

4 - refractory cement (Al_2O_3)

5 - heating element extension

6 - asbestos platform

7 - aluminum holding fixture

8 - quartz tube

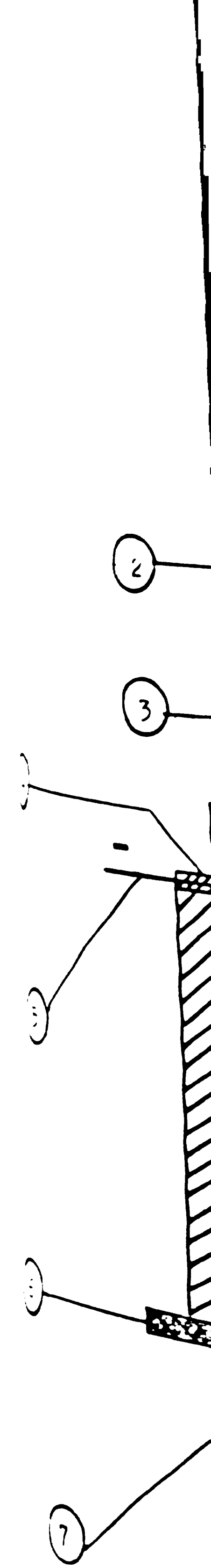
9 - heating element

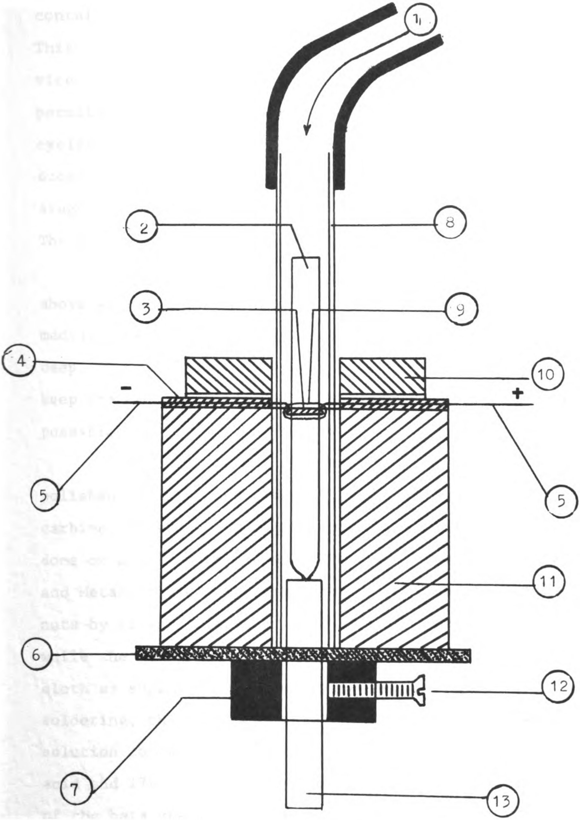
10 - fire brick

11 - porous fire brick

12 - adjusting screws

13 - aluminum holding rod





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containing the crystals for a period of five minutes. This operation was achieved by an automatic timing device. After such cooling, further cyclic annealing was permitted to proceed. Although the earlier heating cycles were about one hour long each time, they were reduced to as short as fifteen minutes towards the final stages of the preparation of flat boundary bicrystals. The schematic of the set-up used is illustrated in Figure 7.

The two-phase bicrystal specimens produced by the above were machined with a flying cutter at an intermediate speed. Although the earlier cuts were 0.003 in. deep, the final cuts were less than 0.001 in. deep to keep the mechanical damage to the bicrystals as small as possible.

After the machining operation, the specimens were polished on all sides with 240, 320, 400 and 600 grit wet carbimet papers and 600 grit wheel. The finishing was done on a velvet wheel using 1 micron diamond compound and Metadi fluid. The specimens were joined to steel nuts by silver soldering. The silver soldering was done while the rest of the specimen was kept cool with a wet cloth as suggested by Brindley, et al.[37] After silver soldering, the specimens were chemically polished with a solution containing 66% acetic acid, 17% Ortho-phosphoric acid and 17% nitric acid. Back reflection Laue pattern of the beta grain in contact with alpha single crystal was obtained on two perpendicular faces. The Miller indices

Figure 7. Schematic Diagram of Ten Units for Automatic Local Cyclic Annealing Operation.

1 - air pressure gauge

2 - automatic A.C. powered air valve

3 - air hose divider

4 - air hose

5 - asbestos coated wire

6 - local annealing unit, detail is in Figure (4)

7 - silver soldering joints

8 - insulating connectors

9 - fuse box, 15-17 amperes capacity

10 - ammeter

11 - power supply

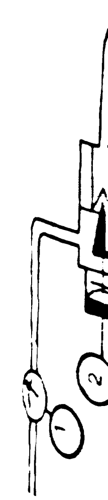
12 - variac

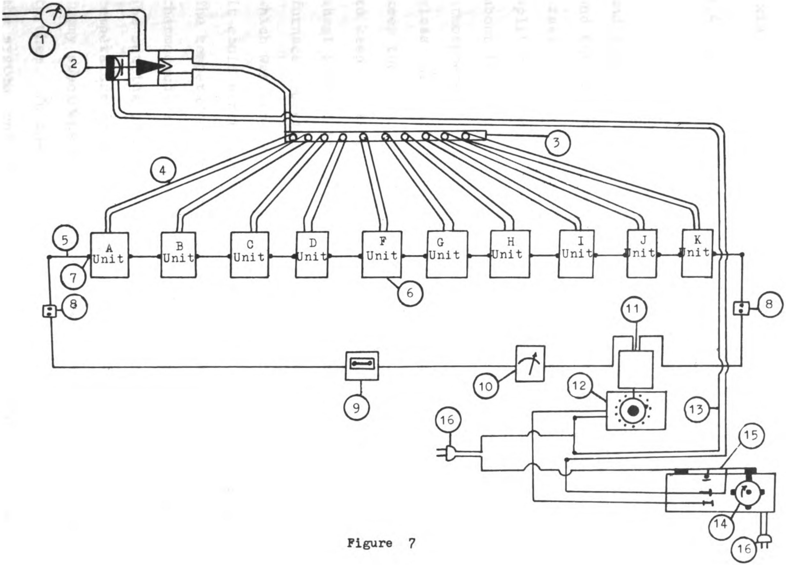
13 - connecting electric wire

14 - electric motor (1 R.P.H.) and timing pulley

15 - spring loaded micro-switch

16 - A.C. electrical input

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of the planes forming these poles were used to determine the crystallographic direction parallel to the tensile axis.

3.4. Mechanical Testing of Specimens at Elevated Temperatures

Tensile tests were performed at 700°, 800°, 900° and 1000°F. The first two of these temperatures are below and the rest are above the ordering temperature of beta brass. These temperatures were obtained by using a split tube furnace having a constant temperature zone of about four inches. The specimens were heated in an Argon atmosphere to prevent oxidation and discoloration. A glass tube with insulating seals at both ends was used to keep the Argon atmosphere around the specimen. Further, to keep the temperature of the specimen uniform, a copper sheet covered the interior surface of the furnace. The furnace temperature was controlled by a thermocouple which was kept very close to the heating element so that it could sense the temperature fluctuations very easily. The temperature of the specimen was monitored by four thermocouples which were in contact with the specimens. The setting of the controller for achieving a required temperature of the specimen was usually achieved with dummy specimens with several thermocouples silver soldered to them. Further, to reduce the heating of the rest of the system, and especially the load cell, water cooling was employ

Figure 8.

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was employed. This entire heating assembly is shown in Figure 8.

The tests were performed using an Instron testing machine with cross-head speeds of 0.02, 0.1, and 0.5 cm/min. The specimens usually reached the required test temperatures in about 45 minutes. A very small crack was left open in the split furnace so that the deformation behavior of the specimen could be studied using oblique lighting. During the heating of the specimen, care was taken to adjust cross-head position so that the specimen did not experience any stress. Similarly, at the end of the test, the load was released so that the specimen would not experience any further stress due to contraction during cooling.

The deformed samples were handled very carefully and were viewed with an optical microscope to study the deformation of Alpha, Beta, and regions near the phase boundary. These specimens were always stored in vacuum desiccators. Further, during the course of this investigation, retesting of any samples tested earlier was avoided since part of the damage in such specimens would anneal out on heating them again. So once a sample was strained and the test stopped at the same stage, it was never tested again regardless of the amount of deformation introduced in the sample.

The strain, strain-rate and stresses experienced by individual phases were also calculated for each specimen

Figure 8. Schematic Diagram of the Apparatus Used for Elevated Temperature Tests.

1 - connecting rod to load cell

2 - water hose clamps

3 - direction of water flow

4 - water cooling coil

5 - thermocouple wires

6 - glass envelope

7 - thermocouple insulator ceramic tubes

8 - thermocouple beads controlling sample temperature

9 - thermocouple beads controlling furnace temperature

10 - specimen

11 - heating elements

12 - gas inlet tube

13 - direction of argon gas flow

14 - water cooling copper coil

15 - connecting rod to cross-head

16 - upper and lower seal to the atmosphere

17 - copper sheet lining

18 - insulated electric furnace

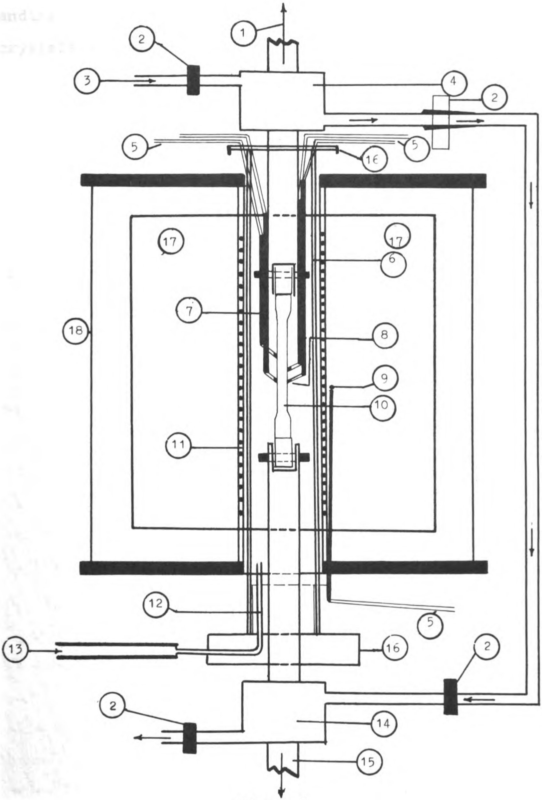

Figure 8

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at various stages of the testing so that a better understanding of the overall deformation behavior of two-phase bicrystals of alpha-beta brass could be achieved.

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CHAPTER IV

RESULTS AND DISCUSSION

4.1. Deformation Behavior of Bicrystals of Alpha-Beta Brass in the Temperature Range at which Beta Exists in the Ordered State (less than 850°F).

4.1.1 Deformation Studies at 700°F

Specimens deformed at 700°F with a cross-head speed of 0.02 cm/min had relatively high yield stresses. In all specimens, the initial deformation took place by single slip in Alpha at regions away from the boundary. As deformation progressed, these slip lines in Alpha became deeper and approached the alpha-beta phase boundary. This left a triangular shaped undeformed region in Alpha in the region near the phase boundary as shown in Figure 9(a). Upon further straining, slip in Alpha interacted with the phase boundary and resulted in deformation in beta phase. This is illustrated in Figure 9(b). Note, T.A. represents the direction of the tensile axis and "a" and "b" are used in all figures to point out alpha and beta phases.

Grain boundary sliding, and slight deformation within the grains were observed in beta regions. These features can be seen in Figure 10. Grain boundary sliding was observed in boundaries that had an orientation

Figure 9a. Interaction of Slip in Alpha with the Alpha-Beta Phase Boundary in a Specimen Tested at 700°F with a Cross-Head Speed of 0.02 cm/min.

Yield Stress of Alpha 3.5 Kg/mm^2.

Critical resolve shear stress of Alpha 1.65 Kg/mm^2.

Total strain 3.4%.

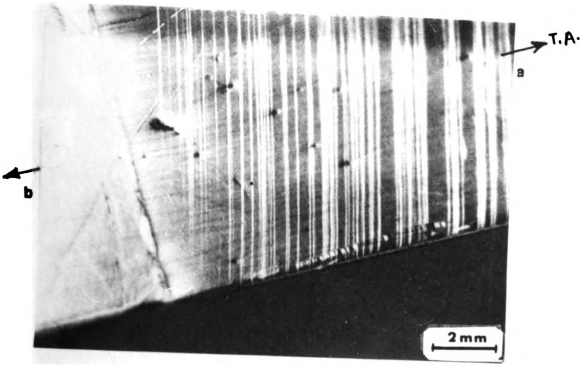

Figure 9(a)

Figure 9b. Cross-slip in Alpha Due to Interaction of Slip with the Boundary. Note: Region marked by X has extensive cross-slip.

Yield stress of Alpha 3.5 Kg/mm^2.

Critical resolved shear stress of Alpha 1.65 Kg/mm^2.

Total strain 14%.

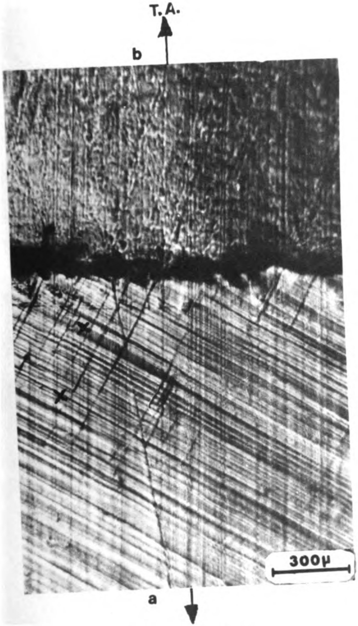

Figure 9(b)

Figure 10. Grain Boundary Sliding in a Beta Region away from the Boundary in a Specimen Tested at 700°F with a Cross-head Speed of 0.02 cm/min.

Yield stress of Alpha 3.5 Kg/mm^2.

Critical resolved shear stress of Alpha 1.6 Kg/mm^2.

Total strain 4%.

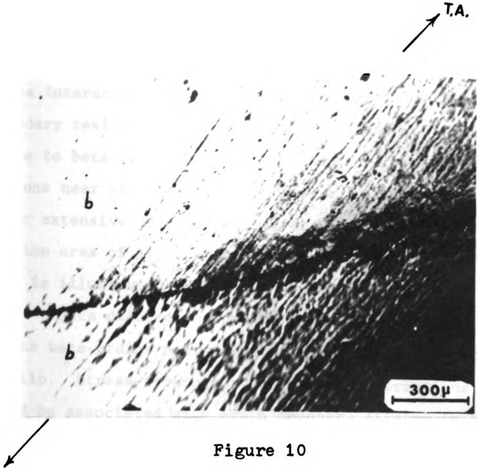

Figure 10

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close to 45° to the tensile axis. Small cracks developed at the grain boundaries in Beta due to non-uniform deformation of grains (caused by different orientations) as shown in Figures 11. These specimens usually failed by grain boundary fracture in Beta.

The initial deformation of specimens strained at 700°F with a cross-head speed of 0.1 cm/min occurred in Alpha by single slip. On further straining, slip from Alpha interacted with the phase boundary. The phase boundary resisted progression of slip from the alpha phase to beta regions. Multiple slip in alpha phase in regions near the boundary can be seen in Figure 12. After extensive deformation, severe reduction in cross section area of Alpha occurred near the phase boundary. This is illustrated in Figures 13 (a) and (b).

Beta deformed by grain boundary sliding and some of the beta grains near the phase boundary also deformed by slip. Stress-strain curves exhibited serrations that could be associated with grain boundary sliding in Beta. The early stages of these curves are similar to the stage I of resolved shear stress versus resolved shear strain curve for alpha brass single crystals.

The deformation behavior of Alpha in a bicrystal of alpha-beta brass, and the interaction of slip with the phase boundary at 700°F were similar to those observed in specimens deformed at room temperature. In all these cases alpha brass deformed first. Deformation of Beta,

Figure 11. Small Cracks Developed in Beta at the Grain Boundaries in a Specimen Tested at 700°F with a Cross-head Speed of 0.10 cm/min. Note: Region Marked with X has the Cracks.

Yield stress of alpha 3.26 Kg/mm^2.
Critical resolved shear stress of Alpha 1.5 Kg/mm^2.
Total strain 10.2%.

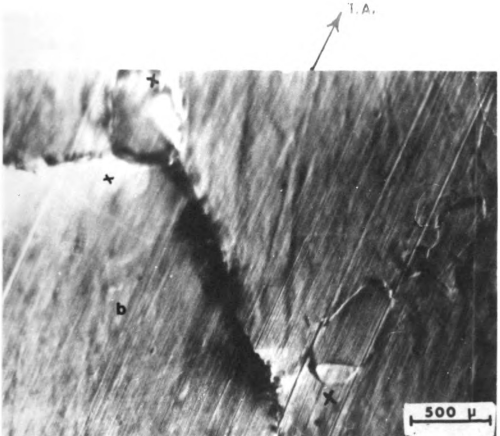

Figure 11

Figure 12. Multiple Slip in Alpha near the Phase Boundary in a Specimen Tested at 700°F with a Cross-head Speed of 0.10 cm/min.

Yield stress of Alpha 3.39 Kg/mm^2.
Critical resolve shear stress of Alpha 1.68 Kg/mm^2.
Total strain 5.6%.

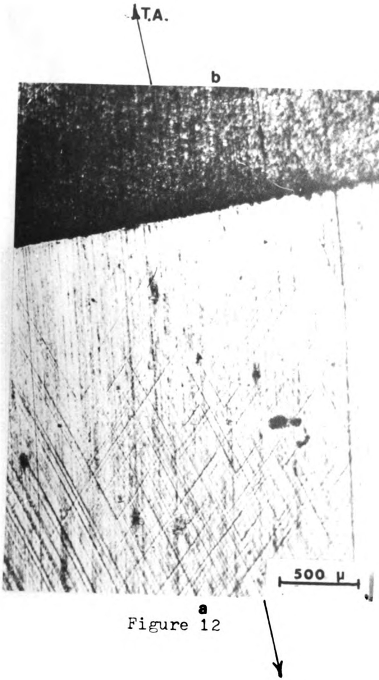

Figure 12

Figure 13a. Interaction of Slip in Alpha with the Phase Boundary in Specimen Tested at 700°F with a Cross-Head Speed of 0.10 cm/min.

Yield stress of Alpha 3.26 Kg/mm^2.

Critical resolved shear stress of Alpha 1.50 Kg/mm^2.

Total strain 10.2%.

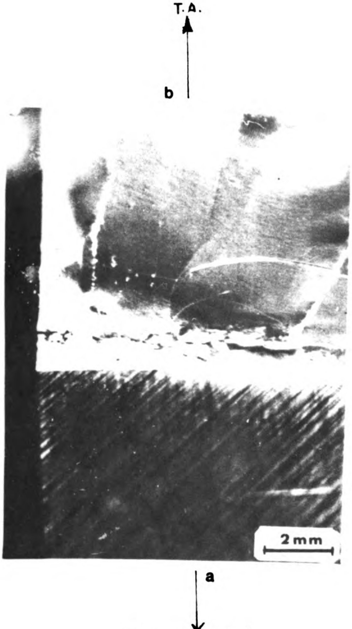

Figure 13 (a)

Figure 13b. Macrograph of a Fractured Specimen. Notice the Severe Deformation in Alpha Near the Boundary (marked by X in this specimen).

Yield stress of Alpha 3.26 Kg/mm^2.
Critical resolved shear stress of Alpha 1.50 Kg/mm^2.
Total strain 10.2%.

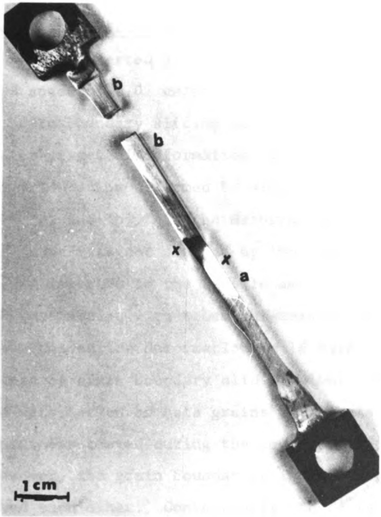

Figure 13 (b)

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however, was accommodated by grain boundary sliding as well as by slip in individual grains. Grain boundary sliding in Beta has never been observed in room temperature tests.[3,4,5] Fracture took place in beta regions at the grain boundaries due to non-uniform deformation.

4.1.2 <u>Deformation Studies at 800°F</u>

Specimens tested at 800°F and strained with a cross-head speed of 0.02 cm/min initially deformed in beta phase by grain boundary sliding as is shown in Figure 14. During later stages of deformation, beta grains with a favorable orientation deformed by slip as can be seen in Figures 15 (a) and (b). If the deformation of a favorably-oriented boundary is not stopped by the change in its orientations relative to the tensile axis in some regions, or by the interaction with other unfavorably-oriented boundaries, the entire deformation could have been accommodated by grain boundary sliding alone. In such cases the deformation of beta grains by slip is unnecessary. In the specimens tested during the course of this investigation however, the grain boundaries in Beta often intersected each other. Consequently, there was resistance for the progression of grain boundary sliding. So, beta grains deformed by slip due to shear stresses imposed by uneven deformation between beta grains. This behavior is illustrated in Figures 16 and 17. In particular, grains having smaller width and sharper edges in

Figure 14. Grain Boundary Sliding in Beta Regions away from the Phase Boundary in a Specimen Tested at 800°F with a Cross-Head Speed of 0.02 cm/mm^2.

Yield stress of Alpha 3.50 Kg/mm^2.
Critical resolved shear stress of Alpha 1.65 Kg/mm^2.
Total strain 3.4%.

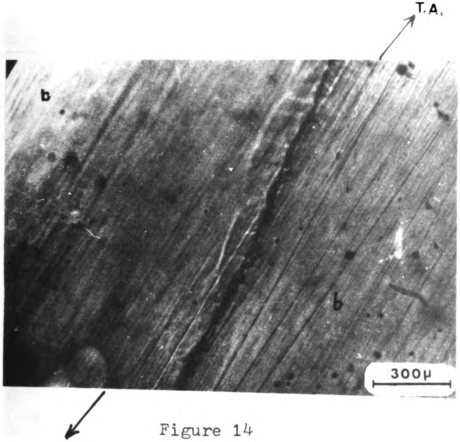

Figure 14

Figure 15. Deformation of Beta by Slip in Favorably-Oriented Grains in Regions away from the Phase Boundary in a Specimen Tested at 800°F with a Cross-Head Speed of 0.02 cm/min.

(a) Fine Slip in Beta.

Yield stress of Beta 2.25 Kg/mm^2.
Critical resolved shear stress of Beta 1.10 Kg/mm^2.
Total strain 4.9%.

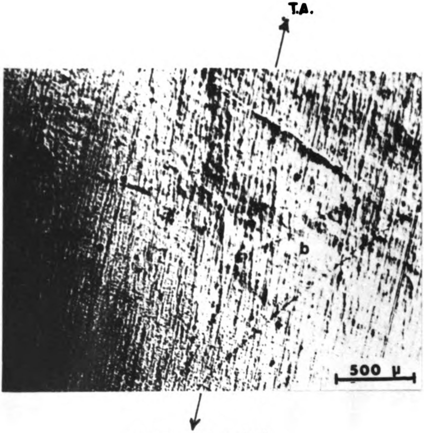

Figure 15 (a)

Figure 15b. Rumple Appearance in Beta.

Yield stress of Beta 2.25 Kg/mm^2.

Critical resolved shear stress of Beta 1.10 Kg/mm^2.

Total strain 4.9%.

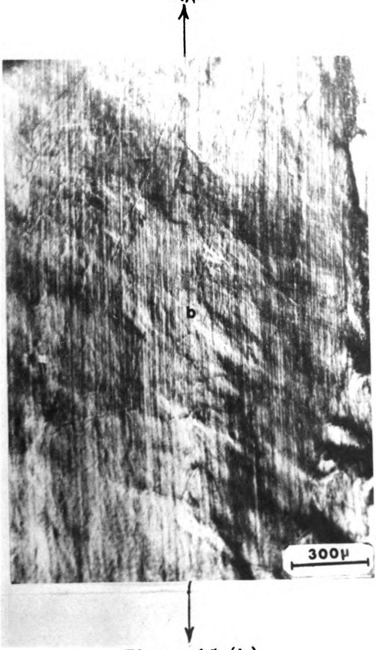

Figure 15 (b)

Figure 16. Slip Initiated by Grain Boundary Sliding in Beta Phase in a Specimen Tested at 800°F with a Cross-Head Speed of 0.02 cm/min. Here Slip is Caused by the Change in the Orientation of the Boundary.

Yield stress of Beta 2.25 Kg/mm^2.
Critical resolved shear stress of Beta 1.10 Kg/mm^2.
Total strain 4.9%.

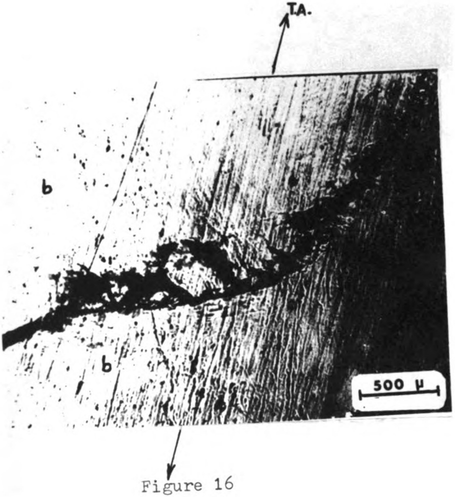

Figure 16

Figure 17. Deformation of Beta in a Most Favorably-Oriented Grain by Coarse Slip, when Grain Boundary Sliding is Unable to Accommodate the Entire Deformation, in a Specimen Tested at 800°F with a Cross-Head Speed of 0.50 cm/min.

Yield stress of Alpha 3.4 Kg/mm^2.
Critical resolved shear stress of Alpha 1.35 Kg/mm^2.
Total strain 12.9%.

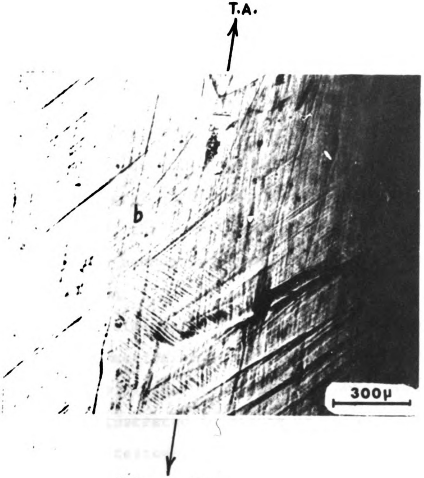

Figure 17

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the vicinity of larger grains deformed more than others. Figure 18 is an illustration of the severe deformation near the tip of smaller grains. Similar observations have been made with respect to non-uniformity in the deformation by Baro,[36] and the observations presented here are in complete agreement with his work. In Figure 19, regions in the two grains "A" and "B" adjacent to their common boundary deformed, since one of the grains had a more favorable orientation with respect to the tensile axis compared with the other. The vicinity of the grain boundary appears to be more deformed because of the deformation in beta grain "A" compared with beta grain "B". In Figure 20, three grain boundaries are meeting at a point, and grain boundary sliding caused cracks in the intersecting region.

In all specimens, Alpha did not deform plastically at all. The interaction of slip from beta region with the phase boundary did not result in any deformation in alpha phase as illustrated in Figure 21.

Specimens tested at 800°F and strained at 0.10 cm/min initially deformed in Alpha by single slip. Slip in Alpha intersected with the phase boundary as can be seen in Figure 22. A rumpled appearance of Beta in the region near the phase boundary can also be seen in this figure. Figure 23 illustrates the interaction of slip from both alpha and beta regions with the phase boundary.

Figure 18. Severe Deformation in Beta Near the Tips of Smaller Grains in a Specimen Tested at 800°F (x and y indicate the small and large grains respectively).

Yield stress of Alpha 2.25 Kg/mm^2.

Critical resolved shear stress of Alpha 1.10 Kg/mm^2.

Total strain 4.9%.

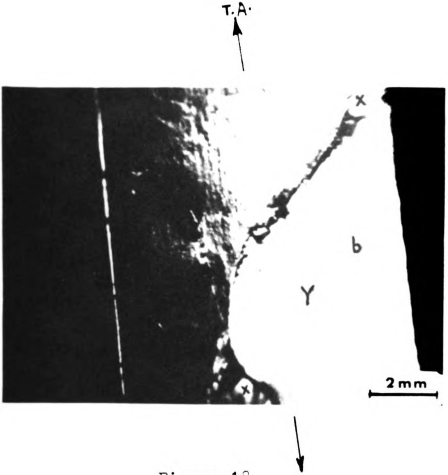


Figure 18

Figure 19. Deformation of Regions in Beta near the Grain Boundary in a Specimen Tested at 800°F with a Cross-Head Speed of 0.02 cm/min. Note: Uneven deformation in adjacent beta grains due to different crystallographic orientations to the tensile axis "A" having more favorable orientation compared to "B" and "C".

Yield stress of Alpha 2.25 Kg/mm^2.
Critical resolved shear stress of Alpha 1.10 Kg/mm^2.
Total strain 4.9%.

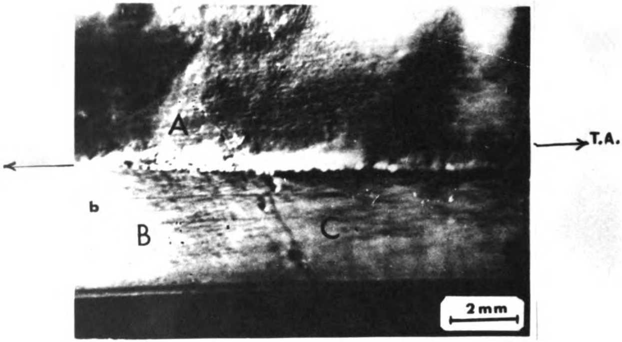


Figure 19

Figure 20. Cracking near the Junction of Three Grain Boundaries in Specimen Tested at 800°F with a Cross-Head Speed of 0.02 cm/min.

Yield stress of Alpha 2.25 Kg/mm^2.

Critical resolved shear stress of Alpha 1.10 Kg/mm^2.

Total strain 4.9%.

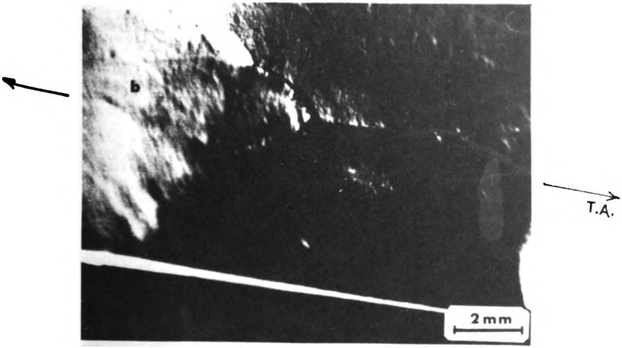


Figure 20

Figure 21. Absence of Plastic Deformation in Alpha in a Specimen Tested at 800°F with a Cross-Head Speed of 0.02 cm/min.

Yield stress of Alpha 2.25 Kg/mm^2.
Critical resolved shear stress of Alpha 1.10 Kg/mm^2.
Total strain 4.9%.

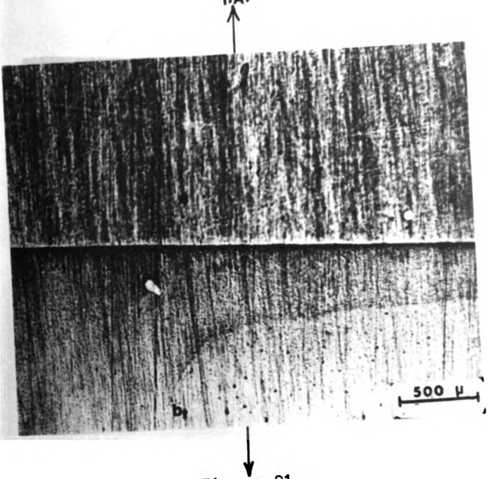


Figure 21

Figure 22. Interaction of Slip in Alpha with the Phase Boundary in a Specimen Tested at 800°F with a Cross-Head Speed of 0.10 cm/min.

Yield stress of Alpha 3.72 Kg/mm^2.

Critical resolved shear stress of Alpha 1.65 Kg/mm^2.

Total strain 3%.

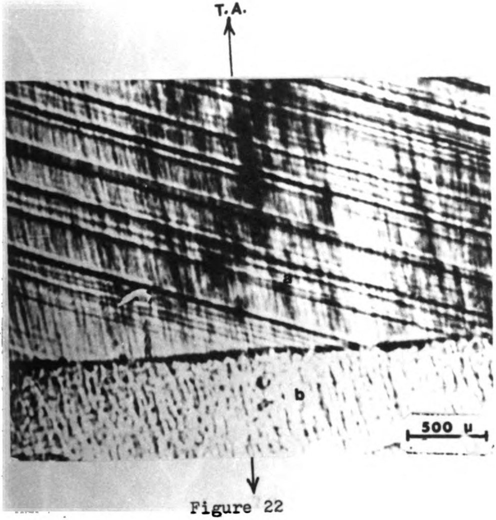


Figure 22

Figure 23. Independent Slipping of Alpha and Beta in the Region near the Phase Boundary in a Specimen Tested at 800°F with a Cross-Head Speed of 0.10 cm/min.

Yield stress of Alpha 2.25 Kg/mm^2.
Critical resolved shear stress of Alpha 1.10 Kg/mm^2.
Total strain 3%.

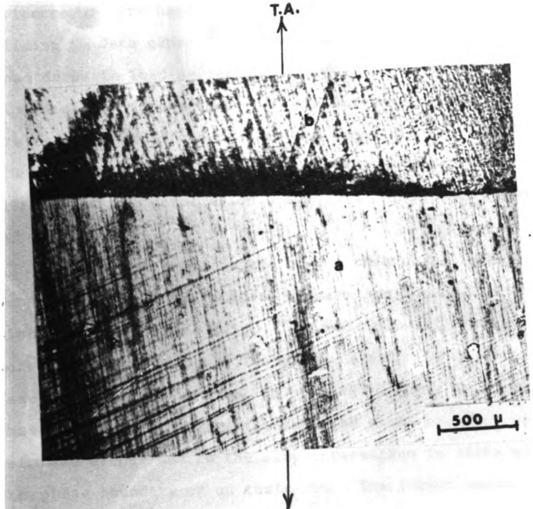


Figure 23

In this specimen Beta also deformed by grain boundary sliding. The deformation in the entire gauge length of beta phase seemed uniform. The grain boundary sliding and slip in Beta can be observed in Figure 24. Stress-strain curves of these specimens did not exhibit any evidence of work-hardening. The process of grain boundary sliding in Beta caused periodical and high amplitude load drops in load-deflection hysteresis at the early stages of the deformation. As the deformation progressed, the amplitude of the load drops decreased and the fluctuations finally stopped. From this point on, the rest of the deformation occurred by slip in alpha and beta grains.

At a cross-head speed of 0.5 cm/min and test temperature of 800°F, initial deformation took place in alpha phase by single slip. Some multiple slip lines were observed in Alpha. The multiple slipping was resulted from the interaction of slip in Alpha with the phase boundary. Beta grains near the phase boundary were deformed either due to the slip interaction in Alpha with the phase boundary or on their own. The latter seems to be more probable since no slip traces were found near the phase boundary. These specimens fractured in beta phase near the grain boundaries in Beta. The fracture progressed along those grain boundaries that were nearly perpendicular to the tensile axis.

Figure 24. Grain Boundary Sliding Leading to Severe Deformation of Beta Grains away from the Phase Boundary in a Specimen Tested at 800°F with a Cross-Head Speed of 0.10 cm/min.

Yield stress of Alpha 3.72 Kg/mm^2.
Critical resolved shear stress of Alpha 1.65 Kg/mm^2.
Total strain 3%.

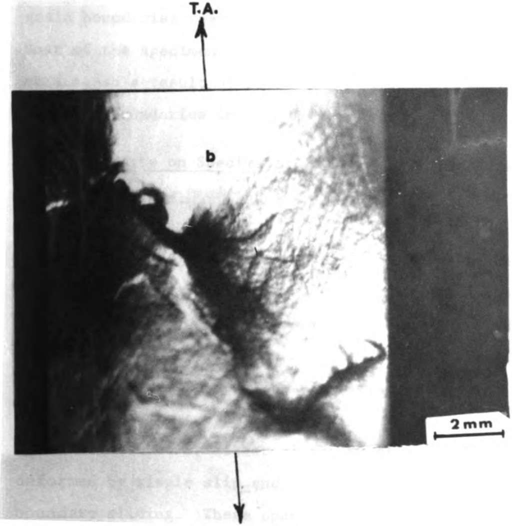


Figure 24

Observations made on specimens deformed at 800°F and at various strain-rates show that at low strain-rates following grain boundary sliding in beta phase, alpha phase deforms in single slip. Grain boundary sliding has not been very significant at high strain-rates, since grain boundaries resist deformation at high strain-rates. Most of the specimens failed in beta phase beyond 15% total strain, as a result of build-up of stress concentration at grain boundaries in Beta.

4.1.3 Remarks on Specimens Deformed at Temperatures Below the Ordering Temperature of Beta Brass (less than 850°F)

Deformation by grain boundary sliding in Beta occurs in all of the specimens. However, grain boundary is important only at high temperatures and low strain-rates. At low temperatures and high strain-rates, specimens failed in beta phase at grain boundaries in Beta. As the temperature was increased, more deformation was accommodated by either of the phases. Alpha phase deformed by single slip and beta phase deformed by grain boundary sliding. These specimens exhibited a high strain-rate sensitivity and resulted in inter-crystalline fractures. At 700°F a cross-head speed of 0.02 cm/min, and at 800°F a cross-head speed of 0.1 cm/min were found to be suitable for imposing uniform deformation in both the phases. The quantitative data for some of these tests are given in Tables I and II. Although in the results and discussion, cross-head speeds are used in Tables I and II, tabulations are made on the basis of actual strain-rates experienced by the individual phases in bicrystals of alpha-beta brass. These calculations are carried out because the lengths of each phase present in the bicrystal specimens are not the same.

$\dot{\varepsilon}_{\alpha}$ is the strain-rate in alpha phase and is calculated by dividing the cross-head speeds by length of Alpha.

ε_{α} is the strain measured in alpha phase after the test is completed.

In this table, the phase that deformed first is also included, since the yield stresses given in this table may belong either to alpha or beta phase. Similarly, strain-rates $\dot{\varepsilon}_{\beta}$ and strains (ε_{β}) were also calculated for beta phase. The average strain was calculated by using the total length of the bicrystals as the specimen gauge length. The average strain was also determined by using the gauge length of the specimens. For example, specimen P tested at 700°F and deformed at a cross-head speed of 0.02 cm/min was actually strained at a rate of 0.0066 cm/cm/min in alpha and 0.026 cm/cm/min in beta phase. The table also includes the ratio of the strain accommodated in alpha and beta phases. These calculations, in terms of strain-rate and strains in each phase, become essential because the gauge lengths of alpha and beta regions in these bicrystals were not the same. The

Table I

Quantitative Results of Bicrystals Deformed at 700°F

Specimen	C.H.S.	$\dot{\varepsilon}_\alpha$	ε_α	Y.S. Alpha	F.P.D.	$\dot{\varepsilon}_\beta$	ε_β	Y.S. Beta
T	0.1	0.0399	0.154	3.26	alpha	0.0365	0.0470	---
M	0.02	0.0074	0.0070	---	beta	0.0085	0.0080	3.50
P	0.02	0.0066	0.390	3.50	alpha	0.0100	0.0260	---

Notations used:

C.H.S. = (cm/min) cross-head speed

$\dot{\varepsilon}_\alpha$ = (cm/cm/min) strain-rate in Alpha

ε_α = (cm/cm) strain in Alpha

Y.S. Alpha = (Kg/mm^2) yield stress in Alpha

F.P.D. = the phase that deformed first

$\dot{\varepsilon}_\beta$ = (cm/cm/min) strain-rate in Beta

ε_β = (cm/cm) strain in Beta

Y.S. Beta = (Kg/mm^2) yield stress in Beta

table continued

Table 1 (continued)

Specimen	C.H.S.	$\dot{E}$	E	$\frac{\dot{\varepsilon}_\alpha}{\dot{\varepsilon}_\beta}$	$\frac{\varepsilon_\alpha}{\varepsilon_\beta}$	$\frac{\dot{\varepsilon}_\beta}{\dot{\varepsilon}_\alpha}$	$\frac{\varepsilon_\beta}{\varepsilon_\alpha}$
T	0.1	0.0178	0.102	0.93	3.2	1.07	0.31
M	0.02	0.0039	0.008	0.87	0.875	1.15	1.14
P	0.02	0.0039	0.034	0.66	1.50	1.51	0.67

C.H.S. = (cm/min) cross-head speed

$\dot{E}$ = (cm/cm/min) strain-rate of bicrystal

E = (cm/cm) average strain in bicrystal

$\frac{\dot{\varepsilon}_\alpha}{\dot{\varepsilon}_\beta}$ = ratio of strain-rate in Alpha to strain-rate in Beta

$\frac{\varepsilon_\alpha}{\varepsilon_\beta}$ = ratio of strain in Alpha to strain in Beta

$\frac{\dot{\varepsilon}_\beta}{\dot{\varepsilon}_\alpha}$ = ratio of strain-rate in Beta to strain-rate in Alpha

$\frac{\varepsilon_\beta}{\varepsilon_\alpha}$ = ratio of strain in Beta to strain in Alpha

Table II

Quantitative Results of Bicrystals Deformed at 800°F

Specimen	C.H.S.	$\dot{\varepsilon}_\alpha$	ε_α	Y.S. Alpha	F.P.D.	$\dot{\varepsilon}_\beta$	ε_β	Y.S. Beta
O	0.5	0.174	0.219	3.4	alpha	0.176	0.039	---
G	0.5	0.161	0.285	3.08	alpha	0.191	0.027	---
H	0.1	0.041	--	2.56	alpha	0.590	0.018	---
R	0.1	0.036	0.036	3.72	alpha	0.032	0.024	---
A	0.1	0.035	0.035	3.01	alpha	0.065	0.035	---
C	0.02	0.008	0.099	2.84	alpha	0.011	0.200	---
K	0.02	0.006	--	---	beta	0.007	0.100	2.24

Notations used above are explained in Table No. I.

table continued

Table II (continued)

Specimen	C.H.S.	$\dot{E}$	E	$\dot{\varepsilon}_\alpha / \dot{\varepsilon}_\beta$	$\varepsilon_\alpha / \varepsilon_\beta$	$\dot{\varepsilon}_\beta / \dot{\varepsilon}_\alpha$	$\varepsilon_\beta / \varepsilon_\alpha$
O	0.5	0.087	0.129	0.99	5.6	1.01	0.18
G	0.5	0.087	0.167	0.84	10.5	1.18	0.09
H	0.1	0.024	0	0.60	0	1.45	0
R	0.1	0.017	0.03	1.13	1.5	0.88	0.67
A	0.1	0.023	0.03	0.54	∞	1.85	0
C	0.02	0.005	0.141	0.70	0.49	1.43	2.02
K	0.02	0.004	0.280	0.97	0	1.03	∞

Notations used above are explained in Table I.

test results of specimens tested at 700° and 800°F with various cross-head speeds are presented on the same basis in Tables I and II. The notations used in these tables are explained in Table 1. Notations used in this table are also valid for the tables presented in the following sections. The stress-strain curves of some of the specimens tested at these temperatures are presented in Figures 25 and 26. The strain is calculated by using the entire gauge length of the bicrystal specimen for plotting these curves. Quantitative observations made on specimens deformed at 800°F are presented in Table III.

4.2 Deformation Behavior of Bicrystals of Alpha-Beta Brass in the Temperature Range at which Beta Exists in the Disordered State (greater than 850°F)

4.2.1. Deformation Studies at 900°F

Specimens deformed at 900°F and strained at a cross-head speed of 0.02 cm/min deformed initially by single slip in beta grains. As a result the beta regions of deformed samples had a rumpled appearance. Grain boundary sliding in beta phase was not a dominant factor. There was no indication of work-hardening in the stress-strain curves. Cracks were opened up at some of the grain boundaries in Beta. Alpha did not deform plastically at all, as can be seen in Figure 27(a), even though Beta deformed in regions away from the phase boundary as shown in Figures 27(b) and 27(c).

Figure 25. Stress-strain Curves for Bicrystals Tested at 700°F and Strained at Various Cross-Head Speeds.

Specimen H cross-head speed 0.02 cm/min

Specimen P cross-head speed 0.02 cm/min

Specimen T cross-head speed 0.10 cm/min

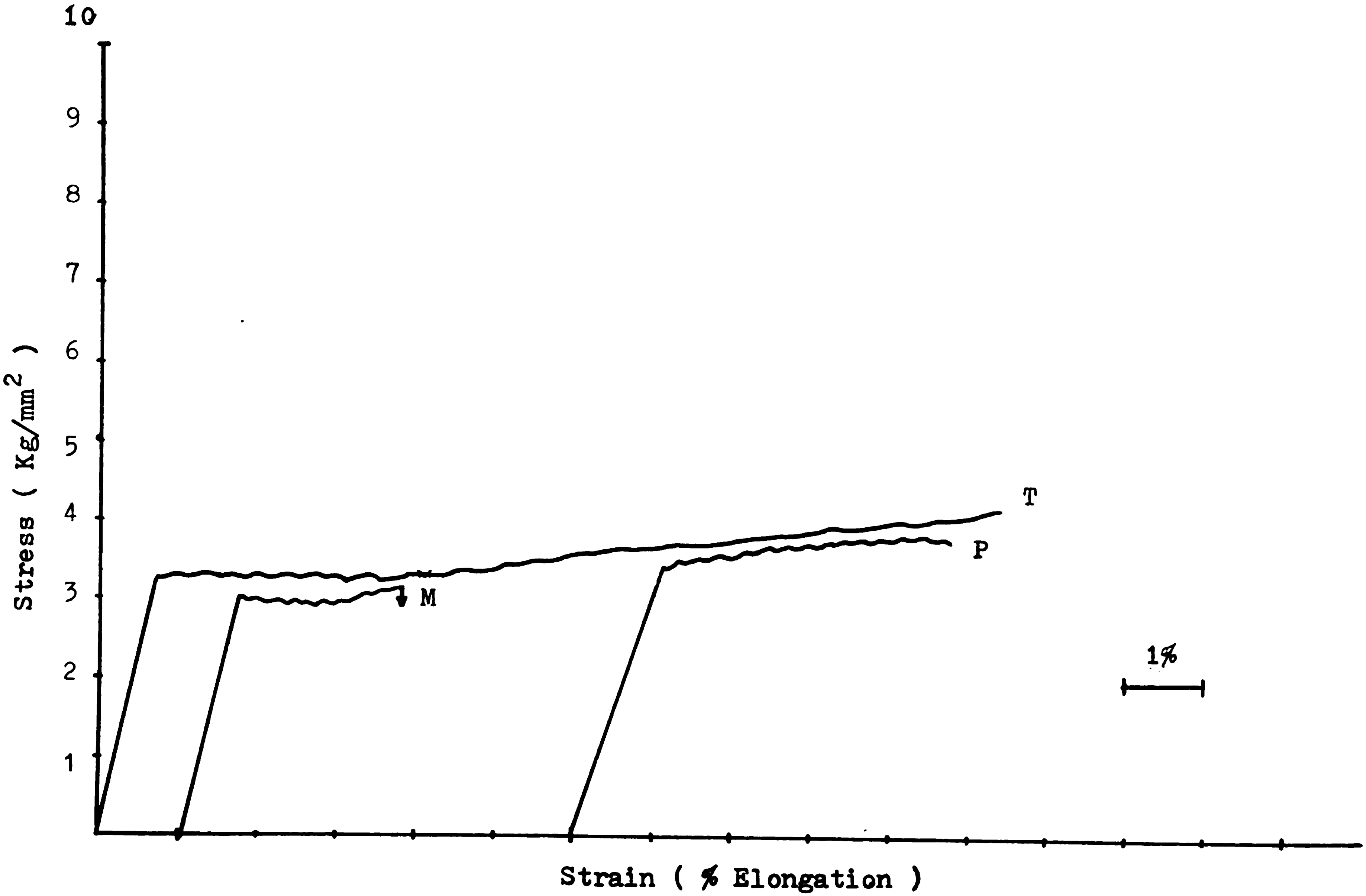


Figure 25

Figure 26. Stress-strain Curves for Bicrystals Tested at 800°F and Strained at Various Cross-Head Speeds.

Specimen K cross-head speed 0.02 cm/min

Specimen R cross-head speed 0.1 cm/min

Specimen R' cross-head speed 0.1 cm/min

Specimen O cross-head speed 0.5 cm/min

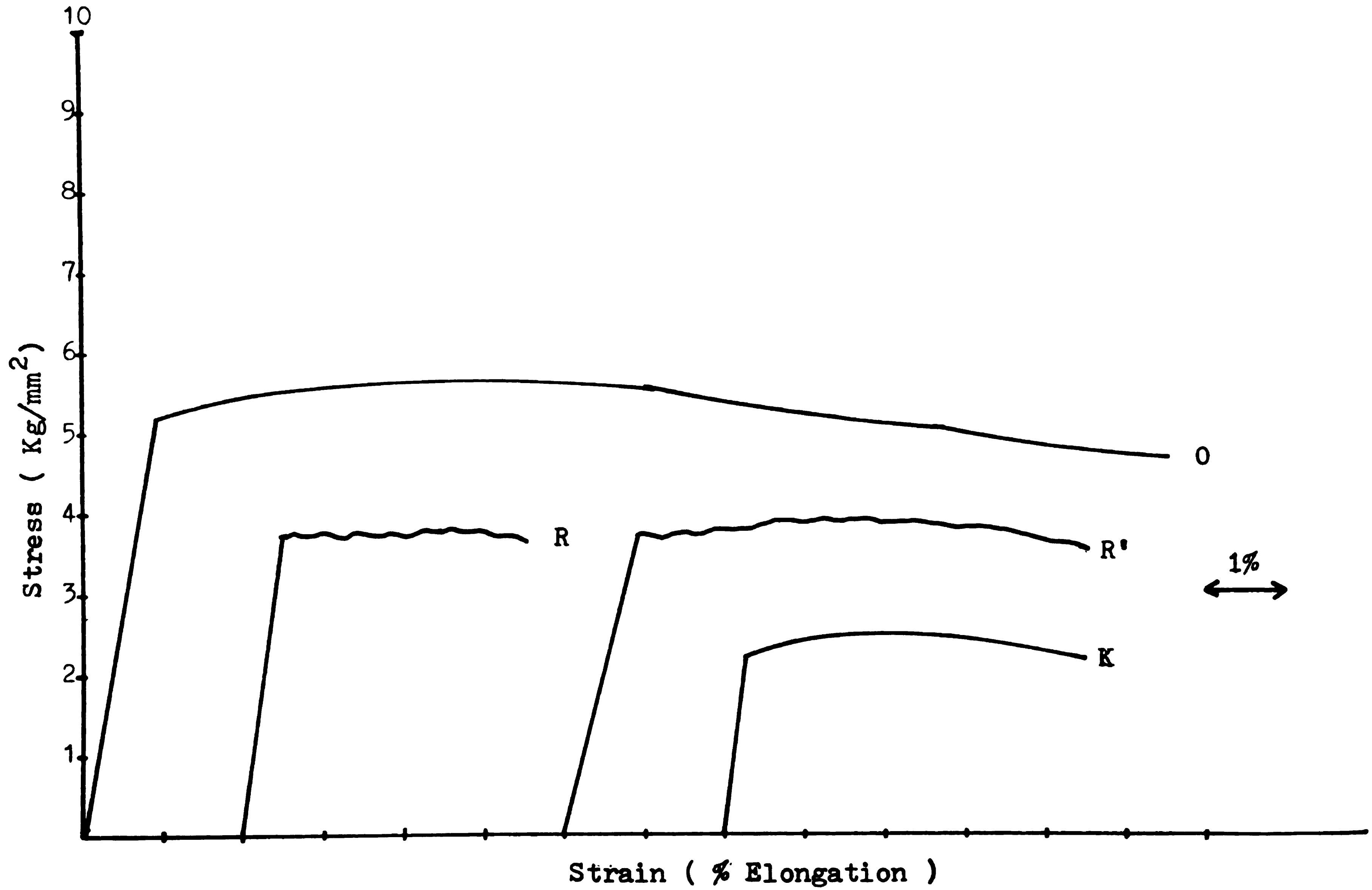


Figure 26

Table III

Qualitative Observations Made on Specimens Deformed at 800°F

C.H.S. cm/min	Deformation of Alpha	Deformation of Beta	Initial deformation	Region of failure
0.02	Single slip in Alpha Very little cross-slip.	Coarse slip and rumples	Alpha	--
0.1	Single slip approaching boundary. Uniform deformation.	None	Alpha	--
0.1	Single slip	None	Alpha	in a grain boundary in Beta.
0.5	Single slip. Non-uniform deformation.	Rumple appearance	Alpha	Beta

Figure 27. Deformation Behavior Near and away from the Phase Boundary in a Specimen Tested at 900°F with a Cross-Head Speed of 0.02 cm/min.

Yield stress of Beta 0.966 Kg/mm^2.
Critical resolved shear stress of Beta 0.48 Kg/mm^2.

(a) No observable plastic deformation in Alpha near the phase boundary.

(b) Severe deformation in beta phase away from the phase boundary

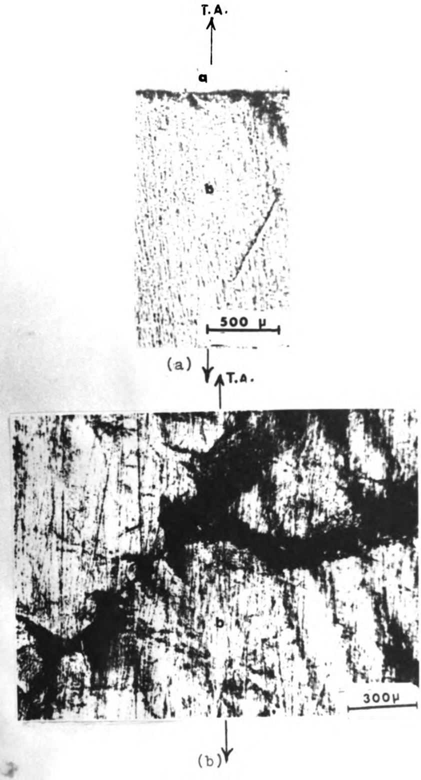


Figure 27

Figure 27 (c) Macrograph Showing Necking in Beta.

Yield stress of Beta 0.966 Kg/mm^2.

Critical resolved shear stress of Beta 0.48 Kg/mm^2.

Total strain 5.9%.

Figure 27 (c)

Observations made on deformed specimens at 900°F and strained with cross-head speed of 0.1 cm/min indicated that the entire deformation took place in Beta. Slip lines formed in Beta during deformation were clearly visible to the naked eye; these slip lines interacted with the phase boundary during further straining. Severe deformation resulted in reduction of cross sectional area in beta phase. This behavior is illustrated in Figure 28.

The observations are similar to ones made at room temperature and deformed at high strain-rates[3,4,5] (cross-head speeds of greater than 0.50 cm/min). In room temperature tests Alpha behaves in a more ductile manner and accommodates almost all of the deformation. Interaction of slip in Alpha with the phase boundary initiates slip in Beta at low strain-rates. At high strain-rates, deformation does not progress through the boundary in room temperature tests. However, in specimens tested at 900°F, the initial deformation occurs by slip in beta regions. The entire deformation at 900°F with a cross-head speed of 0.10 cm/min is accommodated in the beta region. Although slip in Beta interacted with the phase boundary, it does not activate deformation in the alpha region. Further, the absence of any work-hardening in Beta prevents the deformation of Alpha from increased applied stress.

Specimens tested at 900°F and pulled at a cross-head speed of 0.50 cm/min initially deformed in Alpha by

Figure 28. Severe Deformation in Beta Region near the Phase Boundary in a Specimen Tested at 900°F with a Cross-Head speed of 0.10 cm/min.

Yield stress of Beta 1.72 Kg/mm^2.

Critical resolved shear stress of Beta 0.85 Kg/mm^2.

Total strain 12.65%.

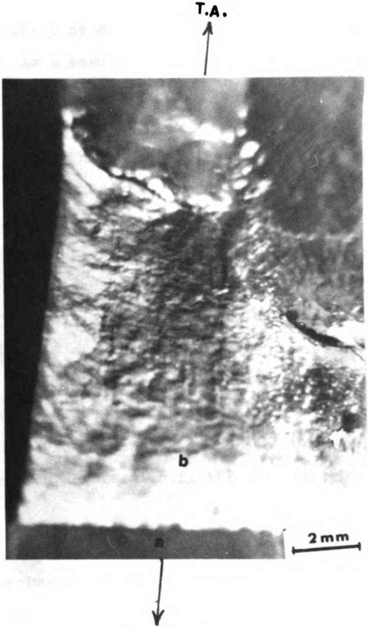


Figure 28

single slip as illustrated in Figure 29 (a). Slip lines in Alpha became deeper and wider on straining as can be seen in Figure 29 (b). The imposing strain-rate was still low enough even at this cross-head speed of 0.50 cm/min to promote climbing of dislocations by thermally activated processes. As a result of dynamic recovery process, the stress level remained low and new slip systems were not activited in alpha phase. However, Beta deformed uniformly by coarse slip. This observation can be seen in Figure 29 (c). It was found that a cross-head speed of 0.50 cm/min resulted in uniform deformation in both the phases. The slip traces present in both the phases in regions near the phase boundary can be observed in Figure 30. This deformation behavior would imply that at 900°F, a cross-head speed of 0.50 cm/min was suitable for imposing a uniform deformation in both the phases and that the phase boundary had very little or no effect on the deformation behavior of these bicrystals; whereas, the phase boundary played a more significant role at room temperature tests.[3,4,5]

4.2.2. Deformation Studies at 1000°F

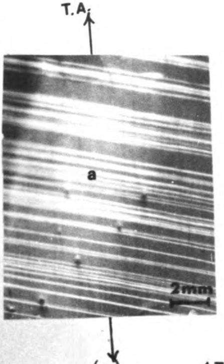
Specimens deformed at 1000°F and strained at a cross-head speed of 0.1 cm/min resulted in total plastic deformation in beta phase by coarse slip. Grain boundary sliding was not too important in the processes and most of the deformation was accommodated by beta grains. Alpha

Figure 29. Deformation of Alpha in Single Slip away from the Phase Boundary in a Specimen Tested at 900°F and Strained at a Cross-Head Speed of 0.50 cm/min.

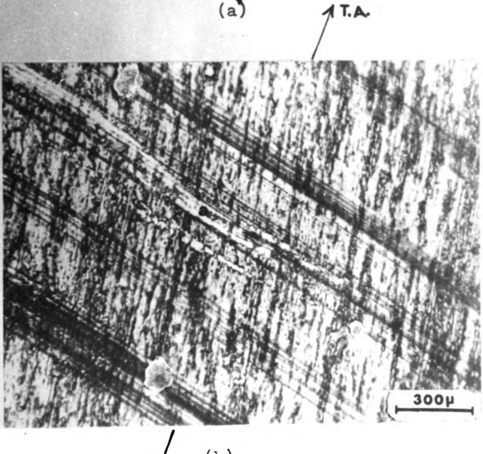
Yield stress of Alpha 4.07 Kg/mm^2.
Critical resolved shear stress of Alpha 1.98 Kg/mm^2.

(a) Macrograph of Slip in Alpha.

(b) Micrograph of Slip in Alpha.



(a)



(b)

Figure 29

Figure 29 (c). Deformation of Both the Phases Near the Phase Boundary in a Specimen Tested at 900°F with a Cross-Head Speed of 0.50 cm/min.

Yield stress of Alpha 3.73 Kg/mm^2.
Critical resolved shear stress of Alpha 1.52 Kg/mm^2.
Total strain 3%.

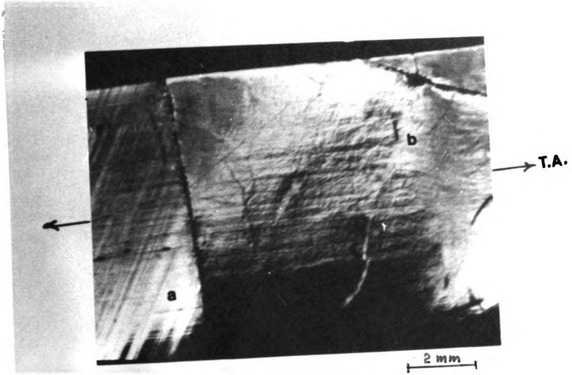


Figure 29 (c)

Figure 30. **Interaction of Single Slip in Alpha with the Boundary. Deformation in Alpha has Progressed Through the Boundary into Beta Regions.**

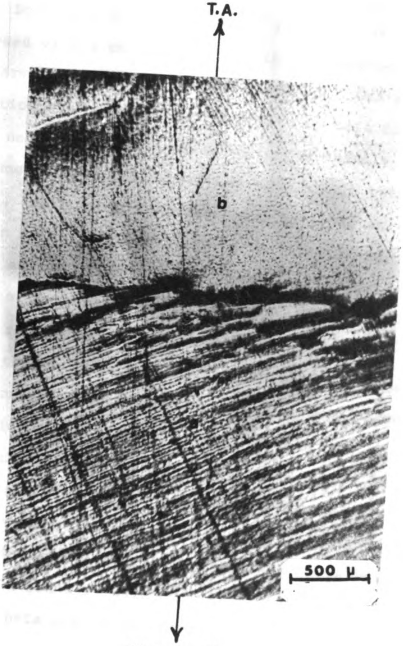


Figure 30

did not deform plastically at all. Necking took place in beta region and the process was similar to a needle point shape deformation, as can be seen in Figure 31.

Specimens tested at 1000°F and strained at a cross-head speed of 0.2 cm/min resulted in deformation of Beta. The deformed beta region appeared rumpled. Grain boundary deformation was not too significant. The beta region present near the phase boundary deformed heavily. Necking took place in beta grains and led to a needle point fracture. Although some cracks were observed at several points in Beta, the deformation at a later stage was entirely accommodated within beta grains. A delay in yielding of Beta had been observed[38] and was also detected here. Since there are no strain-hardening, because of thermally activated dynamic recovery in Beta, the deformation in Beta continued until fracture. Alpha regions did not deform during this process.

Deformation studies of specimens tested at 1000°F and a cross-head speed of 0.5 cm/min indicated that Beta phase was very soft compared with alpha phase and the entire plastic deformation took place in Beta. There was very little grain boundary sliding. At a test temperature of 1000°F, beta grains deformed with less strain-rate sensitivity compared with their grain boundaries. This non-uniformity resulted in cracks in beta phase. There was no plastic deformation in alpha phase.

Figure 31. Needle Fracture in Beta Phase, no Visible Plastic Deformation in Alpha in a Specimen Tested at 1000°F with a Cross-Head Speed of 0.10 cm/min.

Yield stress of Beta 1.10 Kg/mm^2.

Critical resolved shear stress of Beta 0.52 Kg/mm^2.

Total strain 150%.

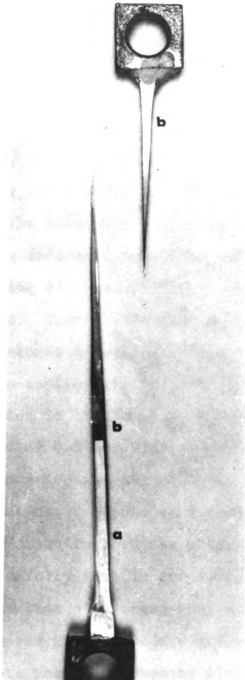


Figure 31

4.2.3. Remarks on Specimens Deformed at Temperatures Above the Ordering Temperature of Beta Brass (greater than 850°F)

Two-phase bicrystal specimens consist of four entities that can undergo deformation. These are single crystal Alpha, grains in Beta, grain boundaries in Beta, and phase boundary between Alpha and Beta. Of these four, at temperatures above ordering, the beta grains accommodate most of the deformation. There is very little grain boundary deformation in Beta at these temperatures. The deformation of these entities are highly strain-rate sensitive. There is no work-hardening of Beta at these temperatures and, as a result, Alpha will not deform unless the applied stress level is high enough to initiate deformation in it. Such a condition exists at a cross-head speed of 0.50 cm/min in specimens tested at 900°F. This observation suggests that at higher temperatures, deformation at higher cross-head speeds, should promote deformation in both the phases. However, this behavior was not fully true in the tests carried out at 1000°F. Deformation at a cross-head speed of 0.50 and 1.00 cm/min promoted grain boundary cracking in Beta; neither alpha nor beta phase deformed by slip. Results indicated that 1000°F was too high a temperature for creating uniform deformation in both the phases.

The deformation of Beta is usually made by creeping at these temperatures. A quantitative data for tests carried out are presented in Tables IV and VI. The qualitative data for the same tests are presented in Tables V and VII. The tabulations are made in a manner similar to the ones presented for specimens tested at temperatures below the ordering temperature for Beta. The stress-strain curves for some of these specimens tested at 900°F and 1000°F are given in Figures 32 and 33.

4.3 Supporting Studies on Single Crystals of Alpha and on Single and Polycrystals of Beta Brasses

In order to understand the behavior of the individual phases present in the two-phase bicrystals, a batch of single crystals of alpha and beta brass plus a large number of polycrystals of beta with grain sizes ranging from coarse to fine were tested at elevated temperatures of 700°, 800°, 900° and 1000°F and at various strain-rates by imposing cross-head speeds of 0.02, 0.1, 0.2 and 0.5 cm/min. Throughout the studies made on bicrystals at elevated temperatures, the deformation behaviors of the individual phases were considered in the interest of arriving at favorable conditions before the tests were stopped. The results obtained from the investigations made on single crystals were also helpful in understanding the features observed during the testing of bicrystals, since the bicrystal tests were never interrupted.

Table IV

Quantitative Results of Bicrystals Deformed at 900° F

Specimen	C.H.S.	$\dot{\varepsilon}_\alpha$	ε_α	Y.S. Alpha	F.P.D.	$\dot{\varepsilon}_\beta$	ε_β	Y.S. Beta
F	0.5	0.172	0.318	3.33	alpha	0.180	0.072	--
O	0.5	0.171	0.113	--	beta	0.179	0.362	4.07
B	0.1	0.053	0.039	--	beta	0.052	0.079	3.39
J	0.1	0.050	0.045	--	beta	0.044	0.199	1.72
I	0.1	0.045	0.464	--	beta	0.045	--	4.53
D	0.1	0.034	0.040	3.57	alpha	0.036	0.018	--
L	0.02	0.011	--	--	beta	0.011	0.117	0.966
E*	0.02 0.1	0.007 0.032	--	--	beta	0.008 0.038	0.343	2.31

Notations used above are explained in Table I.

*Specimen E was reintroduced to normal stress at a strain-rate five times larger than the original strain-rate.

table continued

Table IV (continued)

Specimen	C.H.S.	$\dot{E}$	E	$\frac{\dot{\varepsilon}_\alpha}{\dot{\varepsilon}_\beta}$	$\frac{\varepsilon_\alpha}{\varepsilon_\beta}$	$\frac{\dot{\varepsilon}_\beta}{\dot{\varepsilon}_\alpha}$	$\frac{\varepsilon_\beta}{\varepsilon_\alpha}$
F	0.5	0.088	0.198	0.96	4.4	1.04	0.22
Q	0.5	0.078	0.235	0.96	0.31	1.04	3.20
B	0.1	0.023	0.056	1.01	0.49	0.99	2.02
J	0.1	0.024	0.123	1.14	0.22	0.88	2.64
I	0.1	0.024	0	0.99	0	1.01	0
D	0.1	0.018	0.029	0.93	2.20	1.07	0.45
L	0.02	0.005	0.059	1.00	0	1.00	∞
E*	0.02 0.1	0.003 0.176	0.160	0.93 0.85	0	1.07 1.17	∞

Notations used above are explained in Table I.

*Explained on previous page.

Table V

Qualitative Results of Bicrystals Deformed at 900°F

C.H.S. cm/min	Deformation of Alpha	Deformation of Beta	Initial deformation	Region of Failure
0.02	None	Rumple appearance	Beta	Necked in Beta
0.1	Single slip bands. Widely spaced	Single slip	Beta	----
0.1	Single and multiple slip. Non-uniform deformation.	Rumples. Uniform deformation.	Beta	Non-uniform deformation. No failure.
0.5	Single slip. Uniform deformation. Extensive deformation.	Rumples and slip in Beta. Slip Slip near boundary also.	Alpha	Beta grain boundary.

Table VI

Quantitative Results on Specimens Deformed at 1000°F

Specimen	C.H.S.	$\dot{\varepsilon}_\alpha$	ε_α	Y.S. alpha	F.P.D.	$\dot{\varepsilon}_\beta$	ε_β	Y.S. beta
S	0.5	0.161	0	-	beta	0.187	0.075	2.78
D'	0.5	0.161	0	-	beta	0.179	0.136	2.90
A'	0.2	0.068	0	-	beta	0.129	1.084	1.36
N	0.1	0.046	0	-	beta	0.049	0.280	0.56

Notations used above are explained in Table I.

table continued

Table VI (continued)

Specimen	C.H.S.	$\bar{\dot{\varepsilon}}$	$\bar{\varepsilon}$	$\frac{\dot{\varepsilon}_\alpha}{\dot{\varepsilon}_\beta}$	$\frac{\varepsilon_\alpha}{\varepsilon_\beta}$	$\frac{\dot{\varepsilon}_\beta}{\dot{\varepsilon}_\alpha}$	$\frac{\varepsilon_\beta}{\varepsilon_\alpha}$
S	0.5	0.087	0.035	0.87	0	1.16	∞
D'	0.5	0.085	0.065	0	0	0	∞
A'	0.2	0.045	0.376	0	0	0	∞
N	0.1	0.023	0.136	0.94	0	1.06	∞

Notations used above are explained in Table I.

Table VII

Qualitative Results From Specimens Tested at 1000°F

C.H.S.	Deformation of Alpha	Deformation of Beta	Initial deformation	Region of Failure
0.5	None	Extensive deformation	Beta	Needle point fracture in Beta.
0.1	None	Grain boundary sliding	Beta	At grain boundary in Beta.
1.0	None	Grain boundary deformation	Beta	Grain boundary cracking

Figure 32. Stress-Strain Curves for Bicrystals Tested at Elevated Temperature of 900°F and Strained at Various Cross-Head Speeds. (Data points were transferred from an instron chart.)

Specimen L cross-head speed 0.02 cm/min

Specimen J cross-head speed 0.1 cm/min

Specimen Q cross-head speed 0.5 cm/min

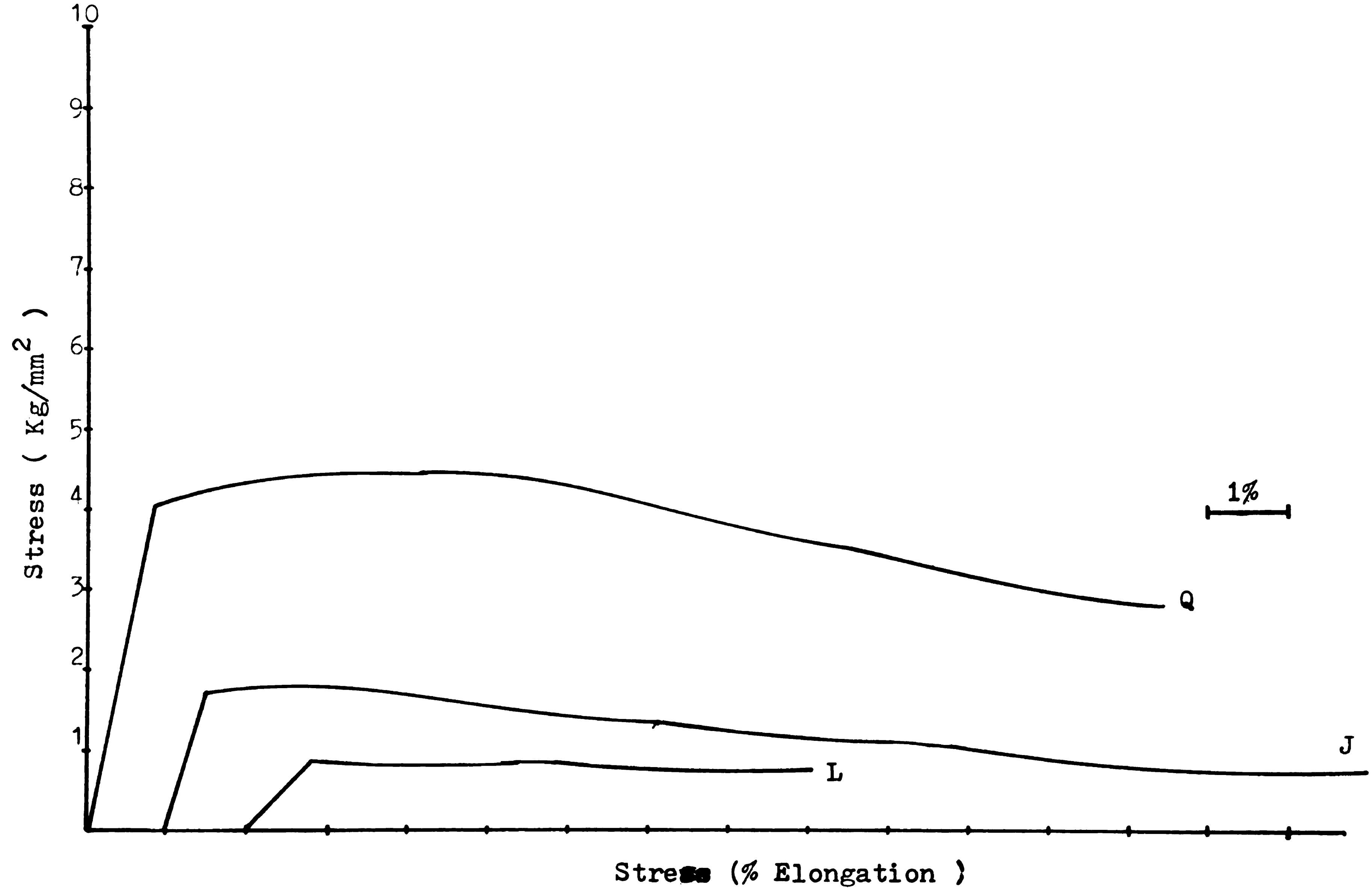


Figure 32

Figure 33. Stress-Strain Curves for Bicrystals Tested at Elevated Temperature of 1000°F and Strained at Various Cross-Head Speeds. (Data points were transferred for an instron chart.)

Specimen N cross-head speed 0.1 cm/min

Specimen A' cross-head speed 0.2 cm/min

Specimen D' cross-head speed 0.5 cm/min

Specimen S cross-head speed 0.5 cm/min

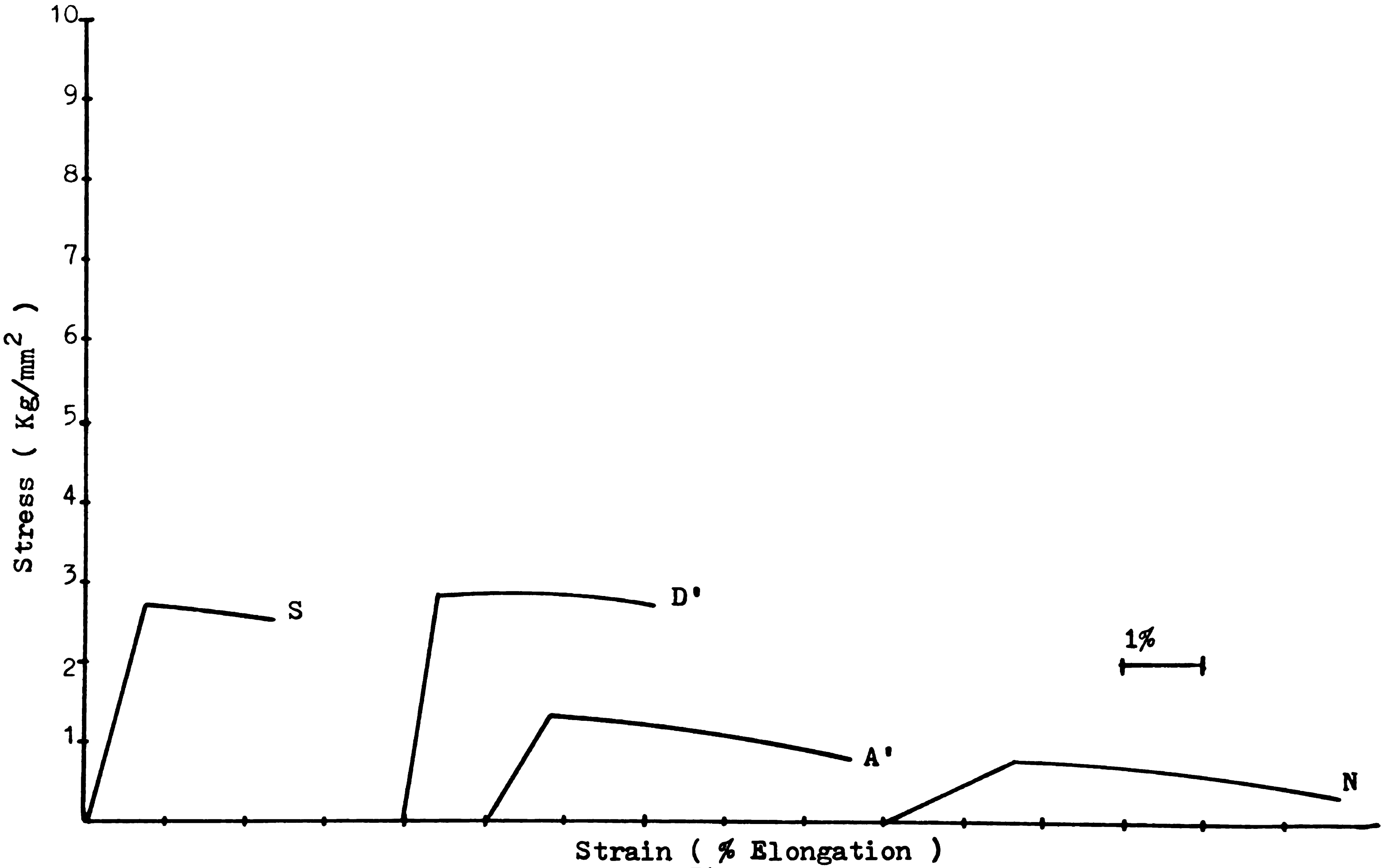


Figure 33

4.3.1 Deformation Behavior of Alpha Brass Single Crystals at Elevated Temperatures

A batch of alpha brass single crystals were tested at 700°, 800°, 900° and 1000°F. Cross-head speeds of 0.02, 0.1 and 0.5 cm/min were employed in these tests. The data on the yield stresses (critical resolved shear stresses) for alpha brass single crystal specimens deformed at 800°F with different cross-head speeds were used to draw a graph in terms of yield stresses (critical resolved shear stresses) as a function of strain-rates. A similar plot was also made from the data for specimens deformed at 900°F. From these graphs, yield stresses (critical resolved shear stresses) for strain-rates of 0.01, 0.02, 0.05, 0.10, 0.15 and 0.20 cm/cm/min at 800° and 900°F were obtained and this data is presented in Table VIII.

Based on this table a graph combining the effect of temperature and strain-rate on yield stress and critical resolved shear stress of alpha brass single crystals is sketched in Figure 34. These graphs illustrate how both temperature and strain-rate influence the yield stresses of alpha brass single crystals. The increase in the temperature and the decrease in the strain-rate, allow the mobile dislocations to move out of a crystal with less resistance. The thermal energy provided by the test temperature, helps to untangle the dislocations

Table VIII

Yield Stress (critical resolved shear stress) of Single Crystals of Alpha-Beta Brass at Various Strain-Rates and Temperatures

Strain rate (cm/cm/min)	Temperature (°F)	Yield Stress (Kg/mm^2)	Critical Resolved Shear Stress (Kg/mm^2)
0.01	800	2.85	1.10
0.01	900	2.70	1.05
0.02	800	2.90	1.15
0.02	900	2.75	1.10
0.05	800	3.10	1.30
0.05	900	2.80	1.20
0.10	800	3.45	1.55
0.10	900	3.10	1.40
0.15	800	3.80	1.80
0.15	900	3.30	1.60
0.20	800	4.15	1.95
0.20	900	3.50	1.75

Figure 34. Yield Stress (Solid Lines) and Critical Resolved Shear Stress (Dotted Lines) for Alpha Brass Single Crystals Tested at Temperatures Ranging from 700° to 1000°F. (Data points are presented in Table VIII.)

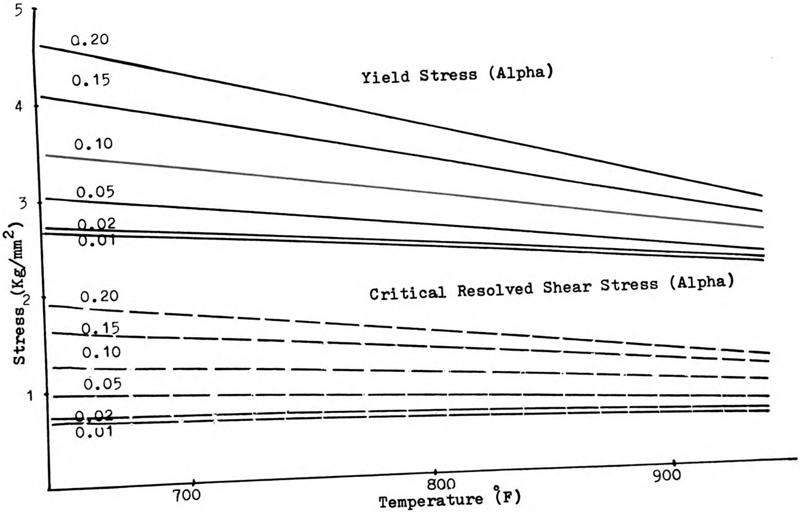

5
4
3
2
1
Stress (Kg/mm2)
0.20
0.15
0.10
0.05
0.02
0.01
Yield Stress (Alpha)
0.20
0.15
0.10
0.05
0.02
0.01
Critical Resolved Shear Stress (Alpha)
700
800
900
Temperature °(F)

Figure 34

locked in their positions, resulting in further deformation. At medium and low strain-rates, where interaction of dislocations with each other is limited, specimens deformed easily.

4.3.2 Deformation Behavior of Beta Brass Single Crystals at Elevated Temperatures

Beta brass single crystals were tensile tested by imposing cross-head speeds of 0.02, 0.10 and 0.50 cm/min. The test temperatures used were 700°, 800°, 900°, and 1000°F. Two of the temperatures are above and two temperatures are below the ordering temperature of Beta. A procedure similar to the one used for alpha brass single crystals was used to obtain the yield stress at strain-rates of 0.01, 0.02, 0.05, 0.10, 0.15 and 0.20 cm/cm/min at these temperatures. The results of such an analysis are presented in Table IX. Based on this table a graph combining the effect of temperature and strain-rate on yield stress of beta brass single crystals is sketched in Figure 35. These graphs illustrate how both temperature and strain-rate influence the yield stresses of beta brass single crystals.

Due to the fact that beta brass experiences structure change above 850°F (454°C) from Cesium-Chloride (CsCl) to B.C.C., a drop in yield stress occurs. Figure 35 illustrates the results obtained from tensile tests performed on single crystals of beta brass at various

Table IX

Yield Stress (critical resolved shear stress) of Single Crystals of Beta Brass at Various Strain-Rates and Temperatures

Strain-rate (cm/cm/min)	Temperature (°F)	Yield Stress (Kg/mm^2)	Critical Resolved Shear Stress (Kg/mm^2)
0.01	700	3.35	1.65
0.01	800	2.75	1.30
0.01	900	1.95	0.92
0.01	1000	0.40	0.20
0.02	700	3.55	1.70
0.02	800	2.90	1.40
0.02	900	2.00	0.95
0.02	1000	0.42	0.21
0.05	700	4.32	2.15
0.05	800	3.40	1.65
0.05	900	2.15	1.05
0.05	1000	0.45	0.22
0.10	700	5.50	2.20
0.10	800	4.15	2.04
0.10	900	2.35	1.13
0.10	1000	0.50	0.24
0.15	700	6.70	3.32
0.15	800	4.95	2.40
0.15	900	2.55	2.25
0.15	1000	0.55	0.25
0.20	700	7.78	3.90
0.20	800	5.57	2.80
0.20	900	2.75	1.35
0.20	1000	0.60	0.29

Figure 35. Yield Stress for Beta Brass Single Crystals Tested at Elevated Temperature and Strained at Various Strain-Rates. (Data is presented in Table IX.)

(a) Strain-rate 0.01 cm/cm/min

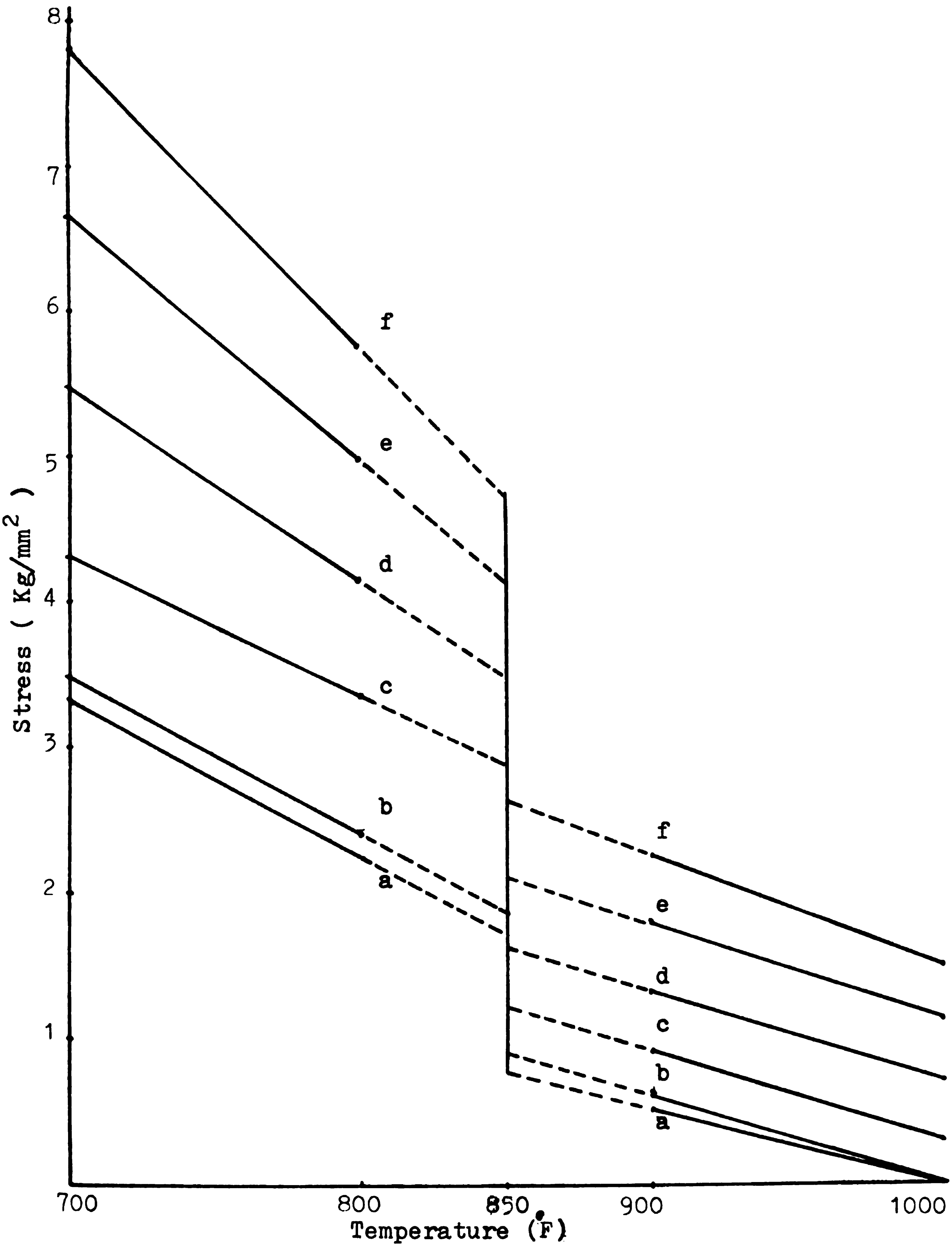
(b) Strain-rate 0.02 cm/cm/min

(c) Strain-rate 0.05 cm/cm/min

(d) Strain-rate 0.10 cm/cm/min

(e) Strain-rate 0.15 cm/cm/min

(f) Strain-rate 0.20 cm/cm/min



Stress (Kg/mm2)
8
7
6
5
4
3
2
1
f
e
d
c
b
a
f
e
d
c
b
a
700
800
850
900
1000
Temperature (°F)

Figure 35

temperatures and strain-rates. A drop in yield stress for all strain-rates in the vicinity of order-disorder transformation can also be seen in Figure 35. It was observed that at low strain-rates, the deformation proceeded by creeping at temperatures above T_c. Coarse slip lines were seen on deformed samples of beta brass single crystals strained at low strain-rates and fine slip lines on specimens strained at high strain-rates. Often heavily deformed beta brass single crystals had a rumpled appearance.

4.3.3 Deformation Behavior of Beta Brass Polycrystals at Elevated Temperatures

Coarse-grained polycrystals of beta brass similar to the ones in the two-phase bicrystals of alpha-beta brass, and some bicrystals of beta brass were tested at 700°, 800°, 900°, and 1000°F by imposing cross-head speeds of 0.02, 0.1 and 0.5 cm/min. Bicrystals either deformed by grain boundary sliding or fractured after 15% deformation. Grain boundary orientation with respect to the tensile axis played an important role. Those grain boundaries parallel or perpendicular to the tensile axis underwent very little deformation and specimens failed at a grain boundary perpendicular to the tensile axis. This is in agreement with the observations made by Mera.[39] Some of the beta grains appeared intact at their centers. Beta grains were reoriented along the tensile axis whenever the deformation proceeded beyond 30% straining. In bicrystals, normally the grain with most favorable-orientation with the tensile axis deformed and the deformed grain accommodated the rest of the deformation. The stress-strain curves did not exhibit any work-hardening. Beta grains as well as grain boundaries, resisted deformation at high strain-rates. Cracks developed at grain boundaries at high strain-rates in bicrystals of beta brass. This phenomenon was observed quite frequently in specimens that underwent incompatible deformation between grains and the grain boundaries.

Medium-grained beta brass specimens exhibited more fractures, since the probability of formation of cracks were increased by the presence of more grain boundaries oriented unfavorable to the tensile axis. Once a local yielding was initiated, the specimen deforms very little and fractures abruptly. Fine-grained specimens of beta brass deformed easier than coarse-grained specimens or single crystals at elevated temperatures. Grain boundaries at room temperature tend to contribute to the strength of beta brass polycrystal; but at elevated temperatures the presence of more grain boundaries weakened the material and polycrystalline beta brass specimens deformed very easily. A procedure similar to the one used in the case of alpha brass single crystal was utilized to obtain yield stress at strain-rates of 0.01, 0.02, 0.05, 0.10, 0.15 and 0.20 cm/cm/min, at various temperatures. The data are presented in Table X, and the graph that

Table X

Yield Stresses of Polycrystals of Beta Brass at Various Strain-Rates and Temperatures

Strain-rate (cm/cm/min)	Temperature (°F)	Yield Stress (Kg/mm^2)
0.01	700	3.55
0.01	800	3.35
0.01	900	2.35
0.01	1000	0.40
0.02	700	3.75
0.02	800	3.55
0.02	900	2.45
0.02	1000	0.42
0.05	700	4.50
0.05	800	4.10
0.05	900	2.65
0.05	1000	0.45
0.10	700	5.70
0.10	800	5.05
0.10	900	3.05
0.10	1000	0.50
0.15	700	6.85
0.15	800	6.00
0.15	900	3.40
0.15	1000	0.55
0.20	700	8.05
0.20	800	6.95
0.20	900	3.80
0.20	1000	0.60

illustrates the data in this table is given in Figure 36. Polycrystals of beta brass also undergo a structure change in the vicinity of 850°F (454°C). The structure changes occur both in the grain boundaries and beta grains simultaneously, since grain boundaries are found to undergo continuous structure change with changing of test temperatures. However, the overall effect of temperature on the yield stress is similar to that of beta single crystals in the vicinity of 850°F.

4.3.4 Deformation Behavior of Two-Phase Muntz Metal at Elevated Temperatures

A number of specimens made out of Muntz metal were tested at 700°, 800°, 900° and 1000°F and were subjected to various strain-rates by imposing cross-head speeds of 0.02, 0.1 and 0.5 cm/min. It was learned that the yield stress is directly related to the strain-rate and inversely to the test temperature. This agrees with the results obtained for most of polycrystalline two-phase materials.[36] Table XI contains the yield stresses for specimens tested at various temperatures and cross-head speeds.

4.4 Factors that Influence Uniform Deformation of Both the Phases Present in a Two-Phase Material

One of the main objectives of this work is to find conditions for obtaining uniform deformation in both the phases present in a two-phase material. The

Figure 36. Yield Stress for Beta Brass Polycrystals Tested at Elevated Temperatures and Strained at Various Strain-Rates. (Data points are presented in Table X.)

(a) Strain-rate 0.01 cm/cm/min

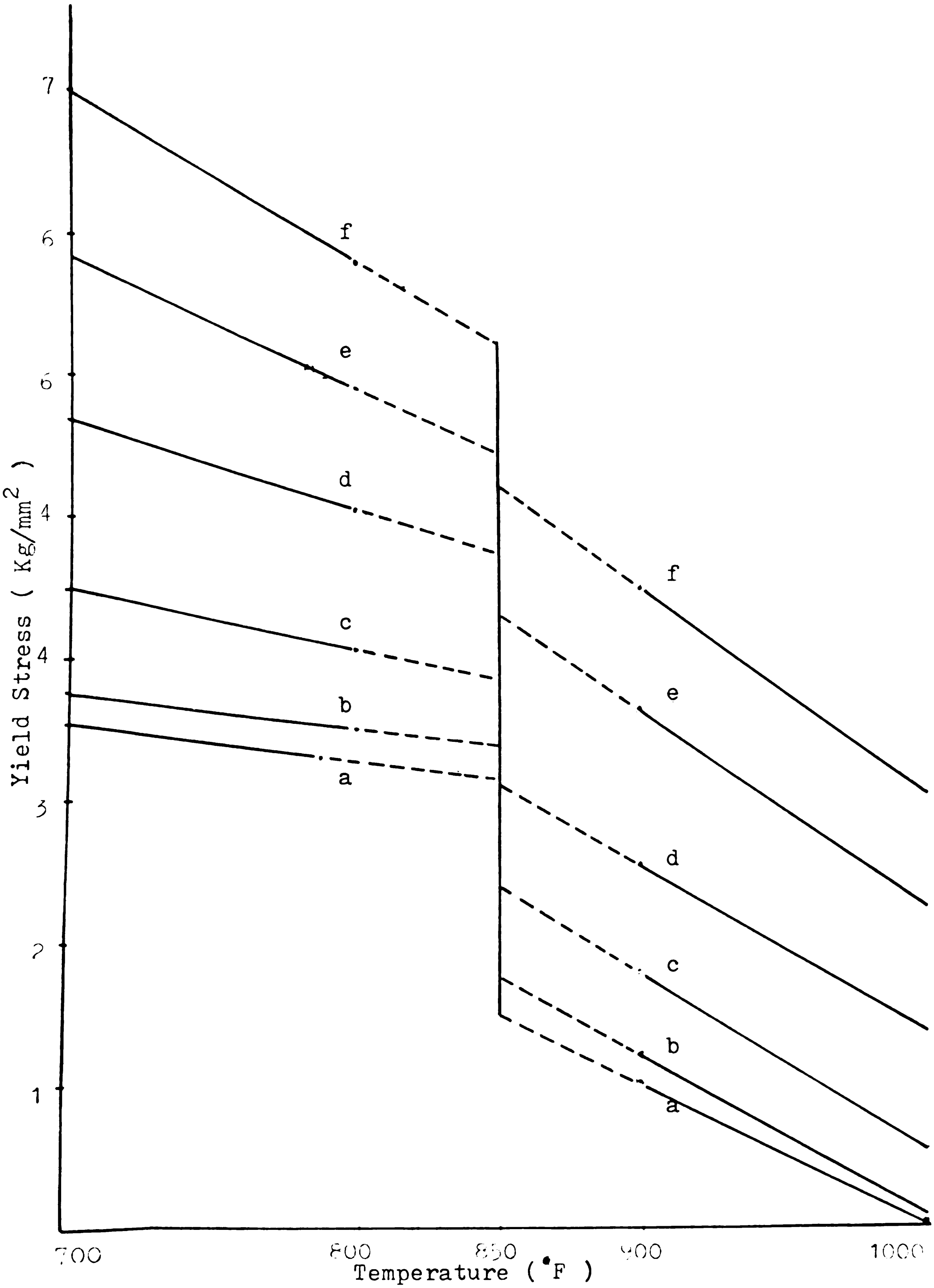
(b) Strain-rate 0.02 cm/cm/min

(c) Strain-rate 0.05 cm/cm/min

(d) Strain-rate 0.10 cm/cm/min

(e) Strain-rate 0.15 cm/cm/min

(f) Strain-rate 0.20 cm/cm/min



7
6
6
4
4
3
2
1
Yield Stress (Kg/mm2)
f
e
d
c
b
a
f
e
d
c
b
a
700
800
850
900
1000
Temperature (°F)

Figure 36

Table XI

Yield Stresses of 60-40 (Cu-Zn) Commercial Brass (Muntz metal) at Various Cross-Head Speeds and Temperatures

Temperature (°F)	Cross-Head Speed (cm/min)	Strain-rate (cm/cm/min)	Yield Stress (Kg/mm^2)
700	0.10	0.0394	9.35
800	0.50	0.197	8.64
800	0.10	0.036	6.4
800	0.02	0.008	4.35
900	0.50	0.212	6.01
900	0.10	0.039	4.55
1000	0.10	0.042	3.96

total number of available slip systems in each of the phases present in two-phase materials are different. As a result, temperature sensitivity, strain-rate sensitivity and phase transformations have been utilized to promote uniform deformation in both the phases. Phase transformation can give rise to different deformation characteristic of a material. Obtaining a uniform deformation in two-phase bicrystals of alpha-beta brass at various temperatures is a complicated process. There are many factors that control the deformation behavior of a two-phase material. For instance there is always a possibility that one phase may still be deforming in an elastic manner while the other is deforming plastically. Generally, at different test temperatures, one phase remains harder than the other. By imposing a proper crosshead speed (strain-rate) on a bicrystal, a uniform deformation could be achieved provided the strain-rate sensitivity of both the phases are different. The role of the phase boundary in the overall deformation behavior of such two-phase bicrystals can also be a controlling factor.

One can formulate the following relationships with respect to two-phase samples (having alpha and beta phases) having both the phases in series and deformed under uniaxial loading:

$$\Delta \ell = \Delta \ell_{\alpha} + \Delta \ell_{\beta} \qquad (1)$$

$$\bar{\varepsilon} L = \varepsilon_{\alpha} L_{\alpha} + \varepsilon_{\beta} L_{\beta} \qquad (2)$$

where: $\Delta \ell$ is the total elongation of the specimen,

$\Delta \ell_{\alpha}$ is the elongation in alpha phase,

$\Delta \ell_{\beta}$ is the elongation in beta phase,

$\bar{\varepsilon}$ is the average strain in specimen,

L is the total length of specimen,

ε_{α} is the strain in alpha phase,

L_{α} is the length of alpha phase,

ε_{β} is the strain in beta phase and

L_{β} is the length of beta phase.

Moreover, at the start of the test when the cross-section area remains constant throughout the sample, one can assume that:

$$\bar{\sigma} = \sigma_{\alpha} = \sigma_{\beta} \quad , \tag{3}$$

where $\bar{\sigma}$ is the stress applied to the total specimen

σ_{α} is the stress applied to the alpha phase, and

σ_{β} is the stress applied to the beta phase.

By combining both equations (2) and (3) one can arrive at:

$$\bar{\sigma}\bar{\varepsilon}L = \sigma_{\alpha}\varepsilon_{\alpha}L_{\alpha} + \sigma_{\beta}\varepsilon_{\beta}L_{\beta} \quad . \tag{4}$$

Equation (4) holds at the start of the tensile test, since all the components of the bicrystal specimens have a uniform cross-section area. If both the phases have identical mechanical properties and suitable conditions in terms of temperature, strain-rate and other influencing factors, then both the phases would undergo a uniform deformation. Otherwise, one has to consider the following factors:

(1) At test temperature below T_c alpha phase is softer than Beta, even though Beta deforms slightly by grain boundary sliding mechanism. However, at low test temperatures and strain-rates, some regions in Alpha deform first, then Alpha experiences strain-hardening. Consequently, the stress level rises in the specimen and ultimately Beta deforms either by the interaction of slip from alpha phase with the phase boundary or deforms on its own.

(2) Above T_c beta phase is the softer phase. Once Beta deforms, either Beta fractures after 15% straining (low temperature, high strain-rates, and presence of grain boundaries perpendicular to tensile axis) or deforms by creeping (high temperature and low strain-rates) until the end of the test. At temperatures above T_c Beta exists in B.C.C. structure and it has been found out Beta deforms easier than alpha phase. Since there is no way of producing the stress level in the alpha phase (hard phase) required to deform it. It is only by applying a high strain-rate to beta phase at the start of the test that the deformation of alpha phase becomes possible. The stress $\bar{\sigma}$ (average stress), σ_α (stress in Alpha) and σ_β (stress in Beta) are related to each other in a complicated way and can be written as follows:[40]

$\bar{\sigma}(T,F,\bar{\dot{\epsilon}},A(t),\lambda)$ is related to $\sigma_\alpha(T,F,\dot{\epsilon}_\alpha,A_\alpha(t),\lambda_\alpha)$

and $\sigma_\beta(T,F,\dot{\epsilon}_\beta,A_\beta(t),\lambda_\beta)$

where: T is the test temperature,

$\bar{\sigma}$ is the total applied stress,

F applied tensile force or load,

$\bar{\dot{\epsilon}}$ average applied strain-rate and

$A(t)$ instantaneous cross-section area,

λ represents parameters such as grain boundary orientation, crystallographic orientation of individual phases with respect to the tensile axis, structural changes during the testing and other minor factors involved in machining and handling of specimens. This is a complicated fundamental relationship and only by experimental observations can one arrive at suitable conditions for obtaining uniform deformation. However, such a process will be extremely complicated and one may not be able to find a correct solution at all.

In this work, a large number of single and polycrystals of Alpha (single crystals) and Beta (single and poly) brass crystals were tensile tested at 700° and 800°F below T_c, and at 900° and 1000°F above the T_c by imposing cross-head speeds of 0.02, 0.1, 0.2, and 0.5 cm/min. Graphs were made from tabulated results by using yield stresses obtained for specimens deformed at various temperatures and strain-rates. These graphs were instrumental in predicting some of the deformation behavior of a bicryatals tested at certain temperatures.

Once the dimensions of the specimens are measured (length of each phase and cross-section area), a cross-head speed can be selected. Thus, by imposing a desirable strain-rate, both the phases would experience a uniform plastic deformation.

Cross-head speeds of 0.02 cm/min. at 700°F, 0.1 cm/min. at 800°F, and 0.5 cm/min. at 900°F, resulted in producing a uniform deformation in both the phases for specimens with total length of about 2 inches and each phase having nearly the same length.

Grain boundary deformation was observed in most cases where the test temperature was high and the strain-rate was low. Margolin[40] made his studies on beta brass bicrystals and as a result, formulated a set of equations relating the total applied stress to the stress away from the grain boundary and stress in the grain boundary deformation zone. From the slip behavior, it was found out that progression of slip from one grain to the other depends upon the compatibility of strain within the grain and the grain boundaries. The yield stress of specimens tested at various temperatures and strain-rates were affected by the nature and the deformation behavior of grains and their grain boundaries.

Formation of cracks at grain boundaries in beta brass was observed quite frequently.[35] Below the ordering temperature, stress concentration due to incompatibility of strain between grains and their boundaries resulted, since the grain boundary sliding is limited to the first 15% of straining. This limit of 15% strain for grain boundary sliding was obtained from observations based on load fluctuation during the course of the tests. When load amplitude fluction faded out, grain boundary sliding process was assumed to have stopped. Moreover, the grain boundary sliding mechanism accommodates almost all of the strain in specimens with few large grains. Whenever smaller grains and, especially, interpenetrating grains with intersecting grain boundaries were present, grains were sheared from opposite directions. Upon further straining, grains deformed to some extent and ultimately due to excessive stress concentration build-up, grains were lifted from their positions and cracks developed. Figure 18 illustrates the interference of grain boundary sliding with grains. In Figure 37, it is shown that slip initiated at point "A" and slip progressed to point "B". A reorientation of grains with respect to the tensile axis also results. Fracture usually occurs where the local stress concentration becomes excessive. Reorientation of grains and rounding off of the edges of grains in two-phase materials were also cited by Baro.[36]

By imposing different cross-head speeds on various bicrystal specimens tested at different temperatures a uniform deformation and/or deformation in both

Figure 37. A Simulated Model Illustrating the Interaction of Grain Boundary Sliding with Beta Grain.

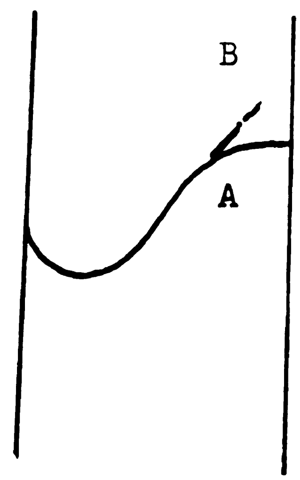


Figure 37

the phases were set up in order that the role of the phase boundary could be studied. The deformed samples revealed in most cases that each phase had deformed on its own without the influence of the other phase. In rare occasions slip progressed through the phase boundary and it produced deformation zones in the other phase. Deformation in either phase proceeded by single slip and phase boundary did not play an important role in the deformation behavior of two-phase bicrystal of alpha-beta brass at elevated temperatures.

The structure change in beta phase in the vicinity of T_c, from order (superlattice structure) below T_c to disordered B.C.C. above T_c, a thermally aided process, plays an important role in the deformation behavior of bicryatals. By increasing the test temperatures, the short range order is decreased and no more superlattice structure would be present. Dislocations can move more easily and contribute to the ease of deformation.

A series of tests were performed on polycrystals of two-phase alpha-beta brass at test temperatures of 700°, 800°, 900° and 1000°F and strained by using cross-head speeds of 0.02, 0.1, and 0.5 cm/min. Below T_c Alpha is soft and beta phase is hard. Above T_c Beta becomes softer than Alpha. In either case, the softer phase plays the role of a matrix. It was also cited by Baro[36] that alpha platelettes, originally present with sharp corners in a beta matrix of a two-phase brass specimen, were rounded off. Some of the photomicrographs showed that the alpha phase became equiaxed and parallel to the tensile axis. The coarsening of microstructure of alpha phase took place due to assembly of phases of the same nature, caused by their rotation during the deformation and also by mass-diffusion along the phase interfaces. At temperatures above T_c where Alpha is harder than Beta, the softer phase (Beta) accommodates almost all of the deformation and reorientation of alpha grains becomes possible.

In order to comprehend the deformation behavior of two-phase alpha-beta brass at elevated temperatures, a description and discussion of a series of simulated models based on actual specimens are helpful. Since shape, orientation, and location of grains with respect to each other and their neighboring grains play important roles during the deformation, step by step additions to some of the simpler models are made in the following sections. The most important factors involved are pointed out and are reviewed below.

Coarse grain alpha brass tested at room temperature and also at elevated temperatures deforms easily. At low strain-rates, specimens did not fracture up to 100% elongation. Grain boundary sliding at elevated temperatures accommodates the straining. At room temperature, grain boundaries resist progression of slip from within one grain to the neighboring grains. Dislocation pile-ups against the grain boundaries were observed.[36]

Coarse grain beta brass tested at room temperature exhibited very little deformation, and specimens fractured before 5% deformation. Grain boundary sliding was observed in beta brass at temperatures above 112°F (50°C). Strain incompatibility between beta grains and their boundaries often led to fracture at grain boundaries.

The deformation of alpha and beta brass bicrystals with three possible grain boundary orientations are discussed next.

Grain boundaries could have three extreme possible orientations with respect to the tensile axis. Figure 38 (a), (b) and (c) are the three possible models. Specimens with grain boundary orientations given in Figures 38 (a) and (b) deform very little, when tested at elevated temperatures and deformation ends in fracture after 15% deformation. Grain boundary sliding is minimal and strain incompatibility causes fracture. Specimens with Figure 38 (c) configuration, when tested at elevated temperatures, undergo grain boundary sliding at low strain-rates. A dislocation model for grain boundary sliding[36] is presented in Figure 39. In this figure a region near the grain boundaries that experiences shear stress deforms by slipping due to grain boundary sliding. At the same instance those regions near the boundary that are

Figure 38. Simulated Models Representing Three Possible Grain Boundary Orientations in Bicrystals of Beta Brass.

(a) Grain boundary is perpendicular to the tensile axis.

(b) Grain boundary is nearly parallel to the tensile axis.

(c) Grain boundary makes 45° with the tensile axis.

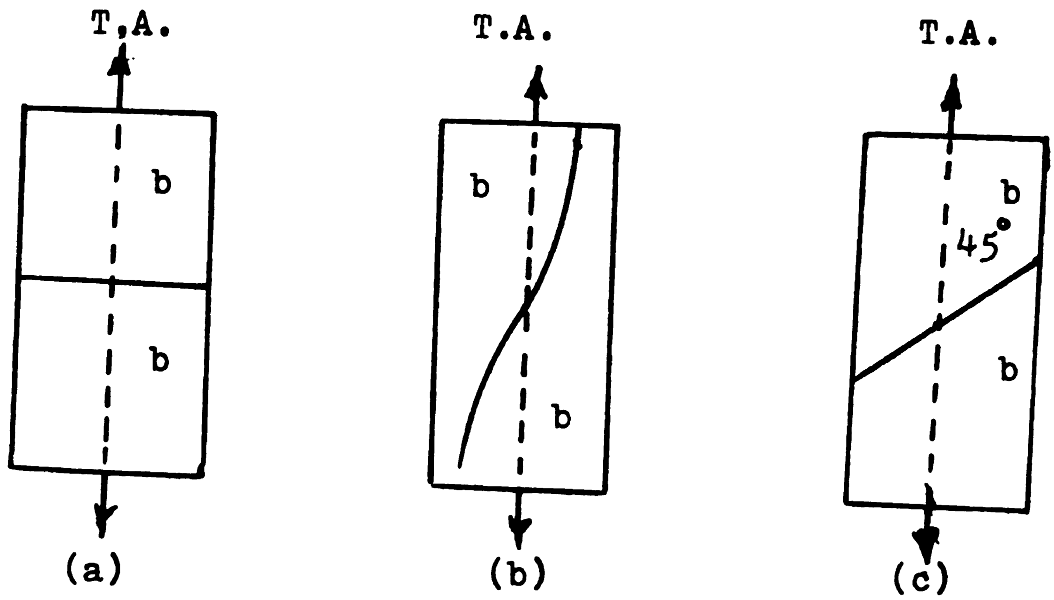


Figure 38

Figure 39. A Dislocation Model for Grain Boundary Sliding in Single Phase Alpha and Beta Brasses.

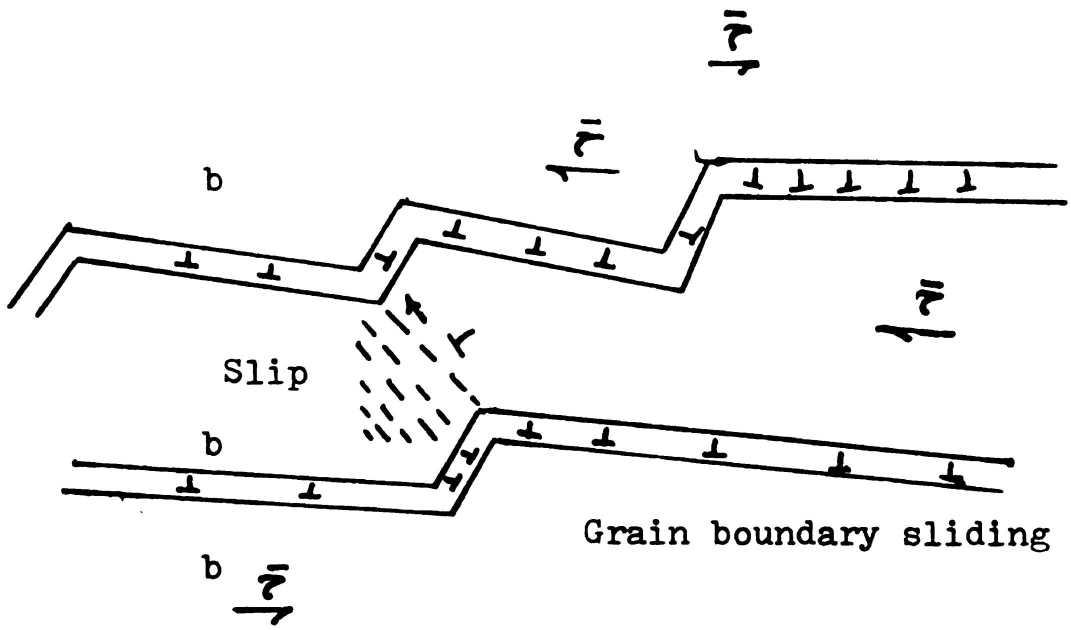


Figure 39

not able to deform by grain boundary sliding cause stress concentration and as a result slip is induced into the beta grains. The experimental observations illustrating this feature may be seen in Figure 24. At high strain-rates and below T_c, one of the grains deforms and deformation continues until needle point fracture occurs. The other grain remains intact. But, at temperatures above T_c, both grains experience some plastic deformation and one of the grains deforms more than the other. In the three mentioned cases, deformation behavior is highly strain-rate sensitive. Grain boundary sliding is the dominant factor with specimens having grain boundary orientation near 45° to the tensile axis.

A step closer to two-phase alpha-beta brass is a model system containing both the phases. Such a model was introduced by Hingwe[3,4] and Nilsen,[5] and the deformation behavior of such specimens was the subject of their investigations at room temperature. In this work, tests were performed at elevated temperatures. Some of the major differences found in the deformation behavior of bicrystals of alpha-beta brass at room temperature and at elevated temperatures are reviewed in Table XII. One can conclude that the deformation behavior of bicrystals of alpha-beta brass is temperature sensitive at a given strain-rate.

Table XII

Deformation Behavior of Alpha-Beta Brass Bicrystals Tested at Cross-Head Speed of 0.10 cm/min. at Low and High Temperatures

	Room Temperature		Elevated Temperature		
	Alpha	Beta		Alpha	Beta
Sequence in deformation of phases	Initial phase	Final phase	Below T_c:	Initial phase	Final phase
			Above T_c:	Final phase	Initial phase
Role of the phase boundary	Is an effective barrier			Did not play an important role	
Slip behavior	Single slip Multiple slip Cross slip	Single slip		Single slip Single slip	Single slip
Maximum stress (Kg/mm^2)	11.14			700°F: 4.50 800°F: 3.60 900°F: 1.72 1000°F: 0.56	
Over all deformation behavior	Non-uniform			700°F: Non uniform 800°F: Uniform 900°F: Non-uniform 1000°F: Non-uniform	
Fracture behavior	No fracture			Along grain boundaries	Needle point in an initially deformed grain

An ideal simulated model duplicating two-phase alpha-beta brass is shown in Figure 40. Let lines AB, BC, CD, DE, FF', IJ, KK' and LA represent the grain boundaries between Alpha-Alpha and Beta-Beta grains. And lines EF, FG, GH, HI, II', JJ', JK, KL and LL' represent the phase boundary between Alpha and Beta grains. Applied normal stress (σ), in the direction specified, causes grain boundary sliding. Grain boundaries such as FF', KK' and IJ undergo grain boundary sliding. This process results in formation of shear stresses across some alpha grain edges. Alpha deforms to some extent and shear stresses are reduced. Further straining results in activation of numerous local stress concentration zones at the interpenetrating phase and grain boundary corners such as corner "B" and neighboring grains undergo plastic deformation. It was observed that phase boundaries do not play an important role in alpha-beta brass bicrystals deformation at elevated temperatures. In the case when shear stresses are present, at the boundaries, it is possible for phase boundaries to accommodate some of the deformation. However, it is not a necessary condition that all grain and phase boundaries undergo some kind of deformation, because deformation proceeds wherever a favorable condition exists. Usually, with the build-up of local stress concentration regions, even the most inaccessible places in a material are deformed.

Figure 40. A Simulated Model of Two-Phase Alpha-Beta Brass Deformed Under Normal Stress.

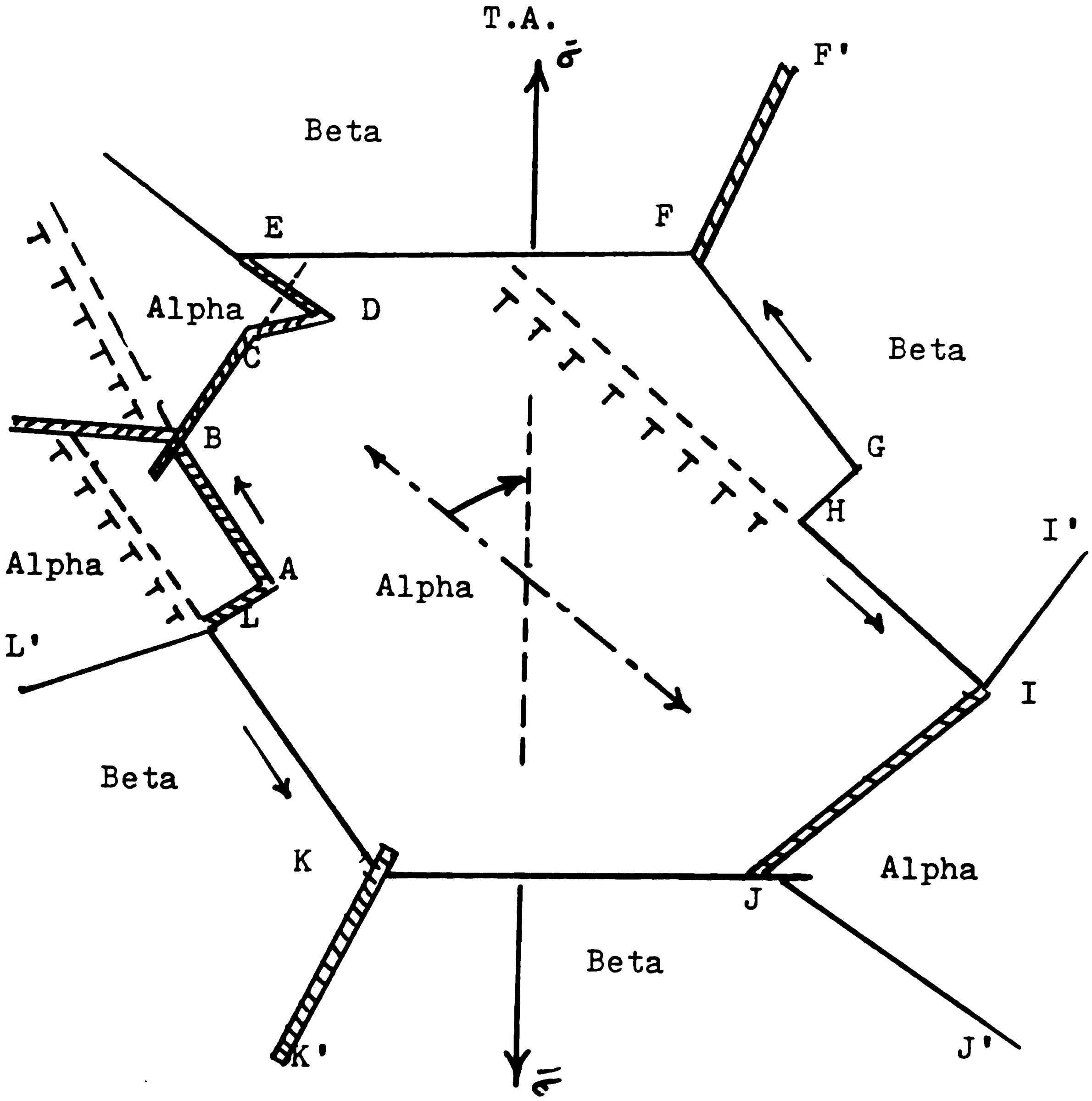

T.A.
σ
F'
Beta
E
F
Alpha
D
C
Beta
B
G
H
I'
Alpha
A
Alpha
L
L'
I
Beta
K
Alpha
J
Beta
K'
J'
σ

Figure 40

A model closer to reality for two-phase alpha-beta brass Muntz metal is shown in Figure 41. The deformation behavior of alpha-beta brass is greatly influenced by the test temperature and strain-rate. At room temperature, the deformation ends in brittle fracture before 30% deformation. At elevated temperatures below T_c, the deformation proceeds in a rather uniform manner. Above T_c where beta phase exists in disordered B.C.C. structure, the hard phase alpha, existing in platelettes shapes, reorients itself with respect to the tensile axis. This behavior is illustrated in Figure 42.

Baro,[36] made observations on samples of 60-40 alpha-beta brass deformed at elevated temperatures and reported that alpha platelettes (harder phase above T_c) were reoriented along their longer dimensions with the tensile axis. Beta phase played the role of a soft matrix, making the alignment of alpha platelettes possible. The sharp edges of the alpha platelettes were rounded off, and a coarsening of alpha platelettes resulted during the deformation.

During the course of this investigation attempts were made to understand the deformation behavior of two-phase materials. A part of this work was devoted to the development of a method by which the deformation behavior studies made on bicrystals of alpha-beta brass could be applied in hot-rolling, extrusion and shaping of materials.

Figure 41. Free Hand Sketch of Alpha-Beta Brass. Note, Alpha Platelettes are Randomly Oriented in the Beta Matrix.

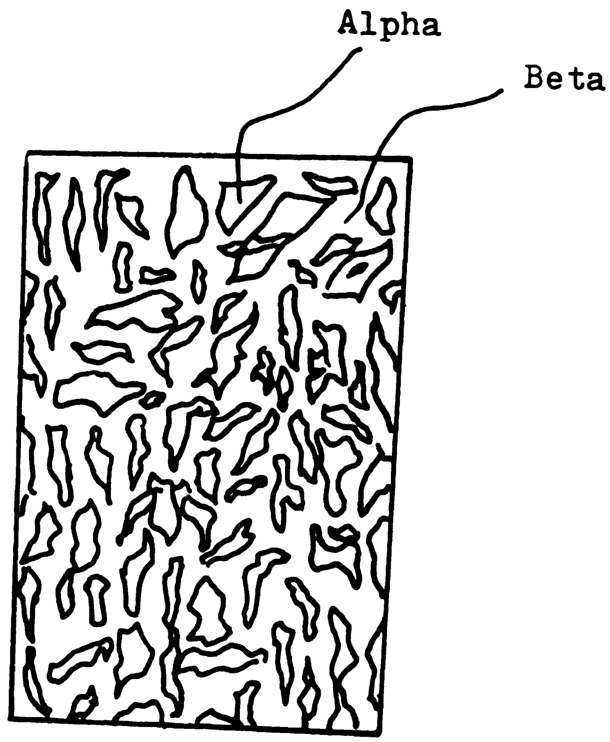


Figure 41

Figure 42. A Simulated Model of Two-Phase Alpha-Beta Muntz Metal (60-40) Brass.

A summary of the observations made on the deformation behavior of two-phase alpha-beta brass bicrystals tested at elevated temperatures is presented in Table XIV.

Table XIV

Summary of the Observations Made on the Deformation Behavior of Two-Phase Alpha-Beta Bicrystals Tested at Elevated Temperatures

	Temperature (F)	Strain-Rate (cm/cm/min)	Initial Deformation	Yield Stress of Deformed Phase (Kg/mm^2)	Critical Resolved Shear Stress (Alpha) Deformed Phase (Kg/mm^2)
I	700	0.005	Alpha	3.40	1.60
		0.025	Alpha	3.50	1.70
II	800	0.005	Alpha	2.25	1.10
		0.025	Alpha	2.50	1.20
		0.100	Alpha	3.00	1.35
III	900	0.005	Beta	1.00	--
		0.025	Beta	1.75	--
		0.100	Beta	4.00	--
IV	1000	0.050	Beta	1.35	--
		0.090	Beta	2.90	--

Table XIV (continued)

	Sequence of Deformation	Slip Behavior in Alpha if any	Slip Behavior in Beta if any	Mode of Deformation	Region of Failure
I	Grain Boundary Sliding in Beta. Alpha.	Single Slip Multiple Slip	Grain Boundary Sliding. Fine Slip.	Uniform	Beta
	Grain Boundary Sliding in Beta. Alpha.	Single Slip Multiple Slip and Cross-Slip	Grain Boundary Sliding. Single Slip.	Non-uniform	Beta
II	Grain Boundary Sliding in Beta.	Single Slip	Grain Boundary Sliding, Fine Slip, and Rumple.	Non-uniform	Beta
	Grain Boundary Sliding, Slip in Alpha and Slip in Beta Grains.	Single Slip	Rumple Appearance. Coarse Slip.	Uniform	Beta
	Minor Grain Boundary, Alpha, Beta.	Single Slip	Coarse Slip	Non-uniform	Beta
III	Beta Grains	None	Single Slip, Rumple	Non-uniform	Beta
	Beta Grains	Minor Slip	Rumple.	Non-uniform	Beta
	Beta	Beta Grains, Alpha	Single Slip.	Uniform	Beta
IV	Beta Grains	None	Extensive Deformation.	Non-uniform	Beta
	Beta Grains	None	Extensive Deformation.	Non-uniform	Beta (needle points)

CHAPTER V

SUMMARY AND CONCLUSIONS

1. Deformation studies on two-phase bicrystals of alpha-beta brass at temperatures below T_c for beta phase show that, at low strain-rates, Beta deforms by grain boundary sliding and the initial deformation occurs usually in Alpha. At high strain-rates, grain boundary sliding in Beta is minimal and most of the deformation is accommodated in alpha phase.
2. Deformation studies on two-phase bicrystals of alpha-beta brass at temperatures above T_c show that, at low strain-rates, Beta deforms by slipping in beta grains and alpha phase does not deform at all. At high strain-rates, both Beta and Alpha deform by single slip.
3. At temperatures above T_c, the individual grains in Beta with B.C.C. structure deform more easily than the grain boundaries in Beta.
4. Uniform deformation of two-phase bicrystals of alpha-beta brass with total lengths of 2.0 inches (Alpha $\simeq$ Beta $\simeq$ 1 in.) can be achieved by using low strain-rate (cross-head speed of 0.02 cm/min) at 700°F, moderate strain-rate (cross-head speed of 0.10 cm/min) at 800°F

and high strain-rate (cross-head speed of 0.50 cm/min) at 900°F.

5. The phase boundary that exists between alpha and beta phases does not play any role in the deformation of such two-phase bicrystals at elevated temperatures. However, the grain boundaries present in Beta regions deform by grain boundary sliding at temperatures below T_c, provided they are favorably oriented and the strain-rate is small or moderate.
6. The deformation behavior of beta phase is highly sensitive to strain-rate. The sensitivity is not affected by variation in temperature, although the yield stress depends strongly on the rate of deformation. On the other hand, the deformation of alpha phase is relatively insensitive to strain-rate.
7. Deformation at temperatures above T_c is dominated by creeping in Beta with an absence of work-hardening. So, alpha phase does not deform unless the initial yield stress is high enough to initiate slip in alpha phase. The plastic deformation of both the phases normally occurs in single slip mode.
8. In room temperature tests, the deformation of two-phase bicrystals was dominated by the interaction of slip in alpha phase with the phase boundary. At high temperatures, each phase deforms on its own and slip interaction with the phase boundary is not of

any significance. This phenomenon is attributed to the thermally activated dynamic recovery of the material, which will dissipate any stress concentration due to the pile-ups. Cross-slip in Alpha was not observed frequently during the course of this work, although it was a very common feature in room temperature deformation.

CHAPTER VI

SUGGESTED TOPICS FOR FURTHER INVESTIGATION

1. Deformation of specimens with phase boundaries inclined at various angles to the tensile axis can be carried out to understand the role of the resolved shear stress on the phase boundary. In this research and previous ones, the phase boundaries were perpendicular to the tensile axis. The effect of temperature on the deformation behavior of inclined boundary specimens at elevated temperatures are still desirable to be pursued.
2. The interaction of slip with the phase boundary is not fully understood at the microstructural levels. Transmission electron microscope studies of thin foils containing the phase boundary should be carried out in order to further explore the behavior of the interaction of slip dislocations with one another at the phase boundary.
3. Studies on the deformation behavior of bicrystals could be extended to multi-crystals containing coarse grains of Alpha and Beta. This will result in a better understanding of the deformation behavior of two-phase materials.

4. As an extension to this investigation, one may utilize some of the procedures used in the text related to obtaining uniform deformation in two-phase materials. A systematic method (incorporating all of the factors involved in the deformation behavior of two-phase materials at elevated temperatures) could be devised for practical applications during hot rolling, hot extruding and shaping of two-phase materials.

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APPENDIX

APPENDIX

DEFORMATION BEHAVIOR OF TWO-PHASE BICRYSTALS OF ALPHA-BETA BRASS HAVING THEIR PHASE BOUNDARIES INCLINED TO THE TENSILE AXIS

The fundamental unit for a two-phase material is a two-phase bicrystal. In the investigation carried out previously by Hingwe,[3,4] and Nilsen,[5] the orientation of the phase boundaries of the specimens tested were perpendicular to the tensile axis and the phase boundaries experienced a tensile stress. Other orientation configurations such as parallel and inclined orientations of phase boundaries with respect to the tensile axis are of great interest. The inclined boundaries will experience shear stresses during tensile deformation. The parallel boundary will require both the phases to have the same strain.

So far, there is not a technique available for preparation of two-phase bicrystal specimens with phase boundaries parallel to the tensile axis. Such studies with grain boundaries parallel to the tensile axis, were carried out elsewhere in single phase materials.[41]

An alternative method, different from that used for producing bicrystals with perpendicular boundaries, is utilized in the preparation of bicrystals of alpha-beta brass with inclined boundaries. A straightforward joining operation is carried out with a prepared alpha brass single crystal having an inclined and carefully polished joining surface, and beta brass stock having equal dimensions and matching inclined surface. Such a joining operation, with associated heat-treatments, does not provide a flat bicrystal phase boundary. The boundary tends to rotate during heat treatment, and one ends up with a phase boundary perpendicular to the length of the specimen.

In order to prepare the inclined phase boundary bicrystals, the entire heat-treatment procedure had to be eliminated. A technique was devised wherein beta brass was melted onto alpha brass single crystal substrates with inclined faces, in quartz tubes. Extremely slow cooling of molten Beta then created the large grains of Beta for the bicrystal specimens. Since, the process took a long time, Argon gas was passed through the quartz tube to avoid any oxidation problems. The probability of obtaining a flat alpha-beta interface was only about one in ten. The results presented in this appendix on the deformation behavior of inclined boundary alpha-beta brass bicrystals are from the specimens having good inclined flat boundaries. Specimens used in this work had phase boundaries inclined to the tensile axis by 60°, 45° and 30°. These specimens were deformed at cross-head speeds of 0.01, 0.1 and 1.0 cm/min. The gauge length of the entire specimen was 1½" long with $\frac{3}{4}$" of which was Alpha and the other $\frac{3}{4}$" was beta brass. Since the gauge lengths in all of the specimens were about the same, the strain-rates are presented in terms of cross-head speeds directly. All of the tensile tests were carried out at room temperature. Some of the tests were stopped for photographing the specimens. Specimens were handled with extreme care.

Specimens having 60° inclined phase boundary with respect to the tensile axis tested at a cross-head speed of 0.01 cm/min initially deformed in Alpha. Single slip in Alpha approached the phase boundary. Figure 43 illustrates deformation in Alpha and Beta is left intact after 2.7% elongation. Upon further straining, at 13.2% elongation, fine cross-slip was observed in Alpha near the phase boundary. Observations are illustrated in Figures 44 and 45. At 21.8% elongation, cross-slip lines became heavier due to the resistance of phase boundary in allowing slip in Alpha to progress through the boundary. Figures 46 and 47 illustrate the extent of the deformation of Alpha and the effectiveness of phase boundary as a barrier in resisting the progression of slip in alpha phase through the phase boundary. At the same strain level of 21.8%, beta regions away from the phase boundary were deformed. The deformation behavior is shown in Figure 48. At 48.9% elongation, a beta grain in contact with another beta grain adjacent to the phase boundary

Figure 43. Single Slip in Alpha Approaching the Boundary.
Boundary makes 60° to the tensile axis.
Cross-head speed: 0.01 cm/min
Stress: 3.358 kg/mm^2 Elongation in Alpha: 2.7%

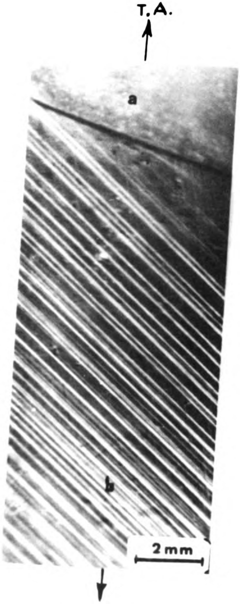

Figure 43

Figure 44. Fine Cross-Slip in Alpha Near the Boundary.
Boundary makes 60° to the tensile axis.
Cross-head speed: 0.01 cm/min
Stress: 4.328 kg/mm^2
Elongation in Alpha: 13.2%

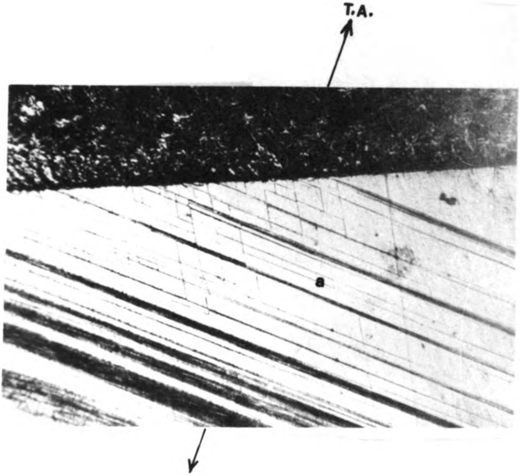

Figure 44

Figure 45. Cross-Slip in Alpha Near the Boundary Region.
Boundary makes 60° to the tensile axis.
Cross-head speed: 0.01 cm/min
Stress: 4.328 kg/mm^2
Elongation in Alpha: 13.2%

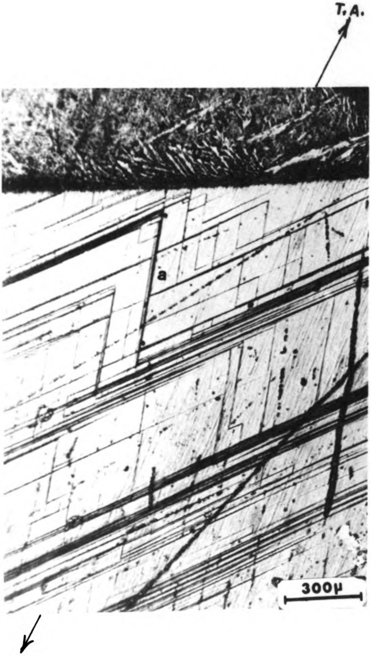

Figure 45

Figure 46. Cross-Slip in Alpha near the Boundary at a Higher Stress.

Boundary makes 60° to the tensile axis

Cross-head speed: 0.01 cm/min

Stress: 5.97 kg/mm^2

Elongation in Alpha: 21.8%

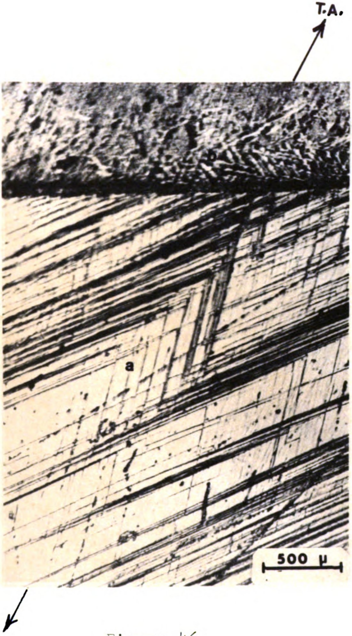

Figure 46

Figure 47. Extensive Cross-Slip in Alpha near the Boundary.
Boundary makes 60° to the tensile axis
Cross-head speed: 0.01 cm/min
Stress: 5.97 kg/mm^2
Elongation in Alpha: 21.8%

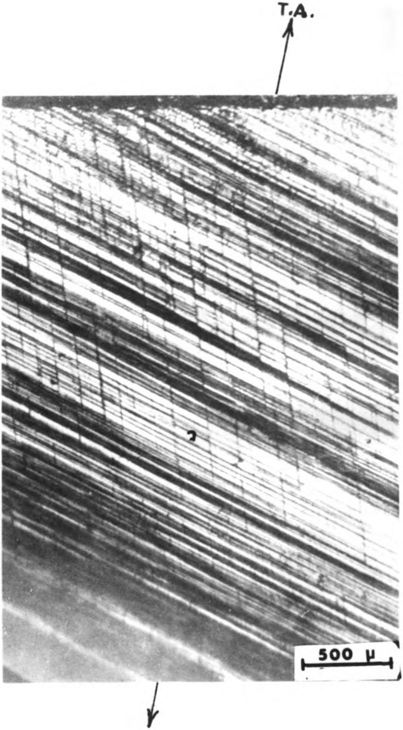

Figure 47

Figure 48. Slip in Beta in a Region away from the Boundary.
Boundary makes 60° to the tensile axis
Cross-head speed: 0.01 cm/min
Stress: 7.1 kg/mm^2
Elongation in Alpha: 21.8%

T.A.

b

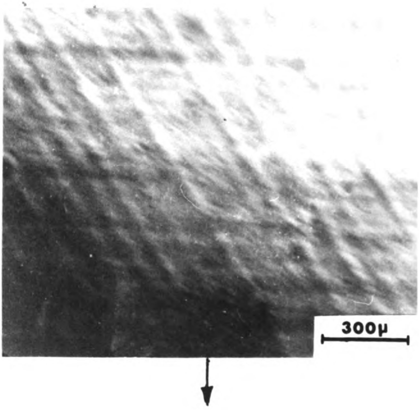

Figure 48

underwent deformation. Figure 49 is the macrograph of the deformed sample after 48.9% deformation of Alpha. Note that the beta grain adjacent to the phase boundary did not deform plastically. The beta region near the phase boundary is deformed the least. The macrograph of deformed specimen after 90.2% deformation is presented in Figure 50.

Specimens with grain boundaries inclined at 45° to the tensile axis and strained at a cross-head speed of 0.01 cm/min initially deformed in Alpha. Figures 51 and 52 illustrate the extensive deformation in Alpha and no observable deformation in Beta near the phase boundary at 37.9% elongation of alpha phase. Cross-slip near the phase boundary is observed as an indication of phase boundary playing an effective role in resisting the progression of slip in Alpha through the phase boundary. However, the beta grain in contact with the phase boundary deformed. This is illustrated in Figure 53. Heavy deformation in Alpha and slip in Beta in a region near and away from the phase boundary are shown in Figures 54 and 55. On one of the faces of the specimen, slip in Alpha interacted with the phase boundary and Beta deformed in the boundary region. This case was considered to be rare and was believed to be insignificant. In most cases, beta grains in contact with the phase boundary and away from the phase boundary deformed on their own. The

Figure 49. Macrograph Showing Slip in Beta at a Higher Stress.

Boundary makes 60° to the tensile axis

Cross-head speed: 0.01 cm/min

Stress: 12.98 kg/mm^2

Elongation in Alpha: 48.9%

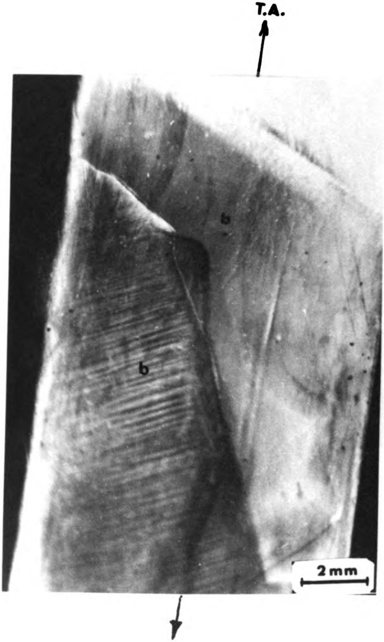

Figure 49

Figure 50. Macrograph of the Tested Specimen.
Boundary makes 60° to the tensile axis
Cross-head speed: 0.01 cm/min
Stress: 18.5 kg/mm^2
Elongation in Alpha: 90.2%

xis

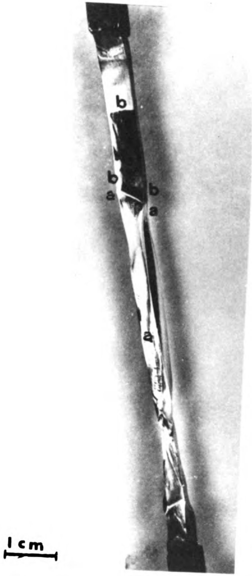

Figure 50

Figure 51. Extensive Deformation in Alpha with no Observable Deformation in the Beta Region Adjoining it.

Boundary makes 45° to the tensile axis

Cross-head speed: 0.01 cm/min

Stress: 9.836 kg/mm^2

Elongation in Alpha: 37.9%

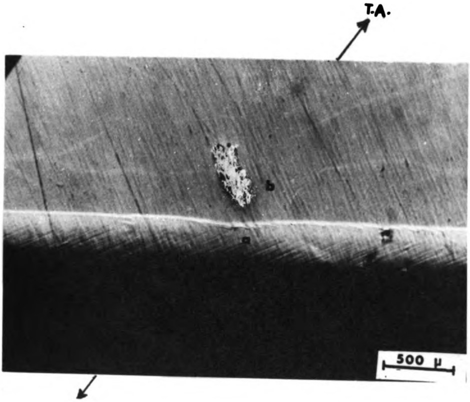

Figure 51

Figure 52. Cross-Slip in Alpha Occurring at the Interface Region.

Boundary makes 45° to the tensile axis

Cross-head speed: 0.01 cm/min

Stress: 9.836 kg/mm^2

Elongation in Alpha: 37.9%

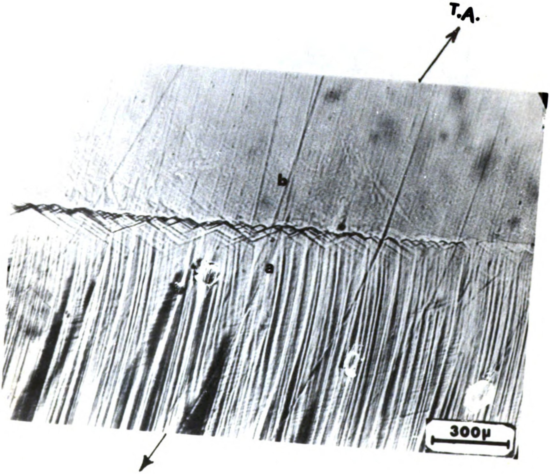

Figure 52

Figure 53. Slip in a Beta Grain that was in Contact with the Boundary.

Boundary makes 45° to the tensile axis

Cross-head speed: 0.01 cm/min

Stress: 9.836 kg/mm^2

Elongation in Alpha: 37.9%

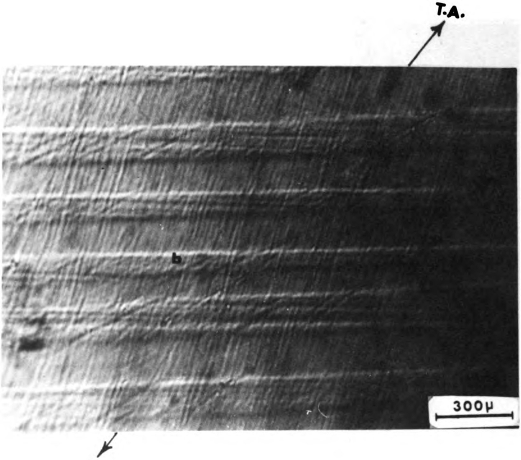

Figure 53

Figure 54. Interaction of Slip in Alpha with the Boundary. Beta has Deformed in the Boundary Region.

Boundary makes 45° to the tensile axis

Cross-head speed: 0.01 cm/min

Stress: 9.836 kg/mm^2

Elongation in Alpha: 37.9%

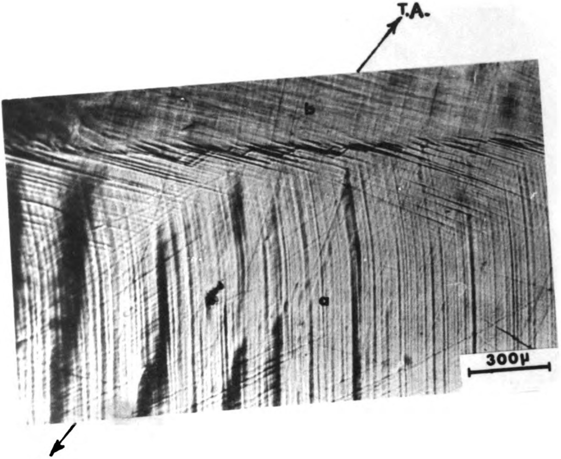

Figure 54

Figure 55. Heavy Deformation in Alpha and Slip in Beta in a Region away from the Boundary
Boundary makes 45° to the tensile axis
Cross-head speed: 0.01 cm/min
Stress: 9.836 kg/mm^2
Elongation in Alpha: 37.9%

p in
ndary
axis

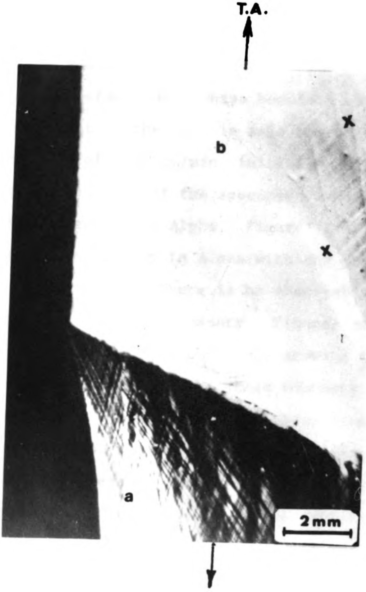

Figure 55

severity of deformation in beta grains in regions away from the phase boundary, shown in Figure 56, was much more pronounced compared with the deformed beta region near the phase boundary as shown in Figure 54, the latter is a result of the interaction of slip in Alpha with the phase boundary.

Specimens with phase boundary orientations of 30° with respect to the tensile axis and strained at cross-head speed of 0.01 cm/min, initially deformed in Alpha by single slip. One of the specimens was photographed after 24.8% elongation in Alpha. Figure 57 illustrates the interaction of slip in Alpha with the phase boundary on one of the faces. There is no observable deformation in Beta near the phase boundary. Figures 58 (a) and (b) are the micrographs of another face showing the interaction of slip in Alpha with the phase boundary. Deformation seems to have gone through the phase boundary. This observation suggests that the orientation relationship required between the Alpha and Beta for slip propagation through the boundary to occur was satisfied. That is:

$$(110)_\beta \,||\, (111)_\alpha \quad \text{and} \quad \langle 111 \rangle_\beta \,||\, \langle 110 \rangle_\alpha \ .$$

Beta regions away from the phase boundary and a grain in contact with Alpha were also deformed more severely than the region near the phase boundary on one of the faces. Figures 59 and 60 illustrate the deformation of beta regions

Figure 56. Macrographs Showing Extensive Slip in Beta Regions away from the Boundary.
Boundary makes 45° to the tensile axis
Cross-head speed: 0.01 cm/min
Stress: 9.836 kg/mm^2
Elongation in Alpha: 37.9%

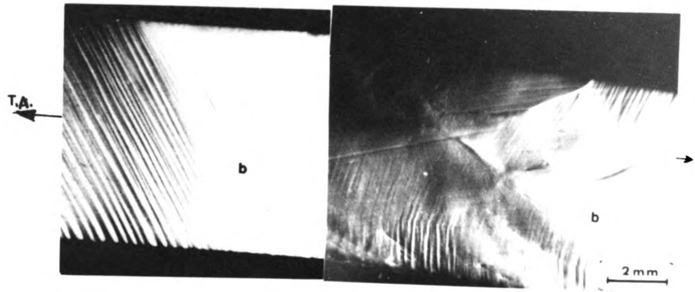

Figure 56

Figure 57. Interaction of Single Slip in Alpha with the Boundary.

Boundary makes 30° to the tensile axis

Cross-head speed: 0.01 cm/min

Stress: 8.52 kg/mm^2

Elongation in Alpha: 24.8%

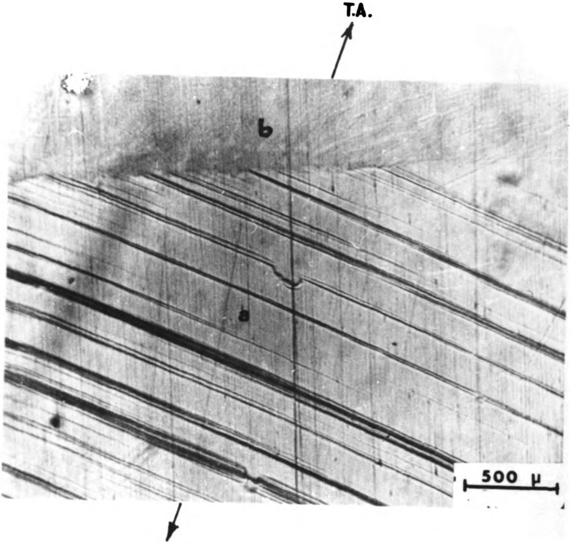

Figure 57

Figure 58. Interaction of Single Slip in Alpha with the Boundary. (Deformation seems to have progressed through the boundary).

Boundary makes 30° to the tensile axis.

Cross-head speed: 0.01 cm/min

Stress: 8.52 kg/mm^2

Elongation in Alpha: 24.8%

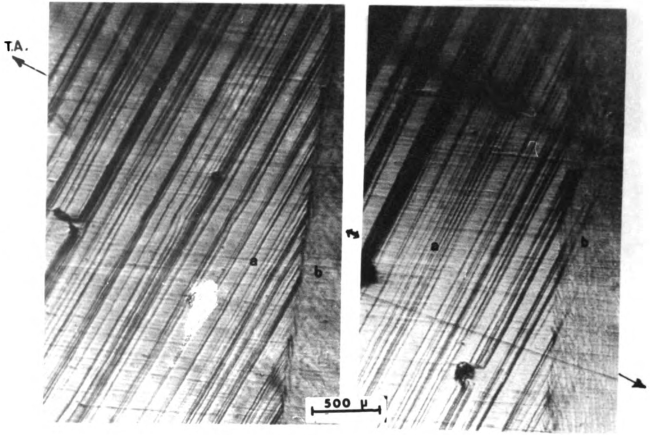

Figure 58

Figure 59. Macrographs of Slip in Beta in a Region away from the Boundary.

Boundary makes 30° to the tensile axis

Cross-head speed: 0.01 cm/min

Stress: 8.52 kg/mm^2

Elongation in Alpha: 24.8%

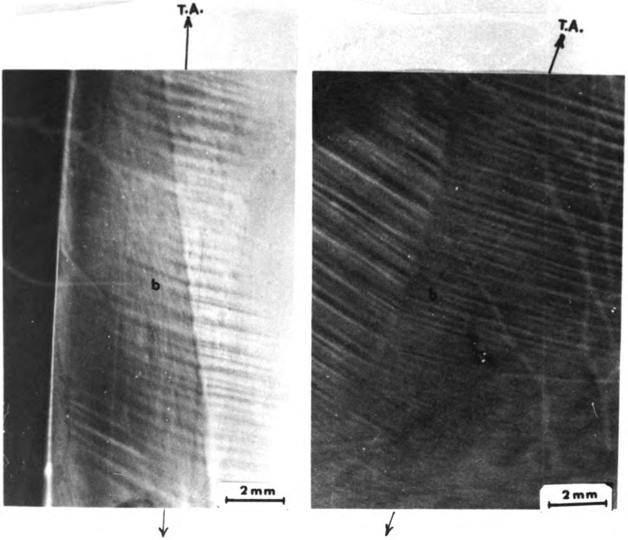

Figure 59

Figure 60. Slip in the Beta Grain that was in Contact with Alpha.

Boundary makes 30° to the tensile axis

Cross-head speed: 0.01 cm/min

Stress: 8.52 kg/mm^2

Elongation in Alpha: 24.8%

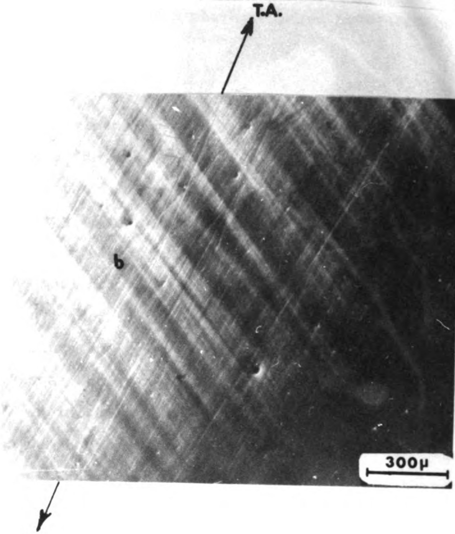

Figure 60

away from the boundary and in the beta grain that was in contact with Alpha. Table XIII contains the data on the deformation behavior of alpha-beta brass bicrystals with inclined boundaries.

Summary

In specimens having their phase boundaries perpendicular to the tensile axis, deformation of Beta is initiated by the interaction of slip in Alpha with the phase boundary. In specimens with inclined boundaries, the interaction of slip in Alpha with the phase boundaries is not the motivating force for creating large scale deformation in Beta. In such specimens slip in Beta normally occurs on its own. This result may be from the shear stresses acting on the inclined boundaries.

Even in specimens with perpendicular boundaries, there can be shear stresses at the phase boundary during deformation because of the phase boundary of the phases involved. However, the results obtained with inclined boundary specimens, in which the boundary experiences shear stresses during loading, are quite different from those observed in perpendicular boundary specimens. This observation suggests that the shear stresses caused by anisotropy are not of significance.

Table XIII

Data on the Deformation Behavior of Inclined Boundary Specimens Tested at Room Temperatures

Angle in deg.	C.H.S. cm/min	Deformation of Alpha	Deformation of Beta	Max. stress Kg/mm^2	Region of Failure
60	0.01	Extensive (3.39)*	Slip all over (7.1)*	18.507	Alpha
60	0.1	Single slip. Very little multiple slip	Crack developed	4.742	Beta near boundary
60	1.0	Single slip. Later rumpled. (3.89)*	Slip away from the boundary.	15.5	------
45	0.01	Multiple slip near boundary. (2.95)*	Slipped both near and away from the boundary. (7.9)*	9.84	------
45	1.0	Extensive (3.12)*	Slip away from the boundary.	12.79	Alpha
30	0.01	Single slip (4.0)*	Extensive slip (8.3)*	8.52	Cracked in Beta near the boundary.
30	1.0	Predominantly single slip. (3.33)*	None	11.74	------

*The stress level (Kg/mm^2) at which the observations were made.