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THE RELATIONSHIP BETWEEN MECHANICAL PROPERTIES, PREFERRED ORIENTATION AND PRESS FORMING BEHAVIOUR OF COPPER AND 70/30 BRASS.

ABSTRACT

A study has been made of the mechanical properties, press forming behaviour and preferred orientation of copper and 70/30 brass in a wide range of conditions produced by temper rolling and temper annealing.

The mechanical properties and R values at 0°, 45° and 90° to the rolling direction were determined by tensile testing, while the stretch forming, deep drawing and redrawing properties were assessed using the respective Erichsen tests. An attempt was made to relate the earing characteristics of deep drawn cups to preferred orientation, which was assessed from pole figures, and to the planar variation in R value.

The stretch formability of both materials was a maximum in the annealed condition, and there was little difference in performance between material temper rolled and temper annealed to the same hardness.

The deep drawing capacity of copper was a maximum after small rolling reductions, while that of 70/30 brass was a maximum in the annealed condition. For both materials annealing to temper increased the deep drawing capacity compared with rolling to temper; for copper the increase was greatest when temper annealing

created a recovered state. Thickness strain and hardness distributions were measured on sectioned cups to assist in the interpretation of the cup drawing results.

The redrawing capacity of copper increased with increase in hardness of the initial strip, but 70/30 brass would not redraw in any condition.

The ear height and position on cups drawn from both materials in all conditions was related to the proportionate variation in R value, and the variation in cup height appeared to be related to the planar variation in R value when considered in a radial direction.

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1. INTRODUCTION

The simplest concept of deep drawing and pressing of metals using a die and a punch has received rapid commercial exploitation since the beginning of this century, until today it is one of the largest mass production techniques. Although in the initial stages of the industry, great reliance was placed on qualitative and intuitive knowledge, the past two decades have seen the deliberate direction of scientific knowledge and research towards the augmentation of the vast practical experience within the industry.

Copper and copper-zinc alloys are two of the most commonly used metals in the deep drawing industry, and along with steel and aluminium alloys constitute a major proportion of press formed production.

This thesis contains the results of a study which was directed primarily towards establishing a more complete picture of the relationship between mechanical properties and press forming properties of copper and 70/30 brass. Much of the previous literature which is relevant to this work has been reviewed, and it is apparent that gaps exist in our present knowledge as well as there being some anomalies in the results obtained.

Several investigations have attempted to relate properties of particular material conditions with pressing performance, whilst others have concentrated on the crystallographic aspects of sheet anisotropy. The few attempts to examine press forming behaviour systematically over a range of material conditions are not unanimous in their conclusions.

This present project was designed to confirm or qualify some of the existing knowledge and to study more fully the effect of sheet temper on mechanical and press forming properties. In particular a comparison is made between strip rolled to temper and strip annealed to temper, to see if either has any economic or metallurgical advantage.

An attempt has also been made to understand the basic differences in press forming behaviour between copper and 70/30 brass and relate these to structural and mechanical properties.

It is important to distinguish between two essentially different modes of deformation involved in most press forming operations. These are stretch forming and deep drawing and both will contribute to the

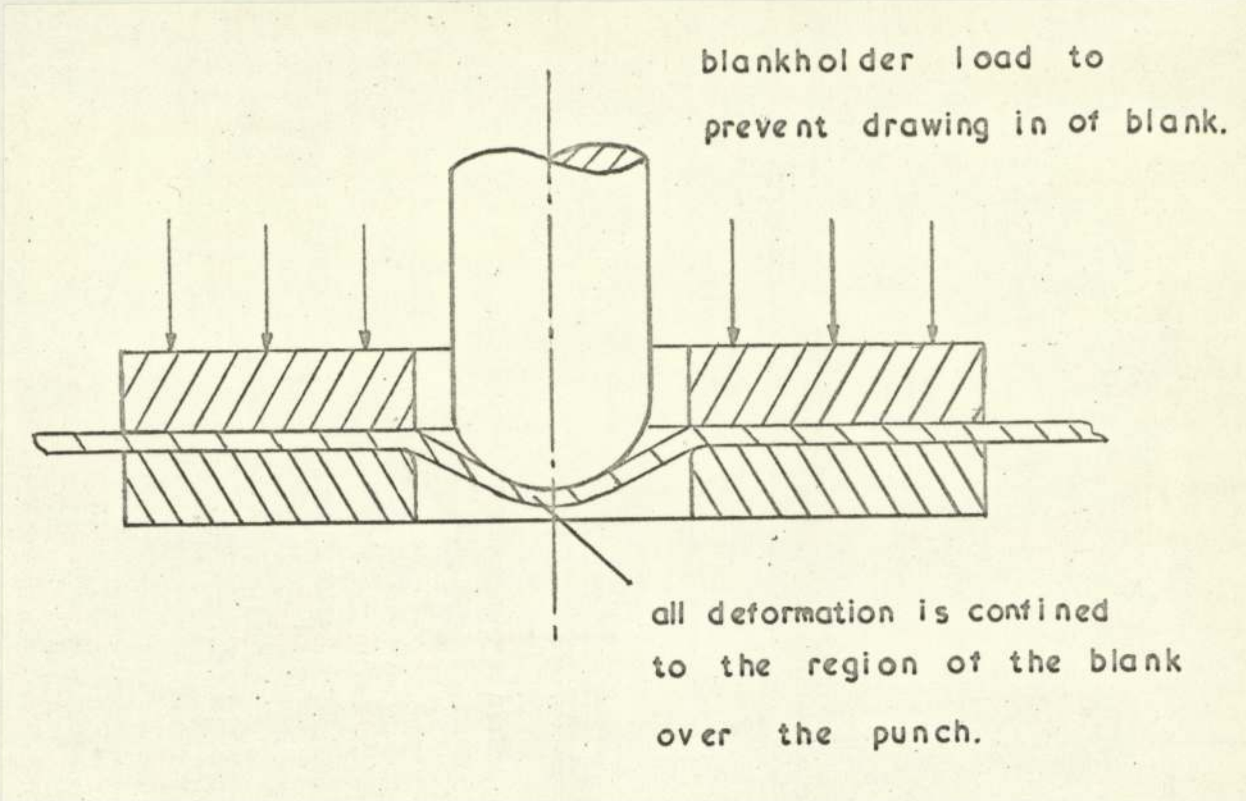


Fig. 1.1 Schematic representation of stretch-forming.

final pressing, although the relative magnitude of their contributions varies widely.

In this work pure stretch forming is defined by the stretching of a rigidly clamped flat blank over the profile of a hemispherically ended punch, Fig. 1.1. Ideally, drawing in of the blank should be absent, and there should be no friction between the punch and the sheet. Assuming an isotropic material, failure would then occur at the pole of the punch nose. In practice, frictional conditions dictate the precise position of the fracture, but generally it occurs nearer the pole as

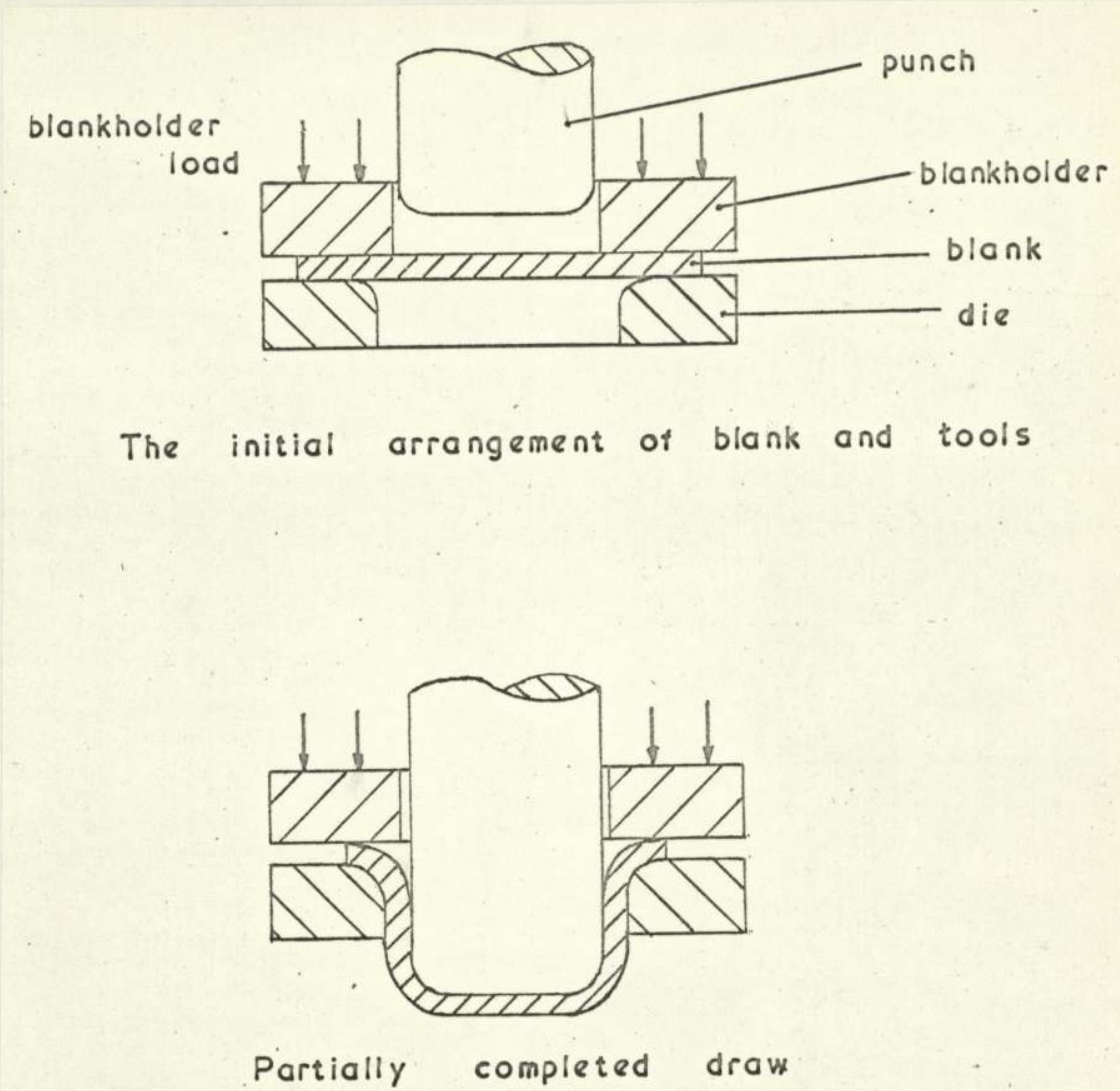


Fig. 1.2 Schematic representation of deep drawing.

the frictional forces are reduced.

Deep drawing is characterised by the forming of a flat bottomed cylinder from a circular blank, Fig. 1.2. A punch with small profile radius compared with the punch diameter is used, as distinct from the case of stretching in which the punch nose is hemispherical. Pressure is applied to the blank between the die and the blankholder to prevent wrinkling. As the punch advances, the initial deformation mode is stretching over the punch nose which, together with subsequent bending under tension over the punch profile radius, causes some thinning. Similarly, thinning occurs in the remaining area of the blank as it is progressively bent and then unbent round the die profile radius. The load on the unsupported area of the blank produces a radial tensile stress in the flange causing a "drawing in" process at the blank periphery. This reduction in diameter induces hoop compressive stresses which cause wrinkling of the flange. Therefore it is essential to provide a blankholder pressure, which allows a more uniform thickening of the blank; the maximum thickness being at the blank periphery. Ironing of the cup wall may be avoided by providing a clearance between the punch and the die which exceeds the sheet thickness by 25% to 75% depending on the material. Failure normally develops within the material stretched over the punch nose radius because of its inability to support the required drawing load. Exceptionally, compressive failure due to flange wrinkling or tensile failure in the drawing zone may occur. Thus the main stress systems likely to be present in the press forming of a simple, symmetrical article are biaxial tension, bending under tension, radial drawing and hoop compression.

Improvements in deep drawing performance have been achieved by effectively increasing the load carrying capacity of the metal at the punch nose radius by, for example, improving lubrication on the flange, or increasing the surface roughness of the punch head to reduce the stretching component. Alternatively, relative strengthening of the metal which transmits the drawing load over the die to the flange has been achieved by a differential annealing technique[1]. This produces a controlled radial variation in temper of the blank so that the rim is softer than the centre, and has been shown to considerably increase drawing capacity.

Since drawability is relatively insensitive to isotropic changes in properties because of the similar effect of such a change on the cup wall strength, on the flange properties and hence on the drawing load, it is not surprising that studies have been made of the effects of controlled anisotropic changes in properties due to preferred orientation on deep drawability. However, it is equally important to ensure that changes in the preferred orientation do not adversely affect the earing behaviour of strip.

2. THE PRESS FORMING OF COPPER AND 70/30 BRASS

2.1 Introduction

The factors which will determine the press forming behaviour of sheet metal may be divided into properties of the material being pressed and non-material properties. This section discusses the important material aspects in relation to copper and 70/30 brass, and briefly deals with some non-material aspects. A more comprehensive account of some of the more practical non-material factors is given in the book by Jevons[2].

2.2 General

After steel, copper and copper-zinc alloys probably represent the largest section of metals produced in sheet or strip for the press forming industry. Cartridge metal, or 70/30 brass is considered to be the best grade from a press formability view point, from the available range of copper-zinc alloys, and is generally used where a long sequence of press operations is required, or when the number of interstage anneals must be kept to a minimum. It has considerably better press forming properties than copper, but also requires much higher drawing loads. For general pressing operations, both copper and 70/30 brass would usually be in the soft condition. The appropriate British Standard

Specification for 70/30 brass calls for an ultimate tensile stress of 18 tons/sq.ins., and a minimum elongation on a 2" gauge length of 50%. Comparable figures for copper would be 12 tons/sq.ins. and 40% elongation respectively, but these values give little indication of probable behaviour under the press.

The mechanical properties of the strip can be varied by cold rolling, and copper is often press formed in a partially work hardened state to further strengthen the final pressing.

Probably the most important difference between copper and 70/30 brass from the press forming aspect is that in work hardening characteristics. For example, after 50% rolling reduction, the hardness of copper is about 120 D.P.N., whereas that of 70/30 brass is about 200 D.P.N. even though the hardness of annealed copper is only about 20 hardness points less than annealed 70/30 brass[3]. This data indicates the greater work hardening rate of 70/30 brass compared with copper.

Superior press formability is obtained by using copper of high purity, and in particular low oxygen content[2]. Consequently there is a greater demand for the more expensive phosphorus deoxidised coppers as opposed to the tough pitch type to withstand more severe pressing operations. Significantly, the presence of the phosphorus may, depending on the concentration, have a beneficial influence, since Roberts[4] has reported that the presence of 0.05% phosphorus in phosphorus deoxidised copper completely suppresses the cube recrystallisation texture.

2.3 The Effect of Grain Size on Pressing Properties of Copper and 70/30 brass

One material variable that can significantly influence the pressing behaviour of sheet is the grain size. Consequently, it is essential to control the grain size by the rolling and heat treatment sequence. The accepted range of average grain size in annealed 70/30 brass for general presswork is 0.035 mm. - 0.045 mm., although in annealed copper, a smaller grain size 0.025 mm., is accepted to strengthen the wall of the pressing, without significantly reducing the sheet ductility[2].

The practical consequences of press forming copper and 70/30 brass with an unsuitable grain size have been discussed by several authors[2,5,6]. It is generally considered that the larger the grain size of the sheet, the better will be the press forming properties although the surface roughness will increase[2,5]. For example, Camenisch[5] recommended a grain size within the range of 0.05 mm. to 0.1 mm. for the more severe press forming of difficult shapes. However, the precise influence of grain size on pure deep drawing or pure stretching is difficult to rationalise from reported industrial experience. Consequently, it is not clear whether the effect of grain size on stretch formability or on deep drawability is the more important for general forming operations.

Kokkonen[7] has concluded that the stretching and deep drawing capacities of 70/30 brass are increased with increasing grain size, but suggests that the corresponding increase in the degree of normal anisotropy observed may be the more important effect. However, Mear and Ford[8] compared the drawability and stretching performance of strip with a fine grain size and strip with a normal grain size for brasses of composition 85/15, 70/30, and 64/36 : Cu/Zn. They found that the drawing performance, as measured by the Swift test, was better in all cases for the fine grained material. For stretching, using an Erichsen test, they found the reverse to be true, normal grain sized alloys giving values 10 - 20% higher than fine grained. Using a round nosed punch on the Swift machine, which enhances the stretch forming component, little difference in performance was observed.

The main difficulty in determining the effect of grain size on press formability is the overriding problem of eliminating the influence of other variables, since it is almost impossible to vary the grain size without affecting other material properties whose contribution to the observed effect is unknown.

2.4 The Effect of Cold Rolling and Heat Treatment

Copper and 70/30 brass strip are produced in various tempers, usually by cold rolling, for press forming operations. However, there is only a limited amount of published data on the effect of cold rolling on formability. There is evidence that the deep drawability of several materials is increased when the sheet is initially in a partially hardened condition. For copper, Wright[9] has shown an improvement in deep drawability after small rolling reductions of the order of 3% compared with that of annealed sheet. This effect has also been observed in aluminium and some of its alloys[10], and 18/8 stainless steel[11].

Kokkonen's[7] results for 70/30 brass disagree with the observed trend. He concluded that rolling reductions up to 10% reduced the deep drawability fairly drastically, although the validity of his conclusion is questioned since only two material conditions were examined after 3% and 10% rolling reduction. It is conceivable that the drawing performance may be significantly different at a rolling reduction of the order of 6%. Alternatively, any difference of opinion as to the effect of sheet temper on drawability may be resolved in terms of the different test machines and tooling used in the investigations, since these will affect the relative contributions of stretching to the deep drawing operation. When the stretching contribution is relatively high, as when a punch with a generous profile radius is used, an annealed blank will usually give a larger limiting drawing ratio. Conversely, when the punch profile radius is small, and hence the amount of stretching reduced, sheet of harder temper may give a larger limiting drawing ratio.

Limited experimental work has shown that a difference in press forming behaviour may be expected between rolled to temper and annealed to temper sheet. Temper annealing, although more difficult to control, was shown by Wright, Long and Green[13] for aluminium/ magnesium alloys, and by King and Turner[14], for aluminium alloys in general, to give superior deep drawing properties.

For copper, Wright[9] has shown an increase in tensile elongation, stretch formability and deep drawability after temper annealing. The results of Fritz and Williams[15] for copper are consistent with this pattern. They found that the elongation of annealed to temper copper was up to 50% greater than that of temper rolled copper at a similar value of ultimate tensile strength. However, this considerable difference in ductility was not reflected in the results of Erichsen cupping tests, from which no meaningful conclusions could be drawn. A decrease in the yield stress to ultimate tensile stress ratio was observed in temper annealed material, but no tests to relate this to deep drawability were carried out.

Recently, German workers[16] have shown an increase in drawability in copper temper annealed after heavy rolling reductions, but no increase was observed in temper annealed 70/30 brass in any condition. Similarly there was no difference in the R value measured in the tensile test or in stretch formability for copper and 70/30 brass between rolled to temper and annealed to temper strip.

2.5 Non-Material Aspects

Some general engineering aspects involved in press forming have been discussed by several authors[2,17,18].

However it is worth mentioning some of the more important factors, since in some cases their effect on formability can obscure particular effects of material properties. For example, the stretch forming behaviour of sheet metal is very much dependant upon the lubrication conditions between the punch and the sheet.

Since copper is often subjected to quite heavy deep drawing reductions the selection of an efficient lubricant is very important. The efficiency of a given lubricant may depend on the speed of drawing. Thus Coupland and Wilson[19] showed that within the range of drawing speeds from 10 ft./min. to 80 ft./min., there was no detectable speed effect for annealed 70/30 brass when lubricated with graphite, but that the efficiency of liquid lubricants was reduced at the higher drawing speeds. Jevons[2] considers that copper is insensitive to drawing speed within the normal industrial range.

Finally, in relation to tool design, Grainger[18] has pointed out that the die radius and the clearance between the punch and the die can be critical to the successful drawing of a particular shape. Compared with 70/30 brass, copper is relatively insensitive to die and punch profile radii, and to die/punch clearance. In fact it is not uncommon to obtain the required pressing depth in copper by ironing a shell of near final diameter, rather than introduce successive draws of smaller reductions. The surface condition of the pressing tools is important since copper is prone to foul steel tools on heavy draws, although the view is expressed[2] that the cause is not a material one, but rather due to the imposition of excessive deformation in one operation.

It would appear, for copper and 70/30 brass, that although drawing speed is not a critical variable in itself, it can adversely affect the lubrication efficiency. Therefore in an investigation into the press forming behaviour of copper and 70/30 brass it is essential to standardise the lubrication and speed of deformation.

3. ASSESSMENT OF PRESS FORMING PROPERTIES OF SHEET METAL

3.1 Introduction

With the rapid commercial development of pressing and deep drawing during the past forty years there has been a growing need for a reliable method of assessing quantitatively the press forming properties of sheet metal without the expense of full scale trials. No single test, or group of tests, yet devised indicates unambiguously the suitability of a material for a particular pressing operation.

Several attempts to remedy this state of affairs by investigating the more fundamental aspects of the deep drawing and stretch forming processes using the plasticity theory have yielded some useful information with respect to the calculation of punch loads, required press capacity and maximum drawing reductions. Most notable in this field is the work of the late Professor Swift and his team at Sheffield[20].

Another more practical approach to the assessment of pressing properties which has been adopted to a limited extent is the use of simulative studies on a laboratory scale. A test developed along these lines could indicate rapidly and reliably, and at relatively little cost, whether or not a sheet would press to a particular shape. Complications usually arise however when precise simulation of a complex pressing is attempted, and with the exceptions of commercial pressings which approach either 'pure' deep drawing or 'pure' stretching, direct correlation is unlikely. In any case, it is virtually impossible to scale down the sheet thickness in the same ratio as the tooling.

A third approach, and one by which much recent research has been directed, is to evaluate the importance of the material properties in the press forming process. A major advance in this direction came with the realisation, through the work of Lankford, Snyder and Bauscher[21] that normal anisotropy might have a beneficial effect on the pressing properties of mild steel. Much of the subsequent work has been concentrated on mild steel, and relatively little attention has been paid to the precise effect of normal anisotropy in non-ferrous metals. For planar anisotropy, the reverse is the case.

3.2 Analytical Approach

The important stress systems in deep drawing a cylindrical, flat-bottomed cup have been summarized by Alexander[22] :-

(a) Biaxial stress system

To a first approximation, a state of plane stress is assumed in the plane of the sheet, which may be predominantly tensile as in stretching over the punch head, or a combination of mutually perpendicular tensile and compressive stresses as in the flange region. Some variation in stress across the sheet thickness will occur due to friction between the blank and the tools and restraint from the blank holder, but the effects are small.

(b) Bending under tension

This occurs over the punch nose radius, and also over the die entry radius with subsequent unbending to form the cup wall. The bending under tension results in a stress variation through the sheet thickness, and also produces some thinning of the blank.

(c) Compression normal to the sheet plane

This may be superimposed on the stress systems (a) and (b) to produce a complex state of stress as may occur for example in ironing.

The most important analysis of the deep drawing process is that due to Swift, who studied the individual effects of the basic stress systems which occur in the process. Subsequently, in association with S.Y. Chung, he supplemented his studies by making a very complete investigation of the forming of an axially symmetrical cup[20]. In particular, their analysis enables the prediction of the stress and strain in the regions of plane radial drawing. They went on to consider the effect of plastic bending and unbending under tension over the die entry radius on the drawing stress and thickness strain, and concluded that the proportion of the total drawing stress due to bending and unbending is about 15%, whilst the maximum thinning strains due to this process are 3 to 4%. They then successfully predicted thickness strains and hardness variation in the cup wall due to radial drawing and bending and unbending, but they did not predict these values for material over the punch nose. Experimentally however, Chung and Swift measured thickness strains and hardness over the punch nose, and showed that with a punch which had a sharp profile radius thinning was mainly confined to the material adjacent to the punch nose radius. With increase in punch profile radius the position of maximum thinning moved towards the pole of the punch, until when the punch nose was hemispherical, so that the contribution of stretch forming to the total forming operation was a maximum, the position of maximum thinning was at the pole of the punch nose. The maximum thinning was numerically greatest in cups formed using a punch with a hemispherical nose, and decreased as the profile radius was reduced, when at small profile radii it increased again, but was then considerably less than the value of thinning observed in cups drawn with a hemispherical ended punch. They also showed that a decrease in die profile radius at a constant punch profile radius increased the thinning over the punch nose radius, but that an increase in the hardness of the initial blank reduced the general level of thinning observed over the punch nose. More practically, the theory developed by Chung and Swift is capable of accurately predicting punch load/punch travel diagrams, from which power requirements for pressing can be calculated. One important limitation of the analysis is that it ignores the influence of anisotropy. A discussion of Swift's work is to be found in the book by Willis[23], other briefer appraisals being due to Alexander[22] and Barlow[24].

Barlow[24] attempted to estimate the maximum cup drawing ratios for some aluminium alloys using a simplified version of the Chung and Swift analysis. He calculated the maximum "nominal" stress near the base of the straight cup wall for various drawing reductions, the limiting reduction being interpolated as that at which the maximum "nominal" stress equalled the ultimate tensile stress. Barlow's results were fairly acceptable when compared with experimental values, and in general his estimates of limiting drawing ratios were rather conservative.

Warwick and Alexander[25] incorporated apparently more realistic instability criteria into Barlow's analysis in an attempt to improve the accuracy and to rank materials in order of drawability. In particular they concluded that biaxial instability gave no useful prediction of the limiting condition; neither did other more traditional criteria such as ultimate tensile strength or total punch load on the drawing machine. However, the work was complicated by the use of inhomogeneous material and their analysis did not account for the effect of thinning over the punch nose radius.

Much of the theoretical work has been concerned with conditions during radial drawing and bending under tension. From the point of view of sheet testing, a comprehensive analysis of the stress state over the punch head which takes into account frictional conditions and punch radius is required, so that tests can be more representative, rather than simulative, of industrial pressings. In addition, any useful analysis of deep drawability must consider material parameters in different directions rather than assume that they are isotropic.

3.3 Simulative Tests

Comprehensive practical reviews on the subject of simulative testing for press forming have been given by Murray[26] and Wright[6].

No well defined approach to simulative testing has been adopted, and therefore the majority of tests differ in design and specification and usually aim to simulate one or more of the important features of a pressing operation as accurately as possible. The tests include simple bending, hole expanding, conical cup drawing, and wedge drawing, but it is possible to categorise some of the tests into two groups which set out to measure parameters of either stretch formability or deep drawability. These groups include respectively the cupping and cup-drawing tests which are the most popular of the available simulative tests.

3.3.1 Stretch Forming Tests

This type of test which includes those of Erichsen, Olsen and Guillery, is often referred to as a cupping test. In principle the tests are similar and generally involve the penetration of a sheet, which is clamped between a sheetholder and a die, by a hemispherical ended punch. The assessment of performance is the depth of the impression at fracture.

There are several differences in engineering design between testing machines; for example, the mode of punch travel may be either manual by a screw mechanism, or mechanical, incorporating an electro-hydraulic system with an automatic cut out at failure. The sheet clamping design may also differ. Jevons[2] describes the essentials of several stretch forming tests in more detail.

Experience has shown that the testing conditions can markedly influence the results obtained, the amount of drawing in and frictional variations being particularly critical. The contribution from the former can be considerably reduced, but seldom eliminated, by the use of a high clamping load. Kaftanoglu and Alexander[27] have demonstrated that drawing in can only be effectively prevented by using blankholder and die with serrated faces to grip the sheet.

The specification Euronorm 14-58 of the European Iron and Steel Community for the Erichsen stretch forming test attempts to standardise testing conditions, so that whilst the above effects are not eliminated, at least there is now some basis for comparison.

The appearance of the outer surface of the stretch-formed dome provides additional information. For example there is a proportional relationship between the degree of surface roughness and the sheet metal grain size, and it is possible to accurately assess grain size by comparison with standard domes.

The mode of fracture gives some indication of the extent of anisotropy in the sheet, a circular fracture perpendicular to the direction of minimum elongation indicating a high degree of anisotropy.

3.3.2 Cup Drawing Tests

The tests most commonly encountered are the Swift and the Erichsen cup drawing tests. Both use a flat bottomed circular punch since this is generally considered to best reproduce conditions of true deep drawing. A pressure is applied to the sheet between a blankholder and a die, which permits uniform radial drawing of the sheet into the die throat. The function generally used to describe the deep drawability is the Limiting Drawing Ratio (L.D.R.), which is defined as the ratio of the largest blank diameter which can be successfully drawn into a cup, to the punch diameter.

The cup drawing tests available are basically similar but differ in tool design and specification for use. Thus the Erichsen test, because of its smaller die profile radius, is generally considered "more severe" than the Swift test. It also specifies closer control over punch/die clearance than does the Swift test. Therefore care should be exercised when comparing L.D.Rs. from different versions of the cup drawing test. It is better to quote for a particular test and make comparisons with that test using the same tooling and sheet thickness.

Hemispherical and hemi-ellipsoidal punch heads have been used in a deep drawing operation with the purpose of introducing an increased but controlled contribution from stretch forming to the total forming operation.

Pearce[28] has reported some success from this approach, but the difficulties in specifying tool design and in tool manufacture to ensure similar contributions from stretch forming and deep drawing in test and commercial pressing appear to limit the general application of this approach.

A fringe use of cup drawing, for which it is most rapid and reliable, is for assessing directionality in sheet metal, as revealed by earing characteristics on a deep drawn cup. Quantitative comparison of directionality can also be made on the basis of ear height, but it is essential to adopt a standard test procedure since the ear height is influenced by several variables. For example, ironing reduces the height of the ears. Blade and Pearson[29], and Wright[30] have been active in producing a specification for this type of test.

3.4 <u>Non-Simulative Tests</u>

The type of non-simulative test most commonly used in assessing sheet metal formability is the uniaxial tensile test. This is probably because of its ease of performance and standardisation, and its quantitative nature in terms of mechanical properties which are readily appreciated.

More recently, the plane strain type of test, either in compression or in tension, has been applied to the evaluation of drawability, since the stress systems encountered often approximate to conditions of plane strain.

Limited use has been made of the hydraulic bulge test to relate stress-strain relationships under conditions of biaxial tension to those obtained in uniaxial tension. It also allows the determination of the stress-strain curve up to higher values of strain than in the tensile test. The test is not simulative of stretch forming because of the absence of friction, so that the rating of materials using this type of test may well correlate poorly with industrial pressings.

3.4.1 <u>The Uniaxial Tensile Test</u>

It is well known that specimen geometry can markedly influence the numerical values of mechanical properties obtained in the tensile test, particularly elongation figures. It is therefore essential to standardize the specimen geometry to obtain comparable mechanical properties.

Vassell[31] considers that the ultimate tensile strength and total elongation are relatively meaningless quantities, since the former is related to the original tensile specimen area and the latter includes a contribution of unknown magnitude from local extension beyond the point of tensile instability.

The limited use of ultimate tensile strength becomes evident when comparing materials in the hard and soft conditions. In the hard condition, where the ductility is low, little reduction in area occurs and so the instantaneous or true stress is numerically similar to the ultimate tensile stress. However, the reduction in area in the case of soft material is considerably greater, and the true stress will be correspondingly higher than the ultimate tensile stress.

It is possible to derive a true stress/true strain relationship from the tensile test either by direct measurement of the cross sectional area during straining, or more usually, by assuming constancy of the gauge length zone during testing, a true stress/true strain curve can be derived from the load/extension diagram obtained during the test. Neither of the methods can be used effectively beyond instability because of the uncertainty of direct measurements within the neck in the former case, and because the assumption of the constancy of the gauge length zone is invalid beyond instability in the second method. In copper and 70/30 brass this limits the use of true stress /true strain data to natural strains of between 0.3 and 0.4.

Apart from the more general use of the tensile test for determining mechanical properties, its most useful functions are for quantitatively assessing the work hardening characteristics and the degree of normal anisotropy in sheet metals. The coefficient generally used to describe the degree of plastic anisotropy normal to the plane of the sheet is referred to as the R factor, or strain ratio, and is equal to the ratio of the strains in the width and thickness directions during a tensile test. In practice, more accurate results are obtained by measuring the strain in the gauge length and width, from which the strain ratio can be derived by assuming that the volume of the gauge length remains constant during the test.

The more commonly quoted parameters relating to strain hardening during a tensile test are the yield stress to ultimate tensile stress ratio, the uniform elongation, and the so called "n" value. The latter is normally measured as the slope of the true stress/ true strain diagram plotted on logarithmic axes and is only valid if the relationship can be approximately represented by the Ludwik equation:-

$$\sigma = K\epsilon^n$$

where σ is the true stress, ϵ is the true strain and K and n are constants depending on the material and its condition. Although there are many objections to the use of the Ludwik equation, it has been shown experimentally to be representative of the strain hardening characteristics of a wide range of materials.

It can be shown theoretically that the "n" parameter is numerically equal to the true strain at tensile instability, although it has assumed more fundamental interpretations such as work hardening rate and work hardening capacity.

A geometric method of determining the arithmetic strain at tensile instability has been suggested by Hill[32], which can be used for all materials irrespective of whether or not they obey an empirical relationship of the Ludwik type.

3.4.2 The Plane Strain Compression Test

The plane strain compression test has been used to a limited extent in the field of research into sheet metal testing. It has the advantage over the tensile test of enabling the determination of the true stress/true strain relationship up to natural strains of 2.0 or more for metal of soft and hard temper.

The widespread use of the plane strain compression test in sheet metal quality control testing is precluded mainly because it is much slower to perform than the tensile test, and generally gives no indication of material ductility. Further, although Holcomb and Backofen[33] have shown that an assessment of anisotropy can be obtained from the plane strain compression test, it is generally no more informative than the tensile test for this purpose, and in some cases is less accurate due to complications introduced by the technique.

In summary, there are numerous tests, simulative and non-simulative, which aim at producing relevant information as to the suitability of sheet metal for industrial press forming operations. With this in mind, the only measure of success one can introduce is the correlation between these tests, either individually or collectively, and industrial performance.

Unfortunately, the few systematic attempts to do this which have been reported are mainly concerned with low carbon steels. No correlation of this type has been published for copper and 70/30 brass. Most of the published information describes attempts to correlate the results of simulative tests with non-simulative tests. No marked correlation has been achieved except between L.D.R. and R value for mild steel, which was first pointed out by Lankford, Snyder and Bauscher[21] in 1950. Less accurately, they showed a correlation between the n value and stretch formability. The recent results of Atkinson and Maclean[34] for mild steel, which were obtained under very closely controlled lubrication conditions, show a correlation coefficient of 0.94 between the L.D.R. and the strain ratio. This emphasises the extreme importance of surface effects on the value of L.D.R. obtained.

4. THE INFLUENCE OF PLASTIC ANISOTROPY ON SHEET FORMABILITY

4.1 Introduction

Plastic anisotropy is a most important consequence of all sheet metal production shcedules, since it is extremely difficult, though not necessarily desirable, to produce sheet which is isotropic. One of the main causes of anisotropy in sheet metal is the presence of a preferred crystallographic orientation, which results in a directionality of mechanical properties within the sheet. Generally, the value of the mechanical properties will vary symmetrically in the plane of the sheet, one common consequence of this being the phenomenon of earing around the rim of a deep drawn product. Less obvious, is the variation in mechanical properties between the plane of the sheet and the direction normal to the sheet.

It is important to distinguish between planar and normal anisotropy, since it is now considered that they may influence formability in different ways. Indeed, since the pioneer work of Lankford, Snyder and Bauscher it has been shown conclusively by several investigators[34,35,36,37] that for mild steel, anisotropy normal to the plane of the sheet has a beneficial effect on deep drawability, whereas it is equally true that planar anisotropy has a detrimental effect in that it causes earing.

4.2 Measurement of Anisotropy

As mentioned earlier, planar anisotropy is a result of directionality of mechanical properties, and one method of assessing its value is by comparison of such properties by performing tensile tests in different directions in the plane of the sheet.

Normal anisotropy is generally measured by the changes in width and thickness during a tensile test. Hill[32] has shown theoretically, that these strains are dependent only on the sheet anisotropy and the test direction, assuming that the state of anisotropy does not change during the test. The ratio of width strain to thickness strain is commonly referred to as the R value where

$$R = \frac{\log_n \ w_o/w_x}{\log_n \ t_o/t_x},$$

where w_o and t_o are the initial width and thickness of the gauge length, and w_x and t_x the width and thickness after a certain amount of strain. By assuming constancy of the gauge length zone during the test, the

R value may be expressed in terms of the gauge length and the gauge width:-

$$R = \frac{\log_n w_o/w_x}{\log_n (l_x \cdot w_x)/(l_o \cdot w_o)} ,$$

where l is the gauge length. This enables the measurement of gauge length instead of the less accurate gauge thickness.

Hill[32] has further shown that by measuring R values in three different directions in the sheet, the total state of anisotropy can be defined indirectly, since the relative values of the three strain ratios indicate the degree of both planar and normal anisotropy. Complete isotropy requires all strain ratios to equal unity whilst other equalities of the strain ratios indicate only planar isotropy.

Holcomb and Backofen[33] considered that the ratio of plane strain strengths in the cup wall, σ_x, and the flange, σ_y, referred to as the β coefficient might be of greater physical significance to press formability than the measurement of the R value in a tensile test. They used the plane strain compression test to determine σ_x and σ_y, but concluded that for materials with limited plastic anisotropy there is little difference between β and R factors, and therefore no justification for experimental determination of β, since the measurement of R from a tensile test is quicker and just as accurate. One exception to this is the case of materials with limited tensile ductility in which it is impossible to measure the R value accurately.

A method of continuously measuring the width and thickness during a tensile test was adopted by Jegaden, Voinchet and Rocquet[38], as opposed to the intermittent measurements conventionally made. The method claims high accuracy, but it is not clear what differences are introduced, if any, by one method or the other.

4.3 Variation of R Value with Tensile Strain

There have been various experimental observations which show that for different metals in different conditions, the R factor may or may not vary with strain during the tensile test.

Lankford, Snyder and Bauscher[21] have shown that R for mild steel is invariant with strain up to maximum load, and Whiteley[35] has reported a similar finding, proposing that any detectable variation is explained by inhomogeneity in the specimen. However, Lilet and Wybo[37] observed that for mild steel the R values are scattered below about 7.5 - 10% elongation, but from 10 - 25%

elongation the R value decreases gradually. On the other hand, Wilson and Butler[36] have reported cases of R increasing with strain for mild steel, and to either increase or decrease for copper depending on the sheet texture. Jegaden,Voinchet and Rocquet[38] found a consistent linear increase in R with strain for mild steel. Randerson[39], working on aluminium, brass and copper, concluded that up to 10% elongation the strain in strip specimens is confined mainly to the length and thickness directions due to constraints imposed by the specimen shape, but that after work hardening has overcome the constraints, further deformation is accompanied by decreases in both thickness and width so that R increases. No explanation was offered for occasional high R values at low strains which decreased with further strain. It is evident that the variation in R value with tensile strain is not well understood, or even well established, so that it is advisable to determine the R value at different strains during a tensile test and then calculate an average value for the test.

4.4 The Dependance of R on Texture

Several attempts have been made to determine the precise relationship between R value and preferred orientation by predicting the width strain and thickness strain from crystallographic considerations.

By considering the distribution of ⟨111⟩ slip directions of ideal orientations on a pole figure in relation to the width and thickness directions of tensile test pieces, Burns and Heyer[40] were able to correlate fairly well experimental R values with preferred orientation for mild steel. In particular, they successfully demonstrated that a cube on corner texture favours a high R value.

More recently, Elias, Heyer and Smith[41] presented a graphical method to relate normal anisotropy with texture in low carbon steel strip. They projected the ⟨111⟩ slip directions on stereograms, with the vertical and horizontal axes corresponding to the width and thickness directions respectively, and the centre corresponding to the tension axis. They assumed that a slip direction became operative if its pole was within 10° of the 45° circle of maximum resolved shear stress. The R value is then the tangent of the angle θ subtended by the thickness direction and the line of the operative slip plane to the tension axis. This implies that for a given slip rotation, R is invariant with strain, since the slip direction moves towards the tension axis such that θ is constant. The method gave reasonable results for strong textures but the use of slip directions is an over simplification, since according to the Schmidt law both the slip plane and the slip direction should be taken into account when determining the slip system of maximum resolved shear stress.

Randerson[39] applied the method of Elias, Heyer and Smith to predict R values in copper, brass, titanium and zinc. For copper and brass it was found that two R values were often predicted, one of which agreed favourably with the experimental value, but the average of the two did not. Thus an R value of 1.0 was accurately predicted for the rolling and transverse directions of cube textured copper, but at 45° to the rolling direction the predicted R value was 0.7 where it should have been zero.

A graphical method of relating planar anisotropy with preferred orientation in aluminium sheet was proposed by Van Horn[42]. He considered the degree of coincidence between a grid of maximum resolved shear stress and the distribution of $\{111\}$ slip planes on a pole figure. A high degree of coincidence indicates a high proportion of slip planes parallel to maximum resolved shear stresses, and so the yield stress should be low. He concluded that slip directions have a negligible influence on planar anisotropy. One limitation of the Van Horn criterion pointed out by Roberts[43] is the high degree of coincidence between the grid and the perfect cube orientation for a tension axis at 0^{o}, 90^{o} and 45^{o} to the rolling direction, which implies planar isotropy.

Roberts improved Van Horn's analysis by separately considering the slip direction on a $\{220\}$ pole figure and showed that in cube textured copper, an R value of 1.0 is predicted in the rolling and transverse directions, and of zero at 45^{o} to the rolling direction.

It is generally assumed that the slip system with the maximum resolved shear stress will be the first to operate. However, the results of Wu and Smoluchowski[44] disagree with this clasical approach . They have shown that the operative slip system in flat single crystal tensile specimens is not always the one of maximum resolved shear stress. They suggest that because of restraints due to specimen geometry, systems of lower resolved shear stress may operate if the resultant slip rotation tends to reduce the specimen in thickness rather than in width. If this argument also applies to polycrystalline metals, it should only affect the R values if there is a marked degree of anisotropy. More usually however, sheet textures are diffuse so that there is a spread around several ideal orientations. Whiteley and Wise[45] have demonstrated one effect of the relative influence of different ideal orientations when they concluded that in low carbon steel the absence of the weak cube orientation correlates more directly with the R value than does the presence of the strong cube-on-corner orientation.

4.5 The Relationship between Normal Anisotropy and Press Formability

Subsequent to the early work of Lankford and his associates[21] few investigations into the effect of normal anisotropy on press forming behaviour had been carried out until Whiteley[35] reported the results of an investigation into a range of low carbon steels. He concluded that normal plastic anisotropy is the main factor influencing deep drawability when stretch forming is restricted. Subsequent results tend to support this conclusion, although Wilson and Butler[36] point out that the scatter in their own results and in those of Whiteley is too great to allow successful prediction of deep drawability on this basis alone. They suggest that more precision would be achieved by the use of more closely standardized cup drawing tests. Due to minimising frictional effects, the work of Atkinson and Maclean[34] approximates to these ideal conditions, but this was for mild steel only.

Few results of L.D.R. and R value determinations have been presented for copper and 70/30 brass, but some results which have been reported are shown graphically in Fig. 4.1. There appears to be little correlation between L.D.R. and R value, but since different workers used different tooling dimensions and lubrication it is not possible to arrive at a valid conclusion at this stage.

The theoretical reasons for the influence of normal anisotropy on deep drawability have been considered by Whiteley[46], and Backofen, Hosford and Burke[47]. Qualitatively their argument assumes that plane strain conditions exist in the cup wall during deep drawing, since the strain in the circumferential direction is zero under the action of the biaxial tensile stress. Since failure usually occurs due to straining in the thickness direction, that is thinning, an increase in resistance to thinning as indicated by

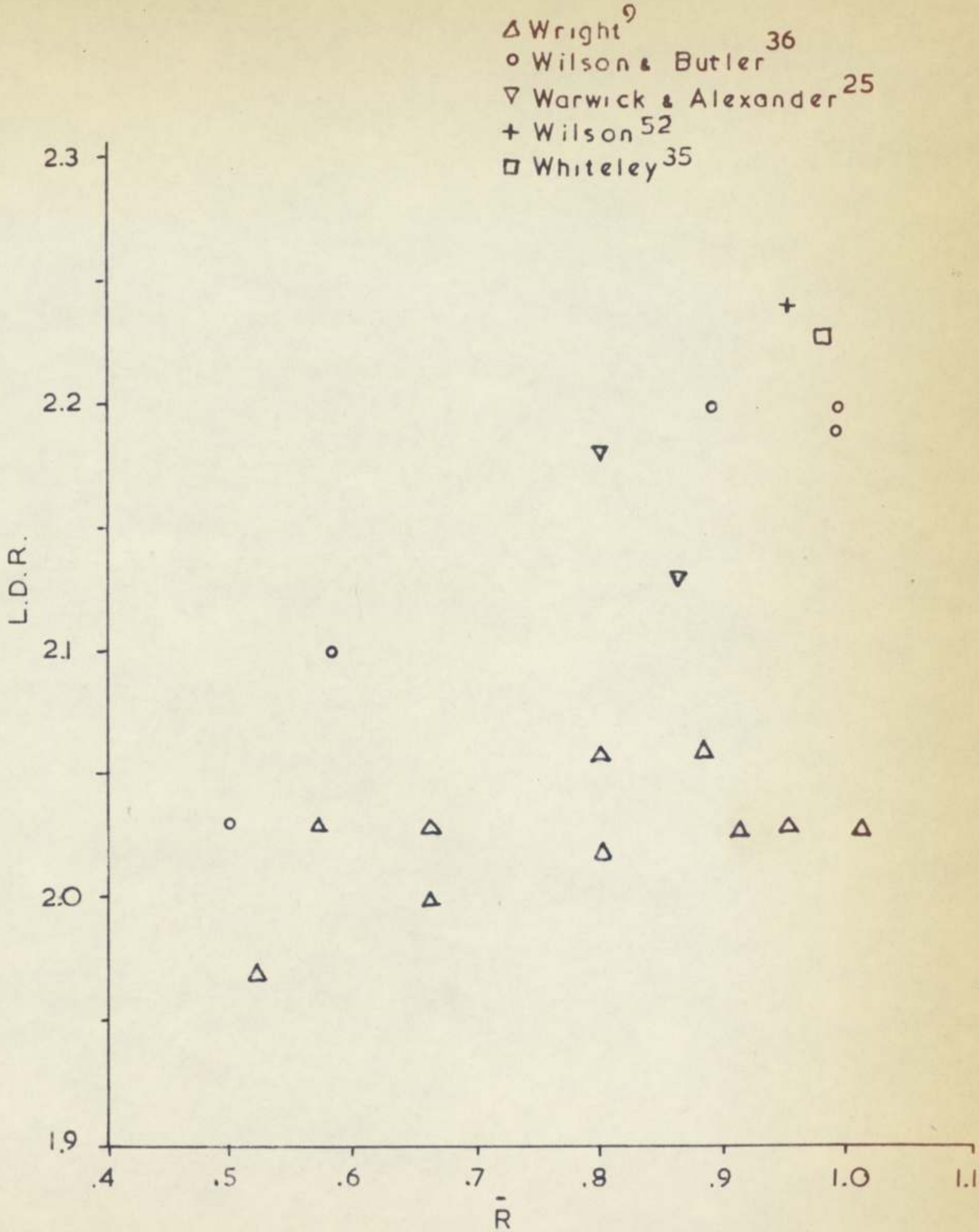


Fig. 4.1 L.D.R. against $\bar{R}$. for copper and 70/30 brass.

a high R value increases the yield strength of the cup wall. Since a corresponding increase in the drawing load does not necessarily occur because of the different stress system in the flange, a net increase in drawing capacity may result. Moreover, Backofen, Hosford and Burke[47] have shown theoretically that a small reduction in yield strength of the flange can occur due to an increase in R value.

Certain qualifications to any fundamental relationship between R and through thickness strength have been discussed by Dillamore[48]. A high R value indicates the operation of slip systems which introduce only a small thickness change in uniaxial tension, but it does not necessarily follow that the same slip systems will operate under conditions of biaxial tension. The main evidence for the relationship between R value and through thickness strength is still only experimental.

The effect of normal plastic anisotropy on stretch forming performance is less well understood. Keeler and Backofen[49], and in more detail Moore and Wallace[50], have predicted from theoretical considerations that a high strain ratio tends to produce more uniform straining in biaxial stretching operations, and therefore should increase stretch forming capacity.

However, Moore and Wallace point out that their analysis does not take into account variations in the work hardening characteristics in the plane of the sheet. In addition it was assumed that instability occurs in one of the two principle stress directions. Whilst in practice this is often the case, it is by no means universally true.

Little experimental verification that a high R value improves stretch formability has been reported. Whiteley and Harper[51] have shown that in a stretching operation on aluminium stabilized steel, where the flange was only partially restrained, the punch penetration increased as R was increased within the range 1.21 to 1.68.

More recently, Wilson[52] has demonstrated an adverse influence of R on stretch formability, since a highly textured titanium alloy with a strain ratio of 5.0 had a lower cup height than mild steel or copper, although the tensile elongation of all three materials was similar. This is also in agreement with the general conclusions of Heyer and Newby[53].

Pearce[54] studied the hydraulic bulging behaviour of several sheet metals with average strain ratios varying from 0.2 to 10.0. He found that the thickness strain at the point of fracture increased with decreasing R value, while the radial strain distribution became more even with increasing R value. Pearce also cold rolled aluminium strips up to 16% reduction to give a range of uniform elongations at a constant R value. He was thus able to demonstrate a linear decrease in bulge height and a linear increase in polar thickness strain with decreasing uniform elongation at a constant R value. At rolling reductions greater than 16%, the R value decreased sharply and a corresponding decrease in the bulge height and an increase in the polar thickness strain was observed. He thus associated an increase in the thickness strain at fracture with a decrease in R value. The main points to emerge from these results are that with aluminium of limited ductility, the bulge height appears to increase with R value, but that in strip of greater ductility the bulge height is not so dependent on R value. However since stretch formability was measured on a hydraulic bulge tester in which friction is absent, it is questionable how well the results of Pearce would correlate with industrial stretch forming.

5. PREFERRED ORIENTATION IN COPPER AND 70/30 BRASS

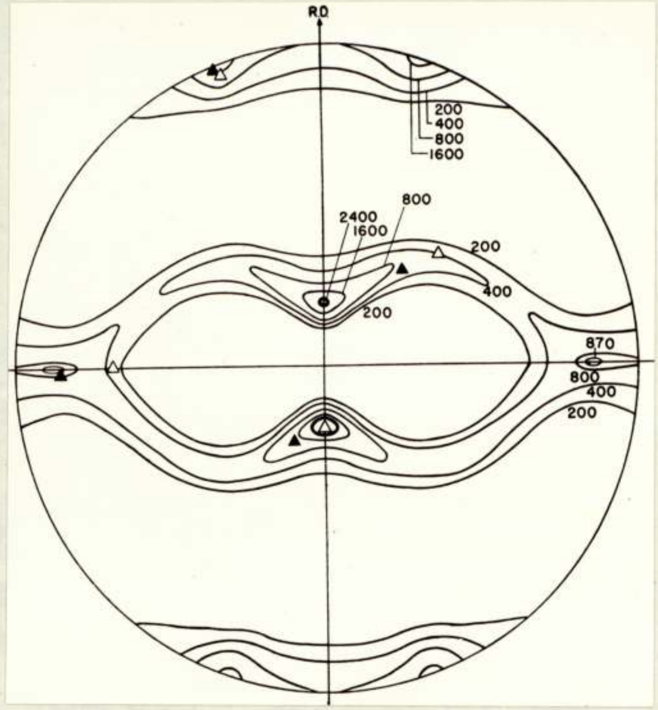
5.1 Introduction

Since the effects of anisotropy in sheet metal formability are largely dependent on textural characteristics, it is relevant to briefly discuss the current knowledge and theories of preferred orientation with particular reference to the types of rolling and recrystallisation textures commonly observed in face-centred-cubic materials such as copper and 70/30 brass. A comprehensive review on the subject of preferred orientation in metals has been given by Dillamore and Roberts[55].

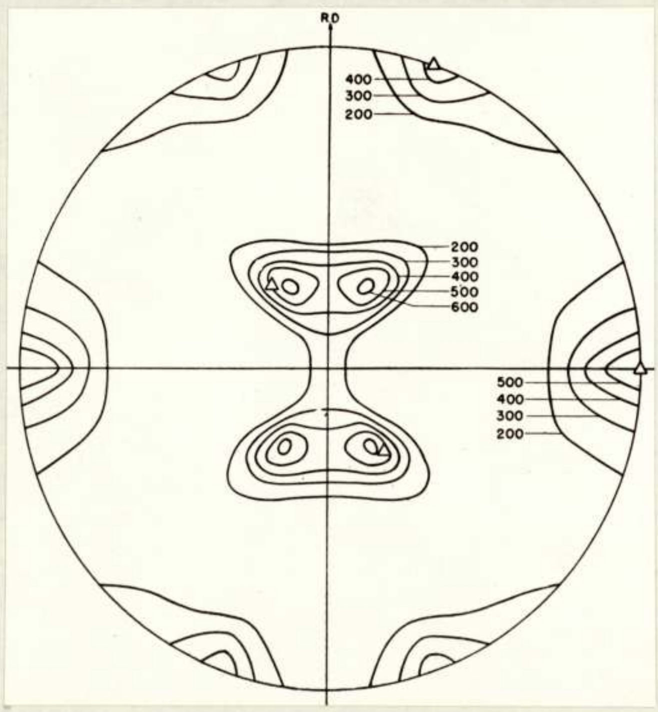
5.2 Rolling Textures in F.C.C. Metals

It is now fairly well established that one of two basic types of texture will be produced in most F.C.C. metals and alloys after heavy rolling reductions (95%) at room temperature. These two textures, commonly referred to as the "pure metal" and "alloy" textures are generally described as those of copper and 70/30 brass respectively and are shown in Fig. 5.1.

Smallman and Green[57] have shown that a transition from "pure metal" to "alloy" type texture can be produced in any F.C.C. metal or alloy at temperatures less than about 0.25 T_m as a result of a decrease in



(a)



(b)

Fig. 5.1 {111} pole figures for the "inside texture" of (a) copper and (b) 70/30 brass after 95% rolling reduction. (Hu, Sperry and Beck[56]) stacking fault energy to a value less than 35 ergs/sq.cm. This stacking fault energy dependence of texture has been associated with the ability of the metal to cross slip.

Dillamore and Roberts[58] have proposed the most comprehensive theory of rolling textures to account fully for the observed orientations. Three basic assumptions were made:-

(a) The stress system in rolling is biaxial.

(b) Deformation in F.C.C. metals occurs by slip on octahedral planes. Cross slip is considered to be of importance to the theory, its extent depending on the material, being a function of the stacking fault energy, on the level of internal stress governed by the amount of rolling reduction, and on the temperature, since cross slip is a thermally activated process. Transmission electron microscopy of F.C.C. metals and alloys supports the conclusion of a dependancy of cross slip on stacking fault energy.

(c) Textures develop as a consequence of deformation on two main slip systems;

accommodating multiple slip in the grain boundary region is postulated to account for the continuity across these regions, but is considered to be of little importance to texture development.

The theory suggests that in all F.C.C. metals the texture varies with increasing amounts of rolling reduction according to:-

"alloy" type → "pure metal" type → $\{112\}\langle 111\rangle$

Metals whose stacking fault energy is low, rolled at room temperature, in which extensive cross slip does not occur, such as 70/30 brass, develop the alloy type of texture described as $\{110\}\langle 112\rangle$, whereas metals whose stacking fault energy is relatively high, and therefore cross slip extensively, such as copper, develop rotations towards the irrational "pure metal" texture and towards the end orientation $\{112\}\langle 111\rangle$. Dillamore and Roberts have shown that aluminium which has a very high stacking fault energy, has a texture described by $\{112\}\langle 111\rangle$ after heavy rolling reductions. For metals with intermediate values of stacking fault energy, the rolling textures lie within the spread from $\{110\}\langle 112\rangle$ to $\{112\}\langle 111\rangle$.

5.3 Recrystallisation Textures

Recrystallisation textures obtained after annealing are generally related to rolling textures prior to annealing by a rotation of 40° about a common $\langle 111 \rangle$ pole[59]. Two main theories have been proposed to account for the observed recrystallisation textures.

The oriented growth theory[60] suggests a random nucleation of the new texture, but that there is preferential growth of nuclei near those orientations having a $\langle 111 \rangle$ pole in common with the matrix, but rotated 40° about it. On the basis of this theory, Beck[60] has explained the origin of cube texture in copper, since a nucleus in this orientation is favoured for growth into all four components of the rolling texture.

The theory of oriented nucleation supposes that the new grains grow from small regions of certain orientations which are already present in the cold worked matrix. The orientation of the recrystallised grains is therefore determined by that of the original nuclei. This theory can obviously account for the retained rolling texture after annealing, but it does not satisfactorily explain textures which are different from the rolling texture.

Neither theory can completely account for all the observed textural features, but Dillamore[59] considers that there is now a weight of evidence in favour of the oriented growth mechanism of recrystallisation, and he has proposed a set of principles which can be applied to accurately predict the recrystallisation textures resulting from known rolling textures, for materials of different stacking fault energies.

The types of annealing textures most commonly reported in copper are the retained rolling texture and the cube texture $\{100\}\langle 001\rangle$. The directionality of mechanical properties produced by the presence of one or the other of these textures can be controlled commercially by obtaining a balanced combination of the two. Directionality is a minimum when the amount of cube texture present is 10 - 20%[4].

The recrystallisation textures in 70/30 brass differ from those in copper, which is not surprising in view of the different rolling textures. Following a heavy cold rolling reduction, 70/30 brass has a low temperature (400°C) annealing texture approximately described by $\{113\}\langle 211\rangle$, and a high temperature (750°C) annealing texture approximating to $\{110\}\langle 112\rangle$.

Small amounts of certain solutes in copper have been shown to have a major effect upon the rolling and annealing textures[57]. For example, phosphorus, arsenic and antimony are particularly effective in transforming the deformation texture from "pure metal" to "alloy" type, while of the order of 10 - 15% zinc is required to produce this transition. More important, even smaller concentrations of these solutes are required to influence the annealing texture. To produce a detectable change in rolling texture at least 0.25% phosphorus is required, but the cube recrystallisation texture may be completely inhibited by the presence of 0.05% phosphorus[4].

6. EXPERIMENTAL PROCEDURE

6.1 Materials

The copper and 70/30 brass used in this investigation were obtained from one of the normal commercial sources. The main impurities are shown in Table 6.1. The copper content of the 70/30 brass was 69.7% and the zinc content is obtained by difference. The analysis showed no trace of phosphorus in the copper.

Main Impurities	Pb	Bi	Ni	Fe	As	Sn
Copper	<.001	<.001	<.001	-	<.005	<.002
70/30 brass	.01	<.002	.01	.02	<.01	.01

Table 6.1

6.2 Processing

The flow diagram, Fig. 6.1, shows the processing sequences used to obtain copper in a range of temper rolled and temper annealed conditions. The processing of the 70/30 brass was identical to that of copper, except that the initial slab thickness was 0.4 in. and the final temper annealing treatments were different. It should be noted that the textures in the starting materials represented in Fig. 6.1 will be different because of the different initial rolling reductions used.

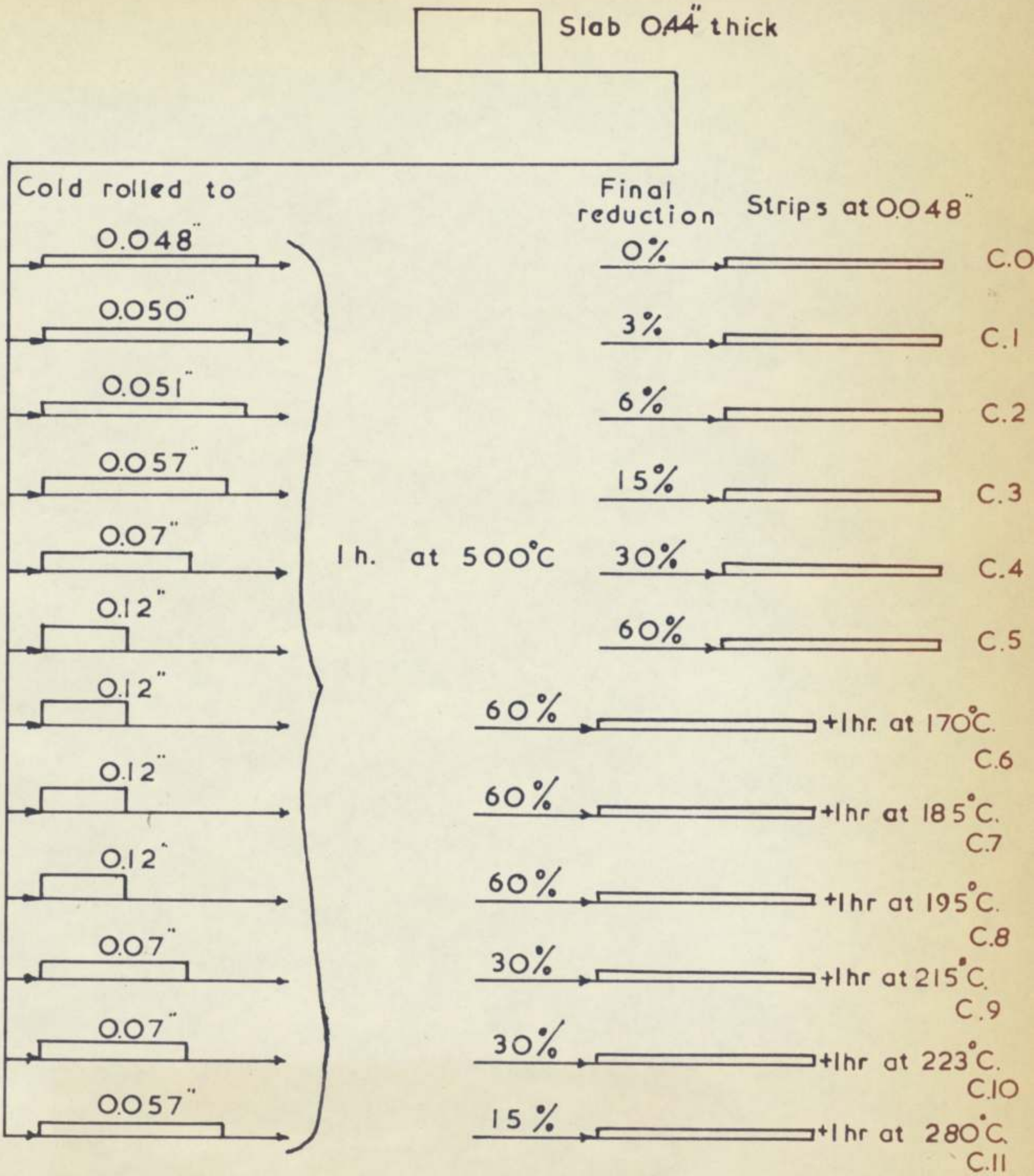


Fig. 6.1 Processing of copper

The conditions and identities of the 70/30 brass strip were as follows:-

B.0	Annealed		
B.1	Cold rolled	3%	
B.2	" "	6%	
B.3	" "	15%	
B.4	" "	30%	
B.5	" "	60%	
B.6	" "	60%	+ 1hr at 285°C
B.7	" "	60%	+ 1hr at 295°C
B.8	" "	60%	+ 1hr at 320°C
B.9	" "	30%	+ 1hr at 315°C
B.10	" "	30%	+ 1hr at 325°C
B.11	" "	15%	+ 1hr at 370°C

Since the grain size is known to affect the mechanical and press forming properties of most materials, it was decided to control its influence by ensuring that all strips had a reasonably constant grain size after their final cold rolling reduction. To achieve this, it was necessary to control the grain size of strips after their intermediate anneal, by investigating the influence of annealing temperature on the grain size of material after its initial rolling reduction. A simultaneous investigation of the hardness variation with intermediate annealing temperature was carried out, so that the intermediate annealing treatment selected would both ensure that the material was in the fully softened condition and possessed approximately the required grain size.

It was found for both copper and 70/30 brass that the grain size for a particular annealing treatment was almost independent of the initial rolling reductions. In addition, for copper, it was found that the grain size was rather insensitive to annealing conditions, so that there was a fairly wide range of annealing temperatures for which the grain size remained approximately constant irrespective of the initial rolling reduction. This can be seen in Fig. 6.2 where the grain size bars indicated refer to the variation in grain size with initial cold rolling reduction for an annealing time of 1 hour. Accordingly, it was decided to use an intermediate annealing temperature of 500°C for the copper, which is consistent with industrial practice, and which produced a grain size within the range 0.025 - 0.035 mm. which again is comparable with that used in practice.

For 70/30 brass, it was found that the grain size increased considerably with annealing temperature between 400°C and 700°C. Therefore, the intermediate annealing treatment for 70/30 brass, 500°C for 1 hour, was chosen to produce a grain size similar to that of the copper.

All strip finally tested was of nominal thickness 0.048 in. Rolling reductions were carried out on a Robertson two-high mill, the exit speed of the strip

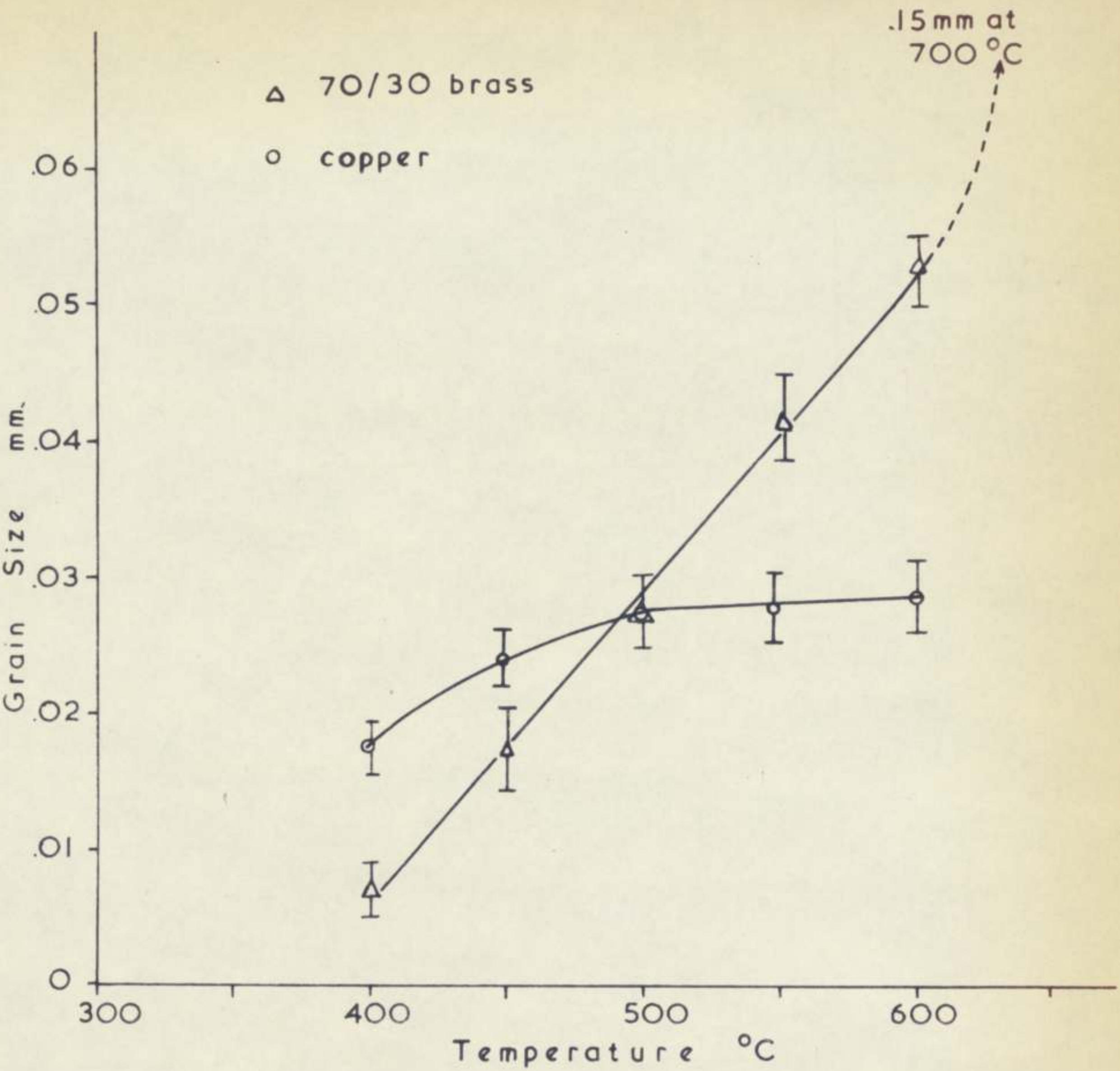


Fig. 6.2 Variation of grain size with intermediate annealing temperature for a time of 1 hour.

being about 25 ft./min. The rolls were lubricated with a light oil. Variation in final gauge was less than ± .001 in. across the strip width and length, and the actual gauge was within .0015 in. of nominal in every case.

All the heat treatments were carried out in an electrically heated air-recirculating furnace. The temperature could be controlled to within ± 5°C at 175°C and ± 2°C at 400°C, and point variation in temperature was very small.

The particular temper annealing treatments used were selected from sets of isochronal curves of hardness against temperature plotted for the larger cold rolling reductions. The isochronal annealing curves for copper and 70/30 brass are shown in Figs. 6.3 and 6.4 respectively. An attempt was made to compare temper rolled and temper annealed sheets on an equivalent hardness basis; thus sheet cold rolled 60% was temper annealed to give hardnesses equal to those of sheets cold rolled by 6%, 15% and 30%.

The increase in hardness of cold rolled 70/30 brass when annealed at low temperatures, 100°C - 250°C, has been observed previously[61], and has been explained as a

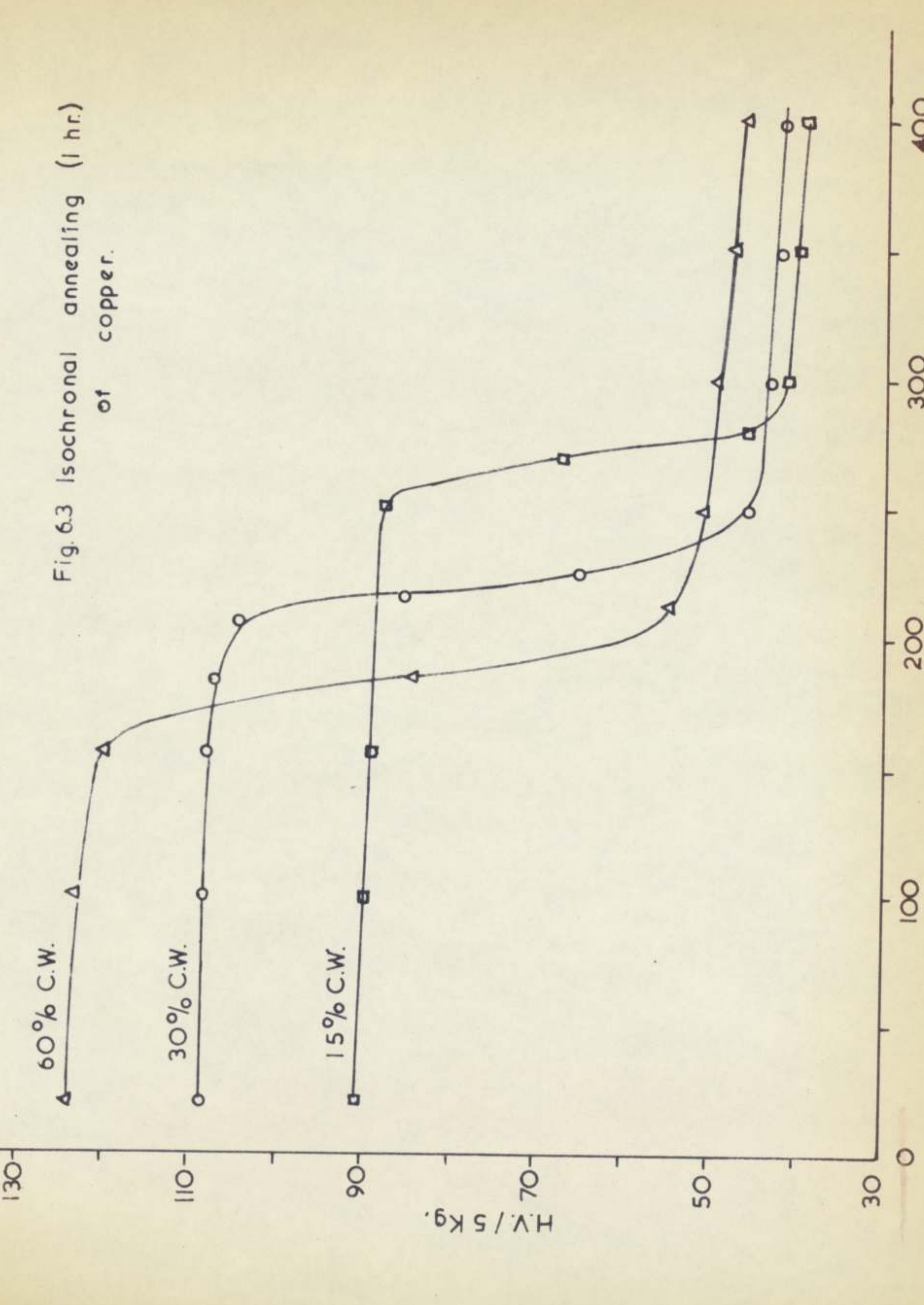


Fig. 6.3 Isochronal annealing (1 hr.) of copper.

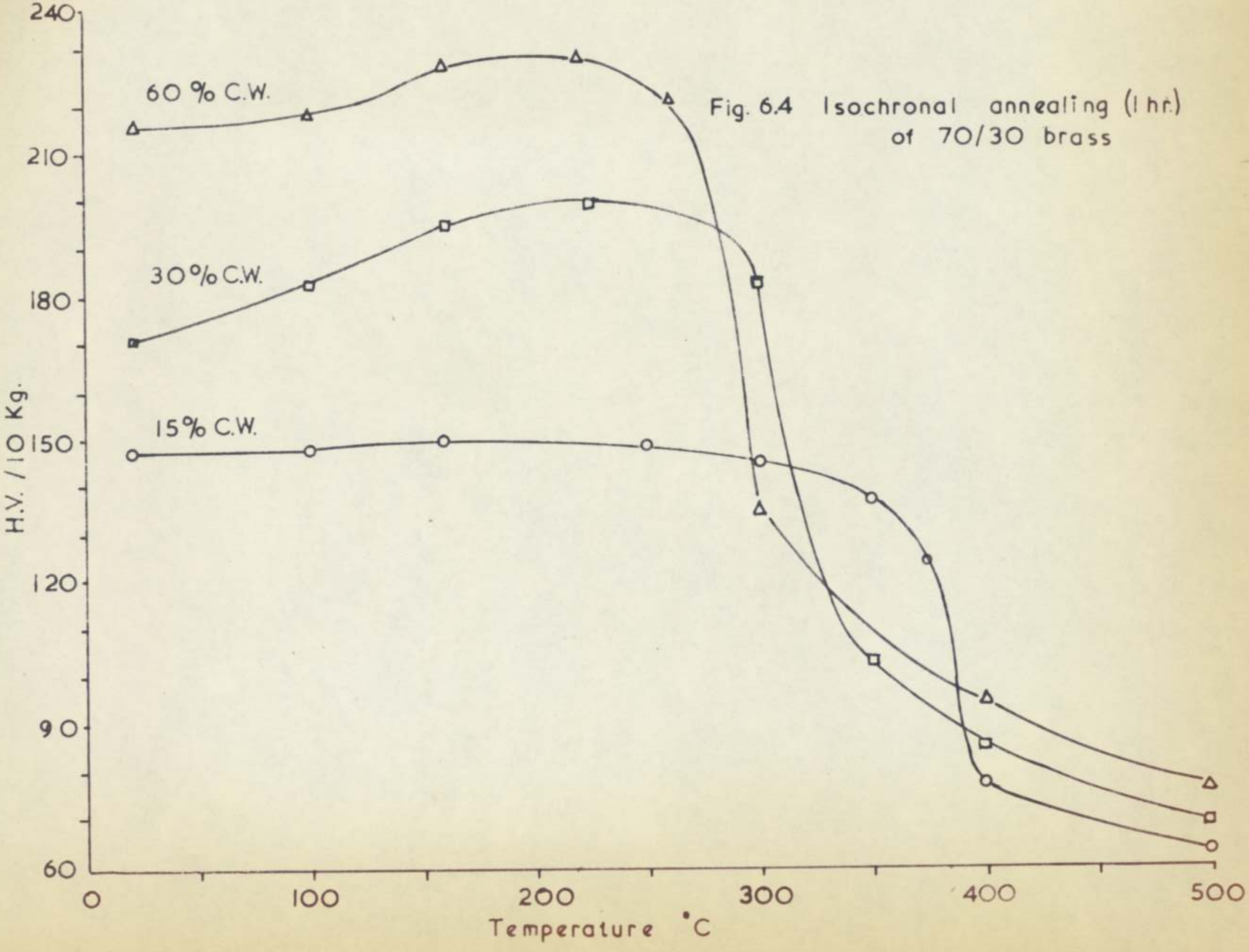


Fig. 6.4 Isochronal annealing (1 hr.) of 70/30 brass

strain-ageing effect.

6.3 Tensile Testing

Duplicate strips 6 in. long and 1 in. wide were sheared at 0°, 45°, and 90° to the rolling direction for each material condition to be investigated. Tensile test pieces were machined to the dimensions of the British Standard specimen shown in Fig. 6.5, extremely small cuts on a high speed milling machine being used to minimise the introduction of effects due to work hardening.

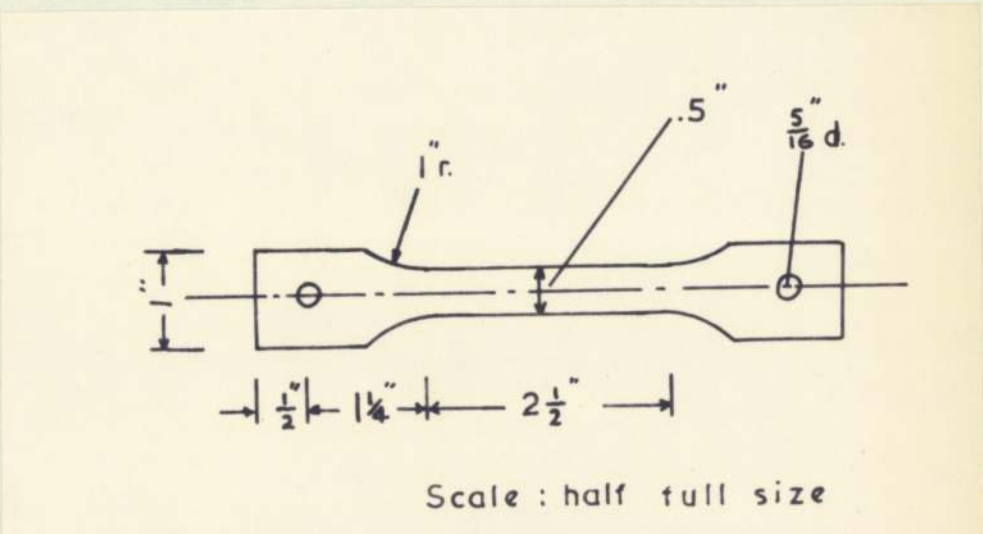


Fig. 6.5

The temper annealing treatments were performed after machining to eliminate work hardening effects due to the machining process.

6.3.1 Test Procedure

A Hounsfield Tensometer with an electrical drive was used for all the tensile testing, a constant cross head speed of 0.06"/min. being used. As each test was interrupted at least once, a variac was incorporated in the electrical drive to ensure a gradual restart to the test.

Although a load-extension curve was plotted using the semi-automatic device fitted to the machine, elongation values derived from this record are relatively inaccurate because it represents the extension of the total specimen length and includes contributions due to beam and other machine deflections. Therefore strain was calculated from measurements made directly on a 2 in. gauge length before and during the test using a cathetometer. Values of stress were derived directly from the load-extension curve. The layout of the equipment is shown in Fig. 6.6.

Measurements of the gauge width and gauge length were made at various stages during the test to enable the calculation of the anisotropy ratio R. The complete tensile test procedure was as follows:-

(1) The thickness and width of the gauge zone

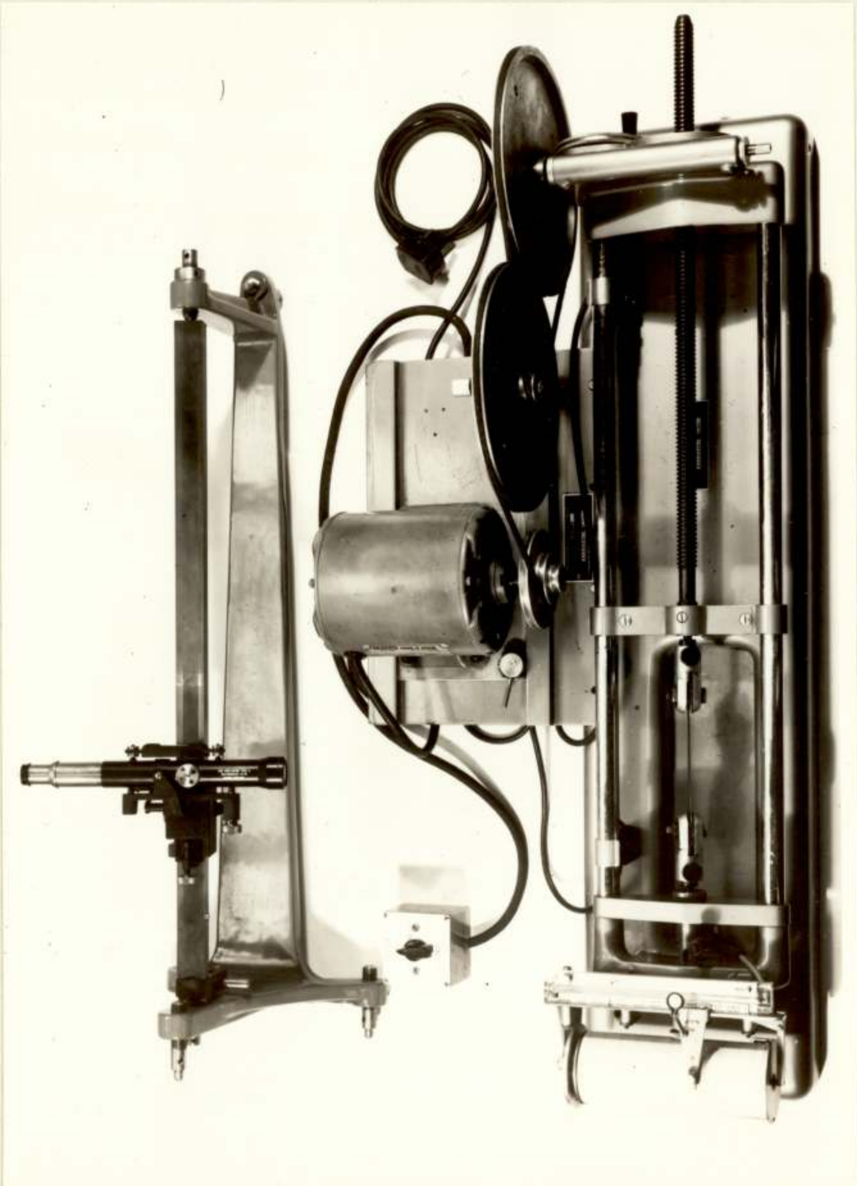


Fig. 6.6 Layout of tensile test equipment.

were each determined as the average of five measurements using a vernier micrometer reading to .0001 in.

(2) The gauge zone was painted with engineers' blue on one side, and using a template, lines were scribed at nominally $\frac{1}{4}$ in. intervals between the specimen shoulders. Only the two gauge marks enclosing the central 2 in. gauge length were required for measuring the R value and the total elongation, the remainder being used to determine the uniform elongation after fracture.

(3) The test piece was fixed into the tensometer and the gauge length measured using a cathetometer.

(4) The test was begun, and then stopped after a certain amount of plastic strain whilst width and length measurements were repeated. The loads at which the measurements were made were marked on the load-extension diagram. The test was continued and two or three more sets of measurements made up to the maximum load. Ideally width and length measurements were made at three points during the test, one after

about 10% elongation, one near to the maximum load, and the third mid-way between the other two. Where the specimen was of limited ductility, one, or in brittle test pieces, two of the measurements were omitted.

6.3.2 Parameters Obtained from Tensile Test

The following parameters were calculated from measurements made during the tensile test:-

(a) Ultimate tensile stress,

(b) true stress at tensile instability,

(c) limit of proportionality (L.P.),

(d) total elongation on a 2 in. gauge length,

(e) n, the work hardening index,

(f) U, the uniform elongation,

(g) R, the anisotropy ratio,

(h) the ratio of limit of proportionality to ultimate tensile stress.

6.4 Press Formability Tests

6.4.1 Erichsen Stretch Forming Test

Cupping and cup drawing tests were carried out on an Erichsen model 123 electro-hydraulic sheet testing machine. For the cupping or stretch forming tests, a modified version of the test conditions recommended in the Euronorm specification 14-58 was used. Briefly, the actual specification suggests that sheet of minimum width 90 mm. should be lubricated on both sides with graphitic grease, and clamped in position with a blankholder load of 1000 Kg. The Erichsen stretch forming value is the depth of penetration after failure occurs due to a slow advancement of the punch. The centres of test impressions should be at least 45 mm. from the sheet edge, and at least 90 mm. from adjacent impressions.

In this investigation, polythene lubrication was used since it is now considered that this minimises surface effects and allows more impartial measurement of the material property as opposed to the effect of the surface finish and lubrication characteristics.

Tooling dimensions are shown in Fig. 6.7.

At least two impressions were made to assess the stretch formability of each material condition.

6.4.2 The Erichsen Cup Drawing Test

The standard version of the Erichsen cup drawing test with a 33 mm. diameter flat nosed punch was used. Tooling dimensions are shown in Fig. 6.8. For each material condition, the critical blank diameter was

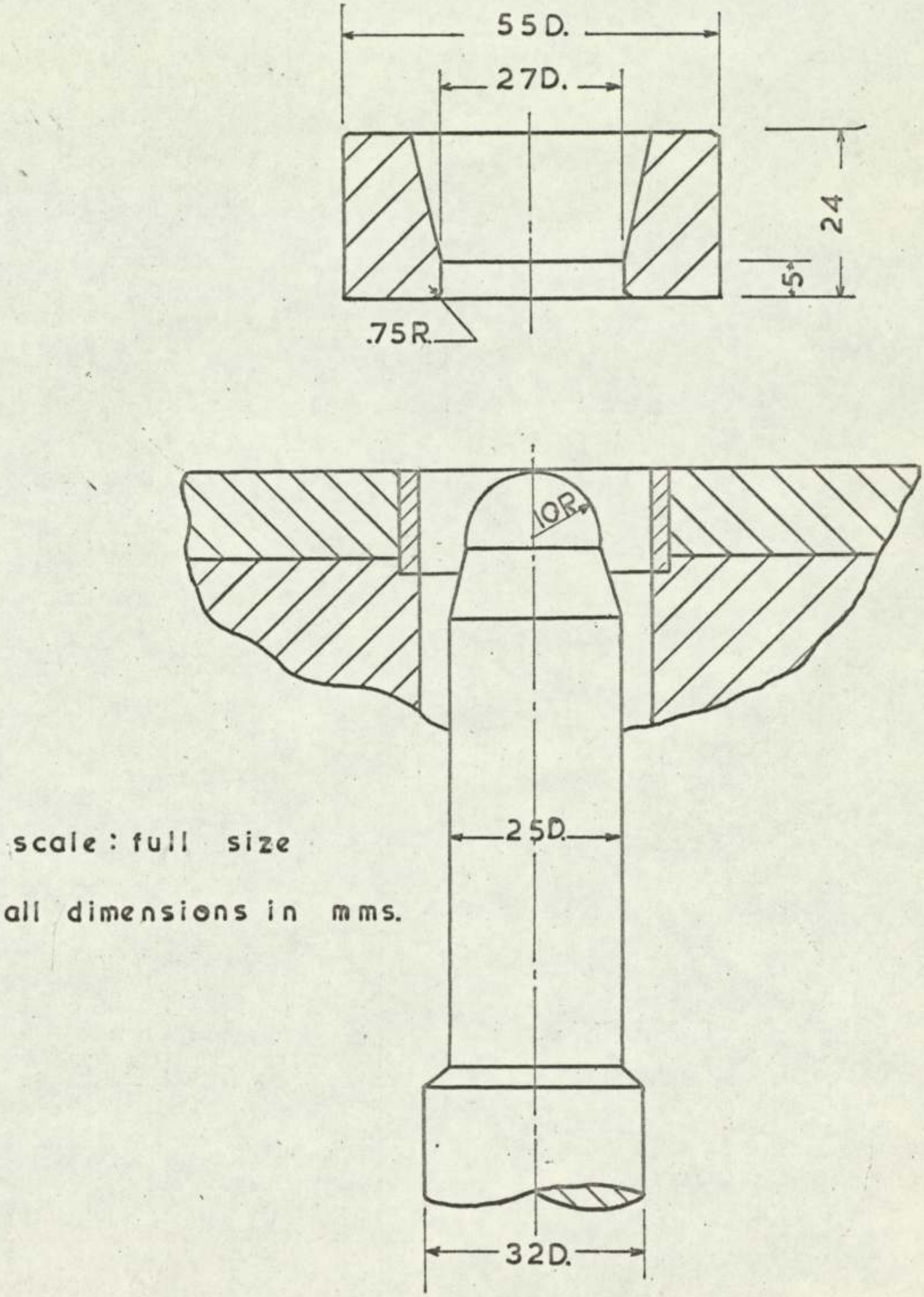
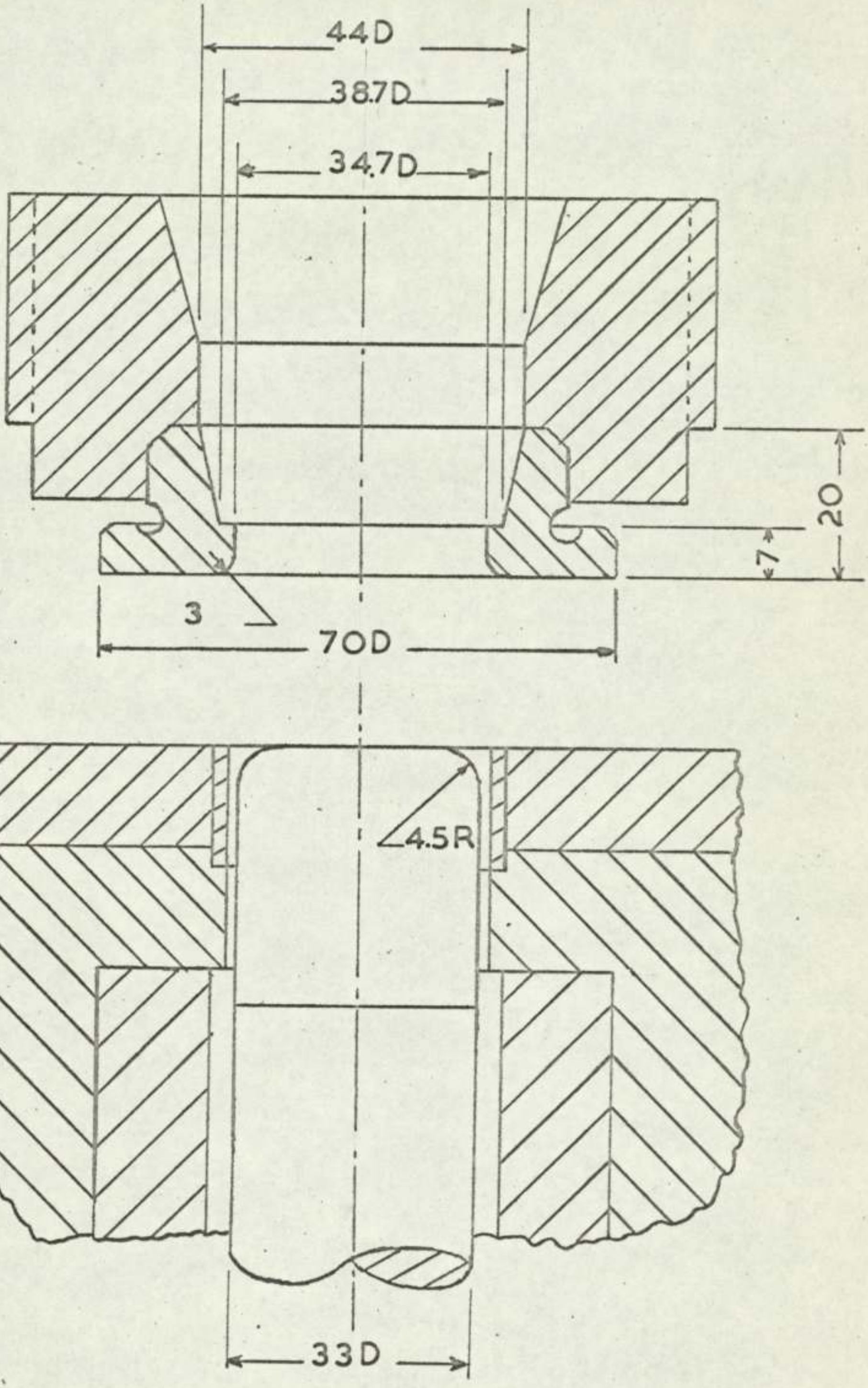


Fig. 6.7 Erichsen stretch forming test tools



scale : full size
all dimensions in mms.

Fig. 6.8 Erichsen cup drawing test tools determined.

The test blanks were punched from strip using the Erichsen blanking press. An adequate range of blanking tools in increments of 1 mm. is provided so that the critical blank diameter could be determined to the nearest millimetre.

The drawing die used was such as to give a clearance of 40% with a sheet thickness of 0.048 in. The standard clearance for the Erichsen cup drawing test is about 20%, but preliminary tests using this clearance resulted in a gross amount of cup wall ironing, to the extent where difficulty was encountered in stripping the cup from the punch. The difficulty was eliminated with the larger die, and in any case it was considered that the influence of ironing should be minimised in a test for deep drawability.

The blankholder pressure used was sufficiently great to prevent wrinkling of the blank, but not so great as to give a spurious value for the C.B.D. The optimum pressure range was determined by drawing a series of cups at different blankholder pressures. For both copper and 70/30 brass a blankholder pressure of 1000 Kg. was used.

As in the stretch forming tests, polythene lubricant was used on both sides of the blank, and the punch speed was 170 mm./min. On the basis of a linear relationship between the drawing load and blank diameter, the C.B.D. can be estimated by drawing successful cups from two blank sizes, and by drawing a third larger blank size which fails during the drawing operation. Since the normal failure load is independent of blank size, and the load for a successful draw is directly proportional to the blank size, the maximum blank diameter can be estimated by extrapolation. The actual C.B.D. was obtained by testing several blanks at the estimated C.B.D. and at the blank size 1 mm. greater. The C.B.D. was taken as that blank size at which at least five successful cups could be drawn, whilst at a blank size 1 mm. greater at least five failures occurred.

Where five successful cups were drawn at one blank diameter, say x mm., and a mixture of passes and fails at x+1 mm. diameter, the C.B.D. is referred to as x + mm. Conversely, when a single failure occurred among five passes, the C.B.D. is described as x- mm.

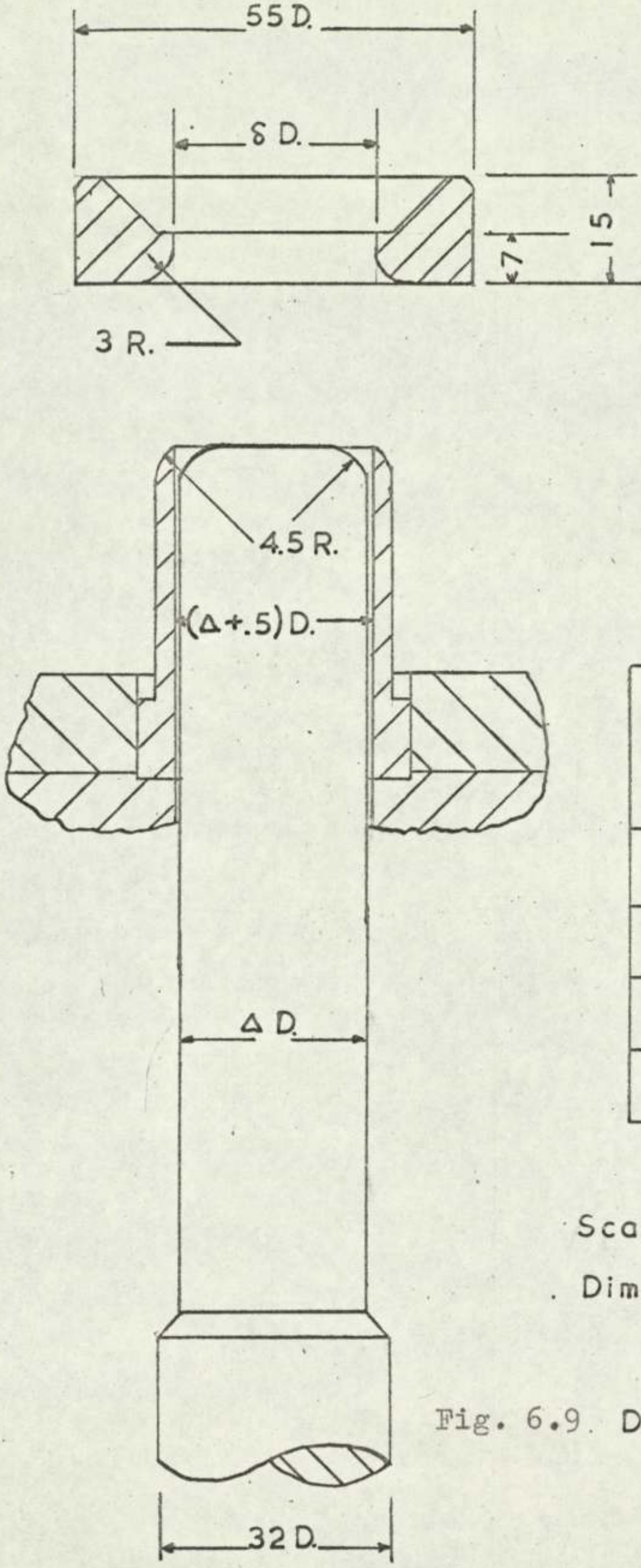
6.5 Redrawing Tests

Redrawing tests were performed on cups drawn using the Erichsen cup drawing test. Comparison of the redrawability of different materials was based on the redrawing of cups drawn from blanks of the critical diameter. Four sets of redrawing tools were available with punch diameters of 24, 25, 26 and 27 mm., the severity of the redrawing test decreasing in that order. Tool dimensions are shown in Fig. 6.9.

Graphitic grease was used as a lubricant and a clearance of 1 mm. was left between the cup and the die. A punch speed of 120 mm./min. was used, the test being assessed merely as a pass or failure. Since only certain tool sizes were available limiting redrawing reductions were not calculated.

6.6 Earing Measurements

Duplicate earing measurements were made on cups drawn from a standard blank size of 64 mm. for both copper and 70/30 brass using identical tooling to that used for the determination of the C.B.D. Since the amount of ironing occurring under these conditions was negligible, the earing measured was essentially material dependent, the influence of other factors being small. The cup height was measured by gripping the cup in a chuck which could be rotated in a horizontal plane, and a dial gauge was used to measure the height of the ears



Punch Diameter (Δ mm.)	Die Bore (δ mm.)
24	26.9
25	279
26	28.9
27	29.9

Scale : full size

. Dimensions in mm.

Fig. 6.9. Direct re-drawing tools for Erichsen cups.

and the troughs. Throughout the work, earing has been expressed in the form:-

$$\% \text{ earing} = \frac{\text{average ear height} - \text{average trough height} \times 100}{\frac{1}{2}(\text{average ear height} + \text{average trough height})}$$

6.7 Hardness Measurements

Using a Vickers hardness testing machine, the hardness of every strip was measured at several points after electropolishing to remove the surface layers. Vickers hardness measurements were also made on sectioned deep drawn cups. The cups were sectioned, cold mounted, mechanically polished and finally electropolished.

6.8 Optical Metallography

Grain size measurements were carried out on specimens which had been electropolished, and etched in dilute alcoholic ferric chloride. The method of assessing grain size was by direct comparison with the standard A.S.T.M. charts for non-ferrous materials. As a check, the grain size was actually measured for a number of specimens on a Vickers projection microscope. The agreement between the measured and the assessed values was to within $\pm$ .005 mm. at grain sizes of the order of .04 mm., so it was considered reasonable to accept the grain size assessments obtained by the comparison method.

The wall thickness of electropolished sectioned cups was measured on the screen of a Vickers projection microscope at a magnification of x100.

Copper and 70/30 brass were both electropolished in a 30% solution of orthophosphoric acid in industrial methylated spirits. A potential difference of about 15 volts was maintained between the specimen and a stainless steel cathode.

6.9 X-ray Diffraction

6.9.1 Back Reflection Technique

To supplement the results of the isochronal annealing studies, x-ray back reflection photographs were taken of all strips in temper rolled and temper annealed conditions. A Siemens x-ray set was used with a copper target operating at 40 Kv. and 16 mA. An exposure time of 1 hour was used, and a nickel filter was used to remove the β radiation.

6.9.2 Pole Figure Determination

A $\{111\}$ pole figure was plotted for each of the final material conditions from reflection data using a Siemens texture goniometer. Copper K_{α} radiation was used throughout the work.

The standard Schulz reflection method was used, and all specimens were thinned to .024 in. by etching, so that the pole figure represented the mid-section of the strip. The goniometer was motor driven and geared to simultaneously cause a 360° rotation of the specimen about the normal to the rolling plane and a 5° rotation of the specimen about the transverse direction in the rolling plane. The spiral locus of these rotations covers the area of a pole figure from the centre to the periphery. However accurate reflection data can only be obtained up to 70° from the centre of the pole figure. Beyond this limit, defocusing of the Bragg reflection occurs for sheet specimens, and the results are not very reliable.

7. EXPERIMENTAL RESULTS

7.1 Studies of Copper and 70/30 Brass in Different Conditions from Cold Rolling and Annealing.

The main object of this section of the work was to produce a fairly wide range of mechanical properties in copper and 70/30 brass by cold rolling and temper annealing treatments. For heavily cold rolled material it was usually possible to create a recovered state by temper annealing; with materials cold rolled 15%, annealing probably caused partial recovery of the lightly worked structure without recrystallisation.

For the heavily cold rolled material, the measurement of R value and n was precluded because of the limited elongation in the tensile test. For the same reasons it is possible that the R values quoted for strips in hard tempers may be inaccurate.

7.1.1 Copper

The results of the investigation on copper are reported in the tables listed below:-

Table 7.1 Tensile test results.

Table 7.2 Summarised results of press forming tests and averaged mechanical properties.

Table 7.3 The true stresses corresponding to ultimate tensile strengths in the tensile tests.

Table 7.4 Results of earing measurements on standard cups.

The value of the average mechanical properties for a particular sheet was obtained from the results of tests at 0^{o}, 45^{o} and 90^{o} to the rolling direction using the formula:-

$$\text{Average} = \frac{\text{Value at } 0^{o} + \text{value at } 90^{o} + 2(\text{value at } 45^{o})}{4}$$

It can be seen from Table 7.2 that the critical blank diameter of copper cold rolled to a Vickers hardness of 108, (C.4), is at least 2 mm. smaller than that of sheet temper annealed to a similar hardness, (C.6). A short investigation was made to see if this increase was observed in strips temper annealed after a 90% rolling reduction, but no mechanical property measurements were made on these samples. The results are shown in Table 7.5.

Identity	Condition	Hv/5kg.	C.B.D.(mm.)
C.4	Cold worked 30%	108	64+
C.5	" " 60%	124	63
C.6	C.W. 60% + 1hr @ 170^{o}C.	107	67
C.12	C.W. 88% + 1hr @ 140^{o}C.	124	65
C.13	C.W. 88% + 1hr @ 150^{o}C.	110	66

Table 7.5

Identity	Condition	L.P. (t.s.i.)			U.T.S. (t.s.i.)			L.P./U.T.S.			El'gn %			U %			n			R		
		0°	45°	90°	0°	45°	90°	0°	45°	90°	0°	45°	90°	0°	45°	90°	0°	45°	90°	0°	45°	90°
C.0	Annealed	2.57	2.39	2.05	13.43	12.29	12.20	.20	.20	.17	51.6	51.3	49.3	45	42	43	.50	.46	.47	.84	.67	.89
C.1	Cold worked 3%	5.32	5.72	6.20	13.29	12.66	16.68	.40	.45	.49	41.0	41.3	42.3	31	28	36	.27	.25	.25	.78	.66	.84
C.2	" " 6%	6.80	6.53	6.74	13.49	12.76	12.77	.51	.50	.52	42.8	46.7	36.8	36	39	31	.27	.28	.23	.81	.64	.94
C.3	" " 15%	10.28	11.47	11.06	14.85	14.14	14.44	.69	.81	.77	30.0	23.4	24.7	25	16	19	.10	.04	.05	.66	.63	.94
C.4	" " 30%	13.22	11.60	13.61	19.02	18.89	19.62	.67	.62	.69	5.1	5.3	5.5	2	1	2	-	-	-	.64	.86	.82
C.5	" " 60%	15.77	14.14	14.97	22.30	22.12	23.56	.71	.64	.64	4.5	3.8	2.9	-	-	-	-	-	-	-	-	-
C.6	C. w'd 60% + 1hr at 170°C	10.32	8.71	11.82	18.04	18.01	19.08	.57	.49	.62	12.1	10.4	9.7	6	3	3	-	-	-	.41	.80	.68
C.7	" 60% + 1hr at 185°C	6.40	4.96	6.97	15.84	14.80	16.29	.41	.34	.43	23.1	27.2	23.1	18	18	14	.15	.18	.13	.50	.81	.69
C.8	" 60% + 1hr at 195°C	3.34	3.12	3.79	13.92	13.75	14.39	.25	.23	.27	39.3	42.2	34.9	34	33	27	.29	.28	.26	.64	.81	.83
C.9	" 30% + 1hr at 215°C	5.88	5.59	6.01	15.07	13.59	14.32	.39	.41	.42	20.0	30.2	19.9	15	22	12	.20	.18	.20	.57	.70	.76
C.10	" 30% + 1hr at 223°C	2.69	2.91	3.42	12.91	12.65	13.33	.21	.23	.26	32.9	37.2	32.4	27	20	24	.34	.31	.22	.71	.79	.79
C.11	" 15% + 1hr at 280°C	2.58	2.69	2.92	12.57	13.11	13.21	.21	.21	.22	34.6	37.3	36.4	25	31	28	.25	.29	.25	.65	.75	.69

Table 7.1 Results of tensile tests on copper.

Identity	Condition	Erichsen Value (mm.)	C.B.D. (mm.)	Hv/ 5kg.	L.P. (t.s.i.)	U.T.S (t.s.i.)	L.P./U.T.S.	Elong^n %	U %	n	$\bar{R}$
C.0	Annealed	13.70	67+	44	2.35	12.57	.19	50.8	43	.47	.77
C.1	Cold worked 3%	11.67	67	65	5.74	12.82	.45	41.5	31	.25	.74
C.2	" " 6%	11.19	68	72	6.65	12.95	.52	43.3	36	.27	.76
C.3	" " 15%	10.16	67	91	11.07	14.39	.77	25.4	19	.06	.72
C.4	" " 30%	9.19	64+	108	12.50	19.10	.65	5.3	2	-	.84
C.5	" " 60%	8.42	63	124	14.76	22.53	.66	3.8	1.2	-	-
C.6	C. w'd 60% + 1hr at 170°C	9.12	67	107	9.89	18.29	.54	10.7	4	-	.68
C.7	" 60% + 1hr at 185°C	9.79	67+	92	5.83	15.43	.38	25.1	17	.21	.70
C.8	" 60% + 1hr at 195°C	11.69	68	67	3.35	13.95	.25	39.7	32	.28	.77
C.9	" 30% + 1hr at 215°C	9.85	67	95	5.76	14.15	.41	25.3	18	.19	.69
C.10	" 30% + 1hr at 223°C	11.14	67+	74	2.98	12.89	.23	34.9	23	.30	.77
C.11	" 15% + 1hr at 280°C	11.30	67	70	2.72	13.00	.21	36.4	29	.27	.71

Table 7.2 Summarised pressing and averaged mechanical properties of copper.

Identity	Condition	True stress (t.s.i.)			
		0^{o}	45^{o}	90^{o}	Av.
C.0	Annealed	19.40	17.50	17.45	17.97
C.1	Cold worked 3%	17.46	16.21	17.28	16.79
C.2	" " 6%	18.34	17.73	16.69	17.63
C.3	" " 15%	18.58	16.40	17.15	17.14
C.4	" " 30%	19.35	19.09	20.07	19.41
C.5	" " 60%	22.63	22.37	23.76	22.79
C.6	Cold worked 60% + 1hr at 170°C	19.23	18.60	19.67	19.02
C.7	" " 60% + 1hr at 185°C	18.67	17.42	18.55	18.02
C.8	" " 60% + 1hr at 195°C	18.66	18.31	18.25	18.38
C.9	" " 30% + 1hr at 215°C	17.28	16.64	15.98	16.63
C.10	" " 30% + 1hr at 223°C	16.34	15.20	16.58	15.83
C.11	" " 15% + 1hr at 280°C	15.72	17.19	16.92	16.75

Table 7.3 True stresses corresponding to U.T.S. for copper.

Identity	Condition	% earing	
		$0^{o}/90^{o}$	45^{o}
C.0	Annealed	3.32	-
C.1	Cold worked 3%	2.62	-
C.2	" " 6%	2.35	-
C.3	" " 15%	0.81	-
C.4	" " 30%	-	5.2
C.5	" " 60%	-	10.0
C.6	Cold worked 60% + 1hr at 170°C	-	6.22
C.7	" " 60% + 1hr at185°C	-	3.15
C.8	" " 60% + 1hr at 195°C	almost flat	
C.9	" " 30% + 1hr at 215°C	-	3.62
C.10	" " 30% + 1hr at 223°C	almost flat	
C.11	" " 15% + 1hr at 280°C	almost flat	

Table 7.4 Earing measurements on standard cups forcopper.

7.1.2 70/30 Brass

The results of the investigation on 70/30 brass are reported in the following tables:-

Table 7.6 Tensile test results.

Table 7.7 Summarised results of press forming tests and averaged mechanical properties.

Table 7.8 The true stresses corresponding to ultimate tensile strengths in the tensile tests.

Table 7.9 Results of earing measurements on standard cups.

7.2 Direct Redrawing Tests

The results of the direct redrawing tests on cups of the limiting size from the Erichsen cup drawing test are reported for copper in Table 7.10. 70/30 brass cups would limiting size would not redraw in any condition.

7.3 Hardness Surveys

The results of hardness surveys on a selection of cups deep drawn from copper and 70/30 brass in the rolled to temper and annealed to temper conditions are shown in Figs. 7.1 and 7.2 respectively. Although these limited surveys demonstrate the general trend of hardness with material condition, they were considered inadequate for

Identity	Condition	L.P. (t.s.i.)			U.T.S. (t.s.i.)			L.P./U.T.S.			El'gn %			U %			n			R		
		0°	45°	90°	0°	45°	90°	0°	45°	90°	0°	45°	90°	0°	45°	90°	0°	45°	90°	0°	45°	90°
B.0	Annealed	5.61	6.06	4.03	21.22	21.86	19.09	.26	.29	.22	57.6	56.4	53.0	50	49	47	.43	.48	.52	1.22	1.00	.67
B.1	Cold worked 3%	10.60	10.73	9.27	22.23	22.28	21.53	.48	.48	.43	46.6	41.8	36.7	37	34	30	.26	.25	.25	1.14	.83	.64
B.2	" " 6%	12.07	11.87	10.71	23.66	23.24	22.96	.51	.51	.48	43.5	41.0	35.6	34	34	29	.25	.19	.20	.97	.89	.65
B.3	" " 15%	9.50	14.65	11.66	26.80	25.94	25.77	.36	.56	.45	32.3	25.3	23.0	21	16	13	.10	.12	.10	1.03	1.08	.85
B.4	" " 30%	13.78	15.37	10.30	33.83	34.59	34.33	.41	.45	.30	9.8	8.7	9.6	2	2	2	-	-	-	1.55	1.23	1.82
B.5	" " 60%	15.23	17.45	17.74	41.57	43.30	44.95	.37	.40	.40	5.2	4.7	4.8	-	-	-	-	-	-	-	-	-
B.6	C. w'd 60% + 1hr at 285°C	14.74	17.30	15.93	30.40	30.62	31.48	.48	.57	.51	26.5	23.0	21.0	17	12	11	.13	.10	.10	.86	1.04	1.10
B.7	" 60% + 1hr at 295°C	10.58	10.55	12.01	27.57	27.03	27.75	.39	.39	.43	40.7	38.3	36.3	33	30	30	.22	.22	.22	.86	.97	.90
B.8	" 60% + 1hr at 320°C	10.19	7.62	10.40	24.90	24.29	24.67	.41	.31	.42	41.3	44.6	40.8	34	37	33	.29	.29	.29	.90	.94	.79
B.9	" 30% + 1hr at 315°C	11.90	10.47	11.50	27.68	26.83	28.11	.43	.39	.41	34.9	32.5	30.3	24	23	21	.20	.19	.20	.80	.98	1.02
B.10	30% + 1hr at 325°C	10.33	8.26	8.93	25.46	24.72	25.57	.41	.34	.35	40.7	40.4	37.8	33	32	30	.26	.26	.26	.82	1.04	1.04
B.11	15% + 1hr at 370°C	8.69	8.20	9.05	23.70	22.47	22.89	.37	.37	.39	43.5	47.5	43.4	35	39	33	.33	.33	.31	.93	1.00	.91

Table 7.6 Results of tensile tests on 70/30 brass.

Identity	Condition	Erichsen Value (mm.)	C.B.D. (mm.)	Hv/10kg.	L.P. (t.s.i.)	U.T.S. (t.s.i.)	$\frac{L.P.}{U.T.S.}$	Elong^n %	U %	n	$\bar{R}$
B.0	Annealed	14.50	70+	70	5.44	20.26	.27	55.9	49	.48	.97
B.1	Cold worked 3%	12.01	69+	114	10.33	22.08	.46	41.7	34	.25	.86
B.2	" " 6%	11.57	69	123	11.64	23.27	.51	40.3	33	.21	.85
B.3	" " 15%	10.10	69	146	12.62	26.12	.48	26.5	17	.11	1.01
B.4	" " 30%	8.13	69	170	14.21	34.34	.40	9.0	2	-	1.46
B.5	" " 60%	-	<55	204	16.97	43.28	.39	4.8	-	-	-
B.6	C. w'd 60% + 1hr at 285°C	8.31	64	175	16.32	30.78	.54	23.4	13	.12	1.01
B.7	" 60% + 1hr at 295°C	10.64	70	142	10.92	27.34	.40	38.4	31	.22	.94
B.8	" 60% + 1hr at 320°C	11.90	70	128	8.95	24.53	.37	42.8	35	.29	.89
B.9	" 30% + 1hr at 315°C	10.08	70	143	11.08	27.36	.40	32.6	23	.20	.94
B.10	" 30% + 1hr at 325°C	12.27	70	131	8.95	25.12	.36	39.8	32	.26	.98
B.11	" 15% + 1hr at 370°C	10.98	70-	123	8.53	22.89	.38	45.5	37	.32	.96

Table 7.7 Summarised pressing and averaged mechanical properties of 70/30 brass.

Identity	Condition	True stress (t.s.i)			
		0°	45°	90°	Av.
B.0	Annealed	31.81	31.03	26.66	30.13
B.1	Cold worked 3%	30.44	29.81	27.99	24.51
B.2	" " 6%	31.66	31.09	30.07	30.98
B.3	" " 15%	32.44	30.05	29.24	30.45
B.4	" " 30%	34.37	35.34	35.18	35.06
B.5	" " 60%	42.35	44.12	45.53	44.03
B.6	Cold worked 60% + 1hr at 285°C	35.51	34.23	34.93	34.73
B.7	" " 60% + 1hr at 295°C	36.57	35.13	35.95	35.70
B.8	" " 60% + 1hr at 320°C	33.27	33.17	32.75	33.09
B.9	" " 30% + 1hr at 315°C	34.45	33.06	34.03	33.65
B.10	" " 30% + 1hr at 325°C	33.77	32.70	33.22	33.10
B.11	" " 15% + 1hr at 370°C	32.07	31.22	30.44	31.24

Table 7.8 True stresses corresponding to U.T.S. for 70/30 brass.

Identity	Condition	% earing	
		$0^o/60^o$	45^o
B.0	Annealed	1.6	-
B.1	Cold worked 3%	-	2.03
B.2	" " 6%	-	2.15
B.3	" " 15%	-	3.06
B.4	" " 30%	-	3.28
B.6	Cold worked 60% + 1hr at 285°C	-	3.08
B.7	" " 60% + 1hr at 295°C	-	3.15
B.8	" " 60% + 1hr at 320°C	-	2.03
B.9	" " 30% + 1hr at 315°C	-	3.01
B.10	" " 30% + 1hr at 325°C	-	2.87
B.11	" " 15% + 1hr at 370°C	-	3.39

Table 7.9 Earing measurements on standard cups for 70/30 brass.

Identity	Condition	Blank dia. (mm.)	Redrawing punch dia. (mm.)			
			27	26	25	24
C.0	Annealed	67	F	-	-	-
C.1	Cold worked 3%	67	F	F	-	-
C.2	" " 6%	68	P	F	-	-
C.3	" " 15%	67	P	F	F	F
C.4	" " 30%	64	P	P	P	P
C.5	" " 60%	63	P	P	P	P
C.6	C. w'd 60% + 1hr at 170°C	67	P	P	P	P
C.7	" 60% + 1hr at 185°C	67	P	F	F	F
C.8	" 60% + 1hr at 195°C	68	P	F	-	-
C.9	" 30% + 1hr at 215°C	67	P	F	-	-
C.10	" 30% + 1hr at 223°C	67	P	F	-	-
C.11	" 15% + 1hr at 280°C	67	P	F	-	-

P indicates a successful redraw
F indicates a failure

Table 7.10 Redrawing tests on copper.

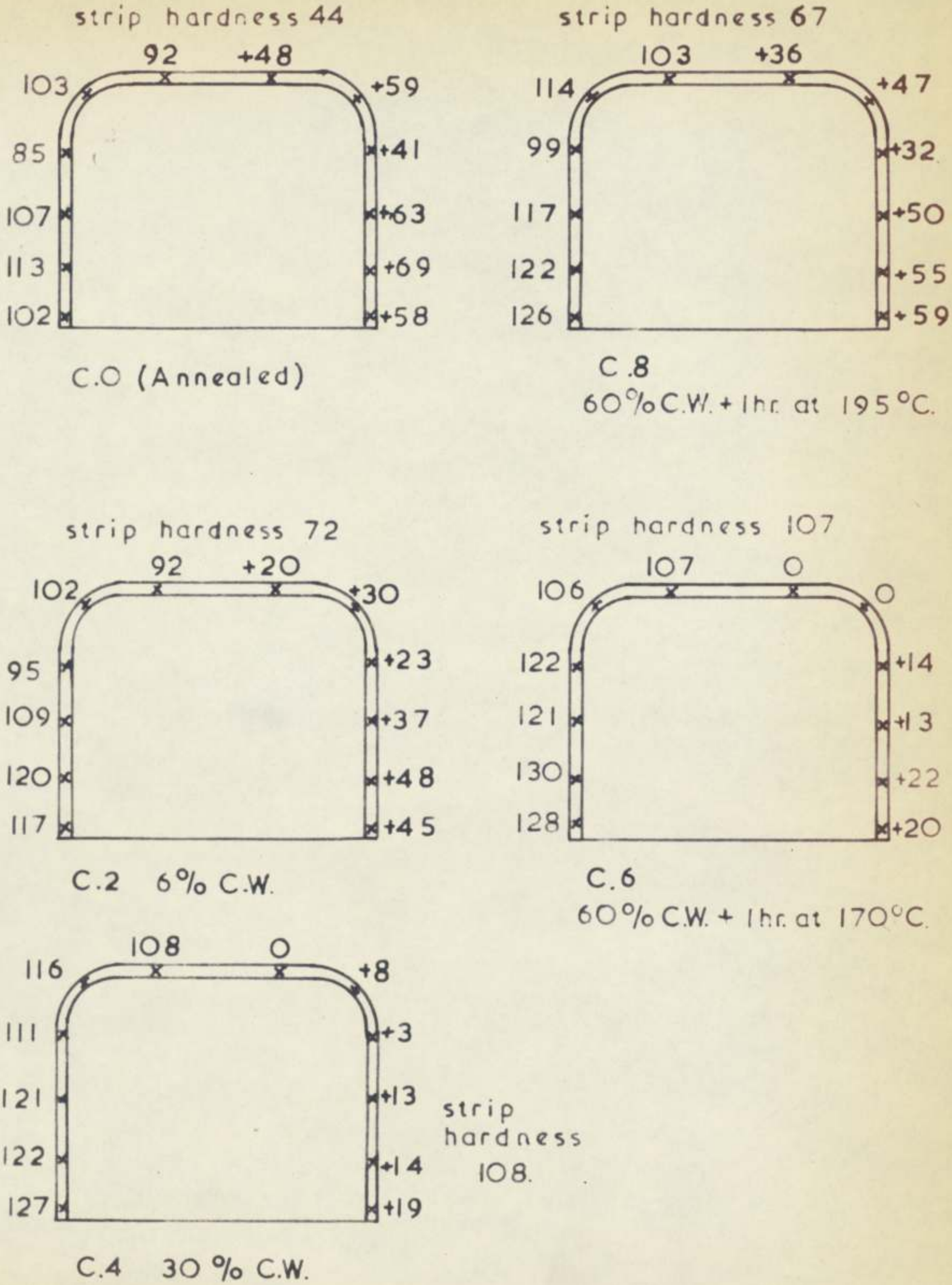


Fig. 7.1 Hardness surveys on sectioned cups drawn from copper.

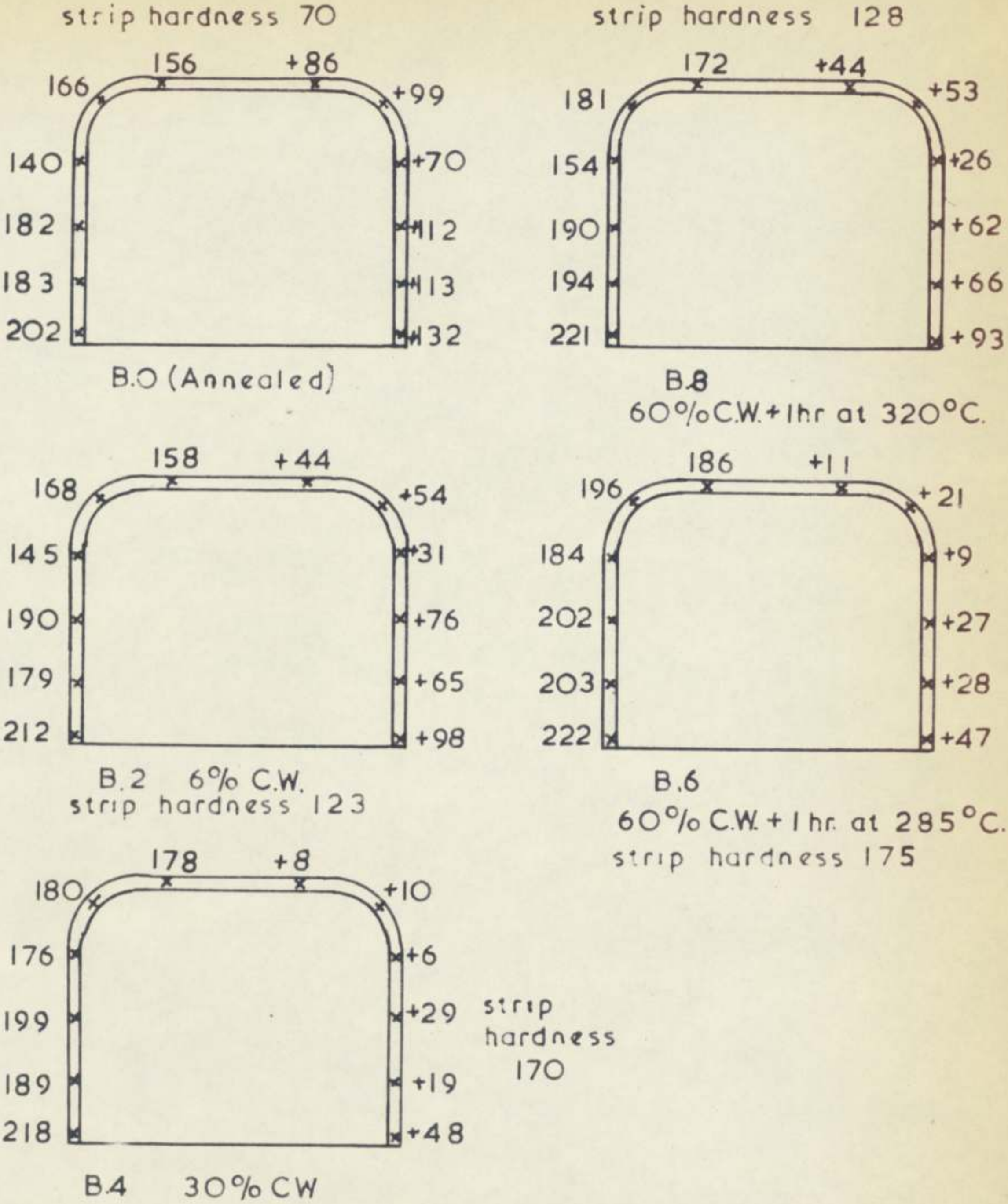


Fig. 7.2 Hardness surveys on sectioned cups drawn from 70/30 brass.

explaining some of the results of the Erichsen cup drawing tests. More complete hardness surveys, together with thickness strain distributions are introduced in the relevant section of the discussion.

8. DISCUSSION

8.1 Relationship between Erichsen Stretch Forming Values and Mechanical Properties

Since there has been some disagreement about which tensile test parameter, if any, should be used to most accurately predict stretch formability it was decided to calculate the degree of correlation between Erichsen values and some tensile test parameters for copper and 70/30 brass. These are shown in Table 8.1. The product moment correlation coefficient is a measure of the degree of correlation between two set of data. A perfect direct relationship between two sets of data would give a straight line when plotted graphically, and also a correlation coefficient of +1. Conversely, a perfect inverse relationship is indicated by a coefficient of -1, whilst a random distribution of points in graphical form is represented by a coefficient of zero. Hence the nearer the correlation coefficient is to ± 1, the higher is the degree of correlation between the two sets of data.

The best relationship for both materials was the general upward trend of Erichsen value with increasing uniform elongation. For copper, the correlation coefficient for this relationship was appreciably better than those between Erichsen value and other tensile test parameters. For the 70/30 brass however, the differences between the correlation coefficients, excluding that of Erichsen value with "n", is not so great and is probably too small to be significant, so that there is equally good correlation between stretch formability as measured by the Erichsen value, and the U.T.S., uniform elongation and elongation to fracture.

Parameter	Correlation Coeff.	
	Copper	70/30 Brass
Uniform elongation	.94	.95
Total elongation	.83	.92
"n"	.85	.86
U.T.S.	-.78*	-.91

Table 8.1

Since the elongation to fracture and U.T.S. are easier to measure than the uniform elongation, one or other of these would be the preferable parameter for predicting Erichsen values. For copper, the correlation coefficient between Erichsen value and uniform elongation is appreciably greater than that for the other parameters. Therefore for the most accurate prediction of stretch forming performance in copper, it is considered worth

* Value low because this relationship was less linear than the others in Table 8.1.

while to measure the uniform elongation in a tensile test. It should be pointed out however that all the correlation coefficients in Table 8.1 represent a high level of significance and could arise by chance considerably less frequently than once in a thousand trials. The relationships between hardness, U.T.S., uniform elongation and Erichsen values are shown in Figs. 8.1 to 8.3 for copper and 70/30 brass.

Several workers have suggested that the "n" coefficient measured as the slope of the true stress/true strain curve should be of greater significance than other tensile test parameters, on the basis that the work hardening characteristics will be of primary importance in determining stretch forming performance. The results show that a high degree of correlation exists between Erichsen value and "n", but not so high as that between Erichsen value and uniform elongation. A possible reason why the "n" value did not show a higher degree of correlation with Erichsen value may be due to the true stress/true strain relationships for copper and 70/30 brass not obeying the Ludwik law precisely, in which case the measurement of "n" is not justified. Since the true stress/true strain relationships were not determined in detail it is only

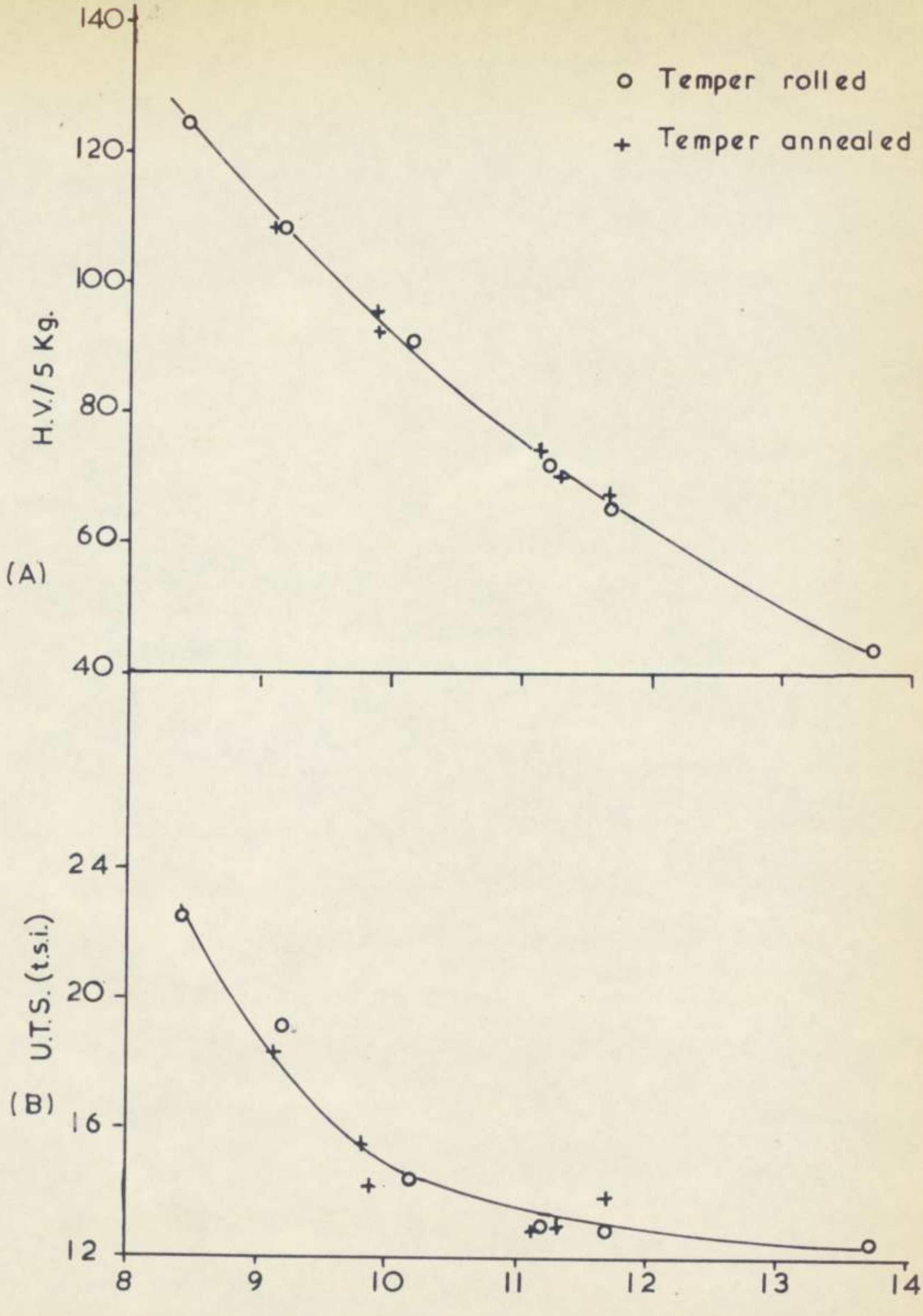


Fig 8.1 Copper : Erichsen value against (a) hardness (b) UTS

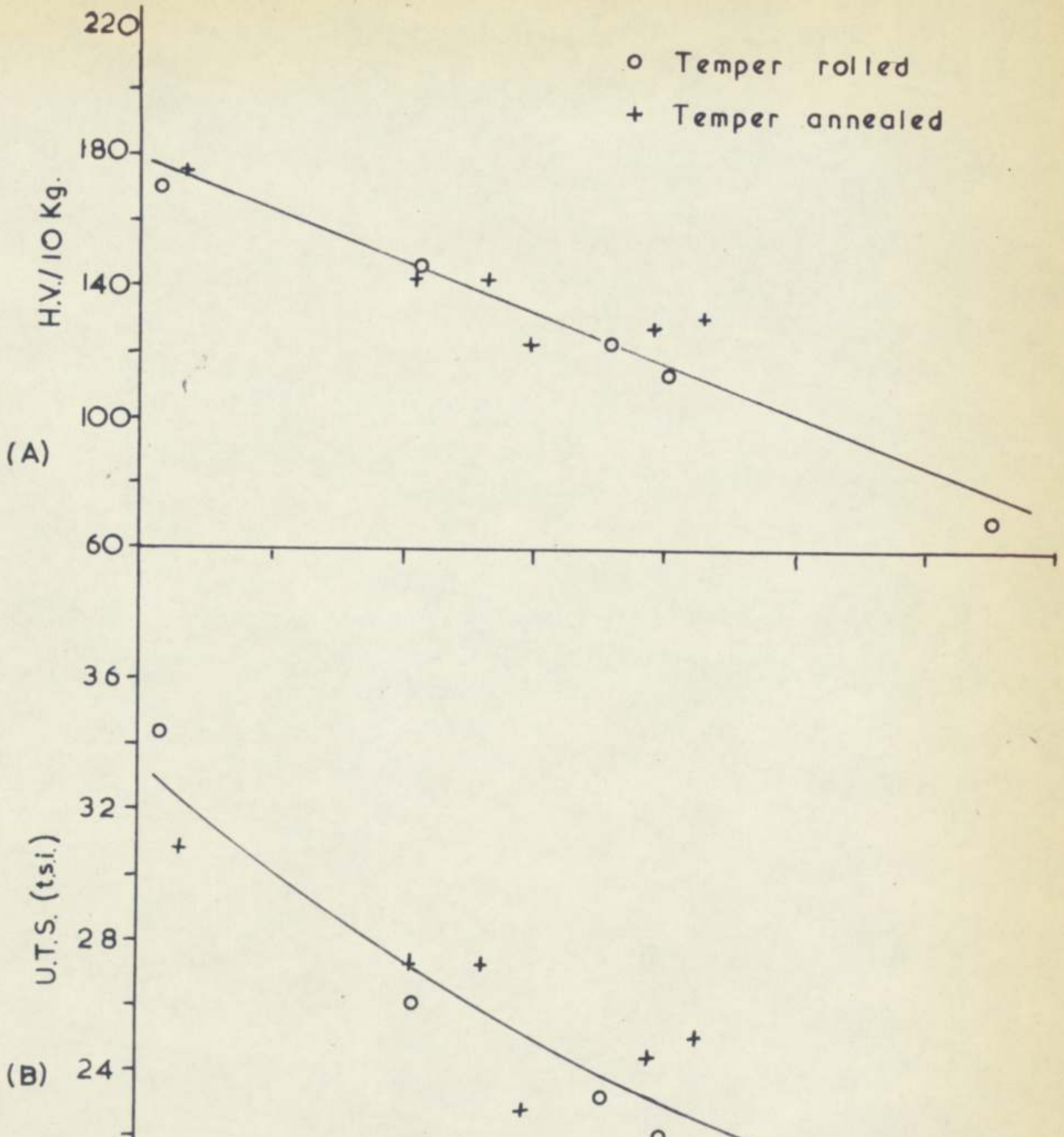


Fig. 8.2 70/30 brass : Erichsen value against (a) hardness (b) U.T.S.

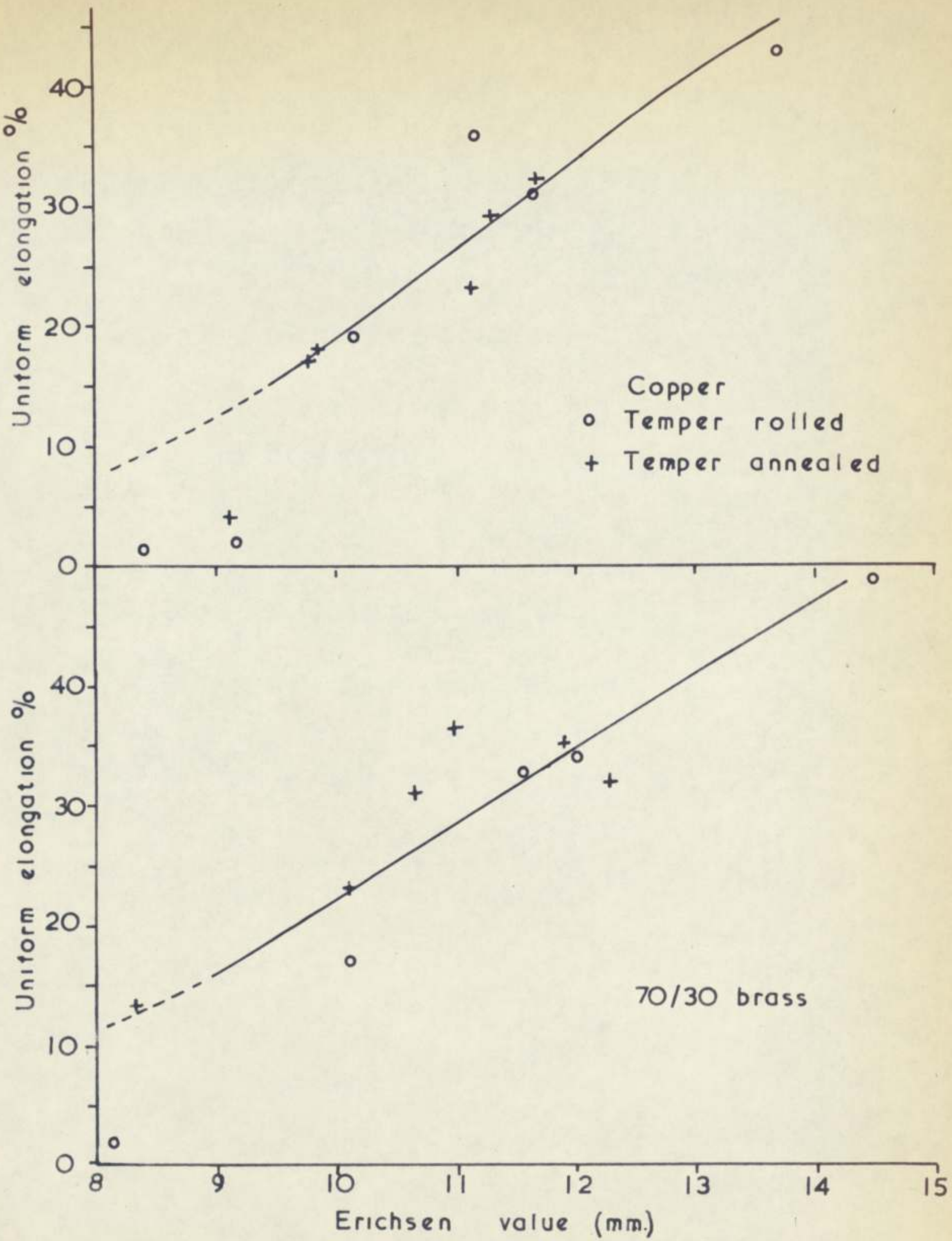


Fig. 8.3 Erichsen value against uniform elongation

possible to state that from the true stress/true strain data considered, the Ludwik law was obeyed approximately. An alternative, and probably more likely, explanation for the lower correlation coefficient between "n" and stretch formability is due to the significant degree of error in the measurement of "n", especially in cold rolled strip.

There is a significant relationship between hardness values and Erichsen values, and also between U.T.S. and Erichsen values for both copper and 70/30 brass. Since the U.T.S. calculated from tensile test data only considers the maximum load and the original cross sectional area, it cannot be regarded as a fundamental physical property. The U.T.S. fails to account for changes in uniform elongation in strips of different temper for example, and can therefore only justifiably be considered as a performance function which directly assesses strength and indirectly assesses elongation or work hardenability. It is not unreasonable to expect some relationship between stretching performance and elongation, or between work hardenability and stretching performance, and hence some reasonable relationship between Erichsen values and U.T.S.

The relationship between Erichsen values and hardness can be rationalized in a similar way. Grimes[10], in considering the physical significance of the hardness test, has pointed out that the hardness value given by a Vickers test on a perfectly plastic material is approximately proportional to the yield stress. He reasons that in a metal which strain hardens, the Vickers hardness number provides a measure of the yield stress as augmented by the indentation process, and thus regards the relationship between Erichsen value and hardness as a relationship between Erichsen value and a parameter containing yield stress and work hardening characteristics. Since the yield stress should be of little relevance to the stretch forming performance, the excellent relationship between hardness and Erichsen value must be extremely dependent on work hardening characteristics.

The influence of the strain ratio R on stretch forming performance has not been studied in detail. Theoretically, it has been shown that increasing R value should produce a greater stable extension in biaxial stretching. Keeler and Backofen[49] concluded that stretch forming performance would be improved by a higher R value only if diffuse necking occurs. However, Moore and Wallace[50] have pointed out that failure due to diffuse necking occurs infrequently in commercial pressing operations, the usual mode of failure being a localized neck. They considered the effect on R on stretch forming when failure was due to localized necking, but still concluded that stretch formability should be improved by an increase in R value. All the treatments of stretch formability have ignored planar variation in mechanical properties. Recently, Pearce[54] has demonstrated an increase in bulge height with increasing R value, but also with increasing uniform elongation, in aluminium cold rolled between 32% and 16%. He further showed that between 16% and 0% rolling reduction, when the R value was constant, the bulge height increased with uniform elongation, but at a decreased rate. He associated the sharp decrease in the rate of increase of R value and bulge height with uniform elongation, with a discontinuity in the relationship between polar thickness strain and uniform elongation. These results appear to support the theoretical predictions discussed earlier. However, there is a significant degree of error in the measurement of R value and uniform elongation on cold rolled specimens, so that the values quoted by Pearce for aluminium cold rolled greater than 16% are questionable.

There appeared to be no systematic dependence of Erichsen value on R value for copper and 70/30 brass. Individual cases could be quoted which support the theoretical prediction that R value should improve stretch formability, but even when considering these cases, the relative magnitudes of the increases in Erichsen value and R value vary widely. In any case, it is difficult to see why stretch formability should be dependent on R value, since deformation can only occur by thinning so that the resistance to thinning in a tensile test, the R value, would appear to be of little importance in determining the extent of deformation. An increase in Erichsen value might be expected if more uniform thinning occurred, but it is difficult to see why this should depend on R value. The R value may become of primary importance in general press forming operations when a stretch formed region must possess sufficient strength to draw in the flange area, when resistance to thinning of the stretch formed region should be high.

Since the Erichsen value depends on material thinning, it seems likely that the extent of deformation during stretch forming will be vitally dependent upon the uniformity of thinning prior to the point of instability. On this basis, the most suitable parameter for predicting stretch forming behaviour would be the uniform elongation measured in the tensile test.

The Erichsen values of annealed 70/30 brass and strips cold rolled up to 6% reduction were superior to copper strip in similar conditions. At rolling reductions greater than 6% however, the stretch formability of copper was marginally better than that of 70/30 brass, but the significance of the result is uncertain because the differences in Erichsen values were small.

Temper annealing of copper and 70/30 brass always resulted in an increase in stretch formability over the same strip in the cold worked condition, but the resultant Erichsen value was generally similar to that of strip which was cold rolled to the same hardness. For copper, there was little difference in Erichsen value between strips which were rolled to temper and strips which were annealed to temper when compared on a hardness or elongation basis. Similarly, there was little difference between elongation values of temper rolled and temper annealed copper except for the temper annealed strip, C.6, whose total elongation was appreciably greater than that of strip cold rolled to the same hardness, C.4. However this increased elongation was not reflected in the cupping test, since the Erichsen values of both the temper rolled sheet and the temper annealed sheet were almost identical.

A comparison between rolled to temper and annealed to temper 70/30 brass on a hardness basis shows a marginal, but probably insignificant improvement in Erichsen value of temper annealed material. When compared on a uniform elongation basis, strips which were rolled to temper stretch formed better than temper annealed strips.

In general, the best stretch formability of the materials investigated was obtained when their uniform elongation was greatest. Significant correlations were also obtained between Erichsen value and the parameters total elongation, U.T.S., "n" value and hardness, although no systematic dependence of stretch formability on R value was observed. For a given elongation, materials rolled to temper stretch formed better than materials annealed to temper.

8.2 Relationships between Deep Drawability and Mechanical Properties

8.2.1 Deep Drawability and R Value for all Materials

According to Whiteley's theory, the C.B.D. should be singularly dependent on R value provided frictional conditions remain constant.

When the copper and 70/30 brass in all conditions are considered together, there is a general upward trend of C.B.D. with R value, which is shown in Fig. 8.4. If the strips of hard temper are ignored, that is those which lie away from the general trend in Fig. 8.4, the correlation coefficient between C.B.D. and R value for the remaining strips is 0.92. This suggests that there is a strong dependency of C.B.D. on R value when the copper and 70/30 brass are considered together.

When the copper and 70/30 brass are considered individually in all conditions, the correlation coefficients between C.B.D. and R value are -0.47 and -0.15 respectively. However it is not suggested that a decrease in R value leads to improved deep drawability, but a more likely explanation is that the R values measured on materials of limited elongation in the tensile test are inaccurate. If the data from specimens of limited ductility is ignored, the correlation

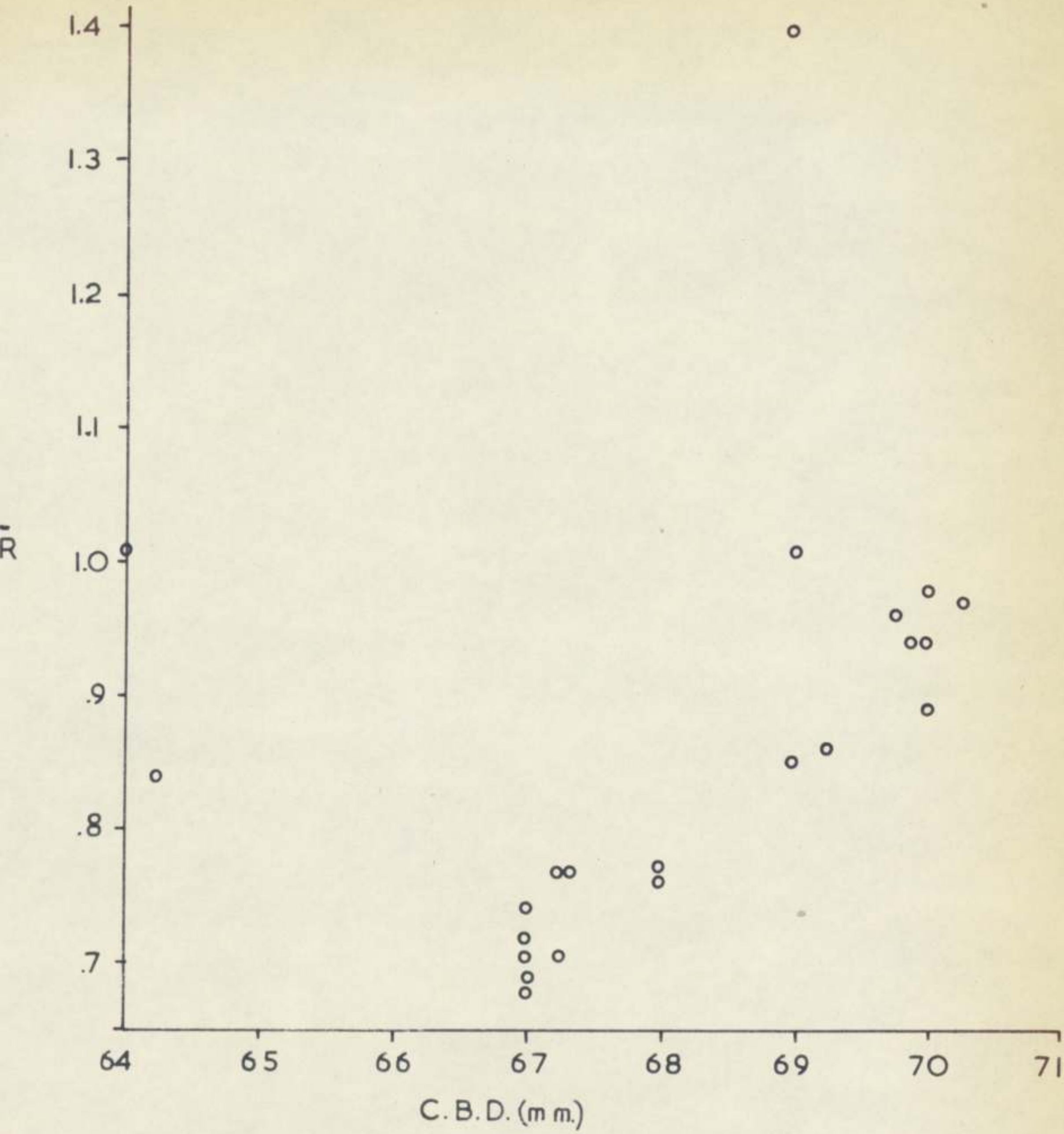


Fig. 8.4 C.B.D. against $\bar{R}$ value for copper and 70/30 brass in all conditions

coefficients for copper and 70/30 brass are 0.65 and 0.22 respectively. The relationship between C.B.D. and R value for copper then becomes significant, with a probability of arising by chance of less than 1%, but the relationship between C.B.D. and R value for 70/30 brass, although having a positive correlation, is not significant.

A comparison of the overall trend between C.B.D. and R value with the individual trends for each material shows the limitations of precise predictions of C.B.D. or R value from published relationships which take into account several different materials and conditions. From the overall trend observed in this work, one can only safely say that the C.B.Ds. of 70/30 brass in soft to medium-hard tempers will usually be greater than those of copper in the same range of temper. Further, since the correlation coefficients between C.B.D. and R value for each material are considerably worse than the overall correlation coefficient, it appears that at best one can only say that the C.B.D. of copper may be moderately influenced by R value, whilst the C.B.D. of 70/30 brass appears to be independent of R value. considering only the range of R values in this work.

8.2.2 Cold Rolled Material

The effect of cold work on the C.B.D. was different for each material. For copper, the C.B.D. reached a maximum in sheet which was cold rolled by 6%, followed by a progressive decrease in C.B.D. with rolling reductions up to 60%. The 70/30 brass showed a decrease in C.B.D. with cold rolling reductions up to 6%, after which the C.B.D. remained constant up to a rolling reduction of 30%. After 60% cold rolling reduction, 70/30 brass would not deep draw at blank diameters as small as 55 mm. This indicates a limiting rolling reduction for 70/30 brass beyond which the material will not deep draw, but within which, the C.B.D. is only 1mm. less than that of annealed material.

Since the overall pattern of mechanical properties with increasing cold work was similar for both copper and 70/30 brass any possibility of an overall correlation between C.B.D. and a simple mechanical property was immediately ruled out. Most of the mechanical properties showed either a progressive increase or decrease with cold work, with the exception of the L.P./U.T.S. ratio, the true stress corresponding to the U.T.S., and in the case of copper the elongation.

Deep drawing capacity, L.P., U.T.S., and L.P./U.T.S. ratio are shown plotted against rolling reduction for copper and 70/30 brass in Figs. 8.5 and 8.6 respectively.

For copper, the ratio L.P./U.T.S. increased rapidly in sheets cold rolled up to 15% and then decreased gradually up to 60% rolling reduction. This behaviour is similar to that of the C.B.D. with rolling reduction which reached a maximum at 6% reduction, then decreased with further rolling reduction. For the 70/30 brass the pattern of the L.P./U.T.S. ratio with increasing cold work was similar to that for copper, but the C.B.D. of 70/30 brass differed from that of copper in that no increase was observed after small rolling reduction, although it is possible that the C.B.D. might increase at rolling reductions of 1 - 2%.

The precise significance of the L.P/U.T.S. ratio will differ according to the initial material condition, since the U.T.S. is primarily a strength parameter for strips of hard temper, but for strips in the soft condition the U.T.S. includes, indirectly, elongation. In all cases the L.P. should approximate to the initial flow stress, so that for materials in soft temper the ratio L.P/U.T.S. effectively provides a measure of initial flow stress, maximum strength and elongation.

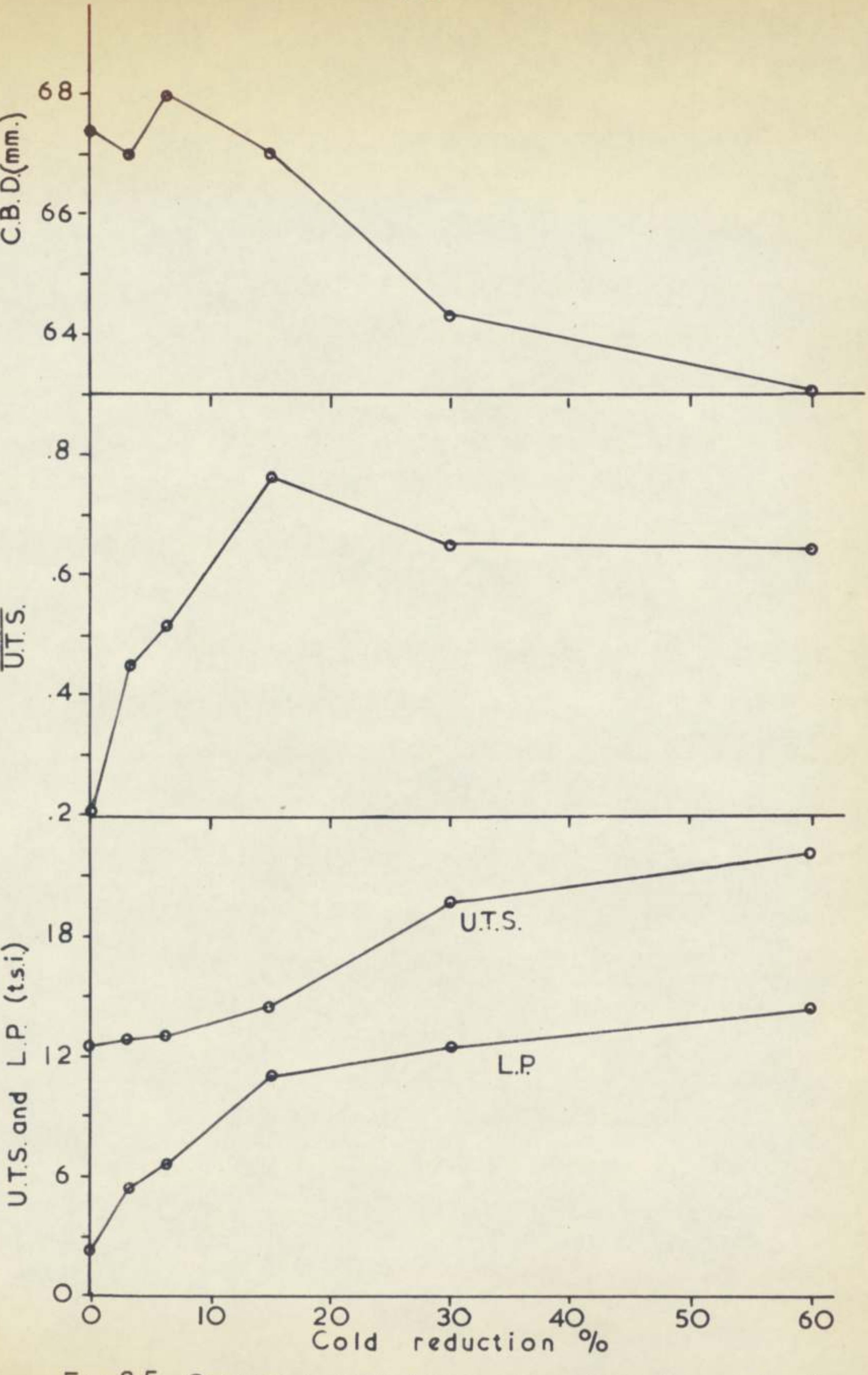


Fig. 8.5 Copper : variations in strength and deep drawing capacity with cold work

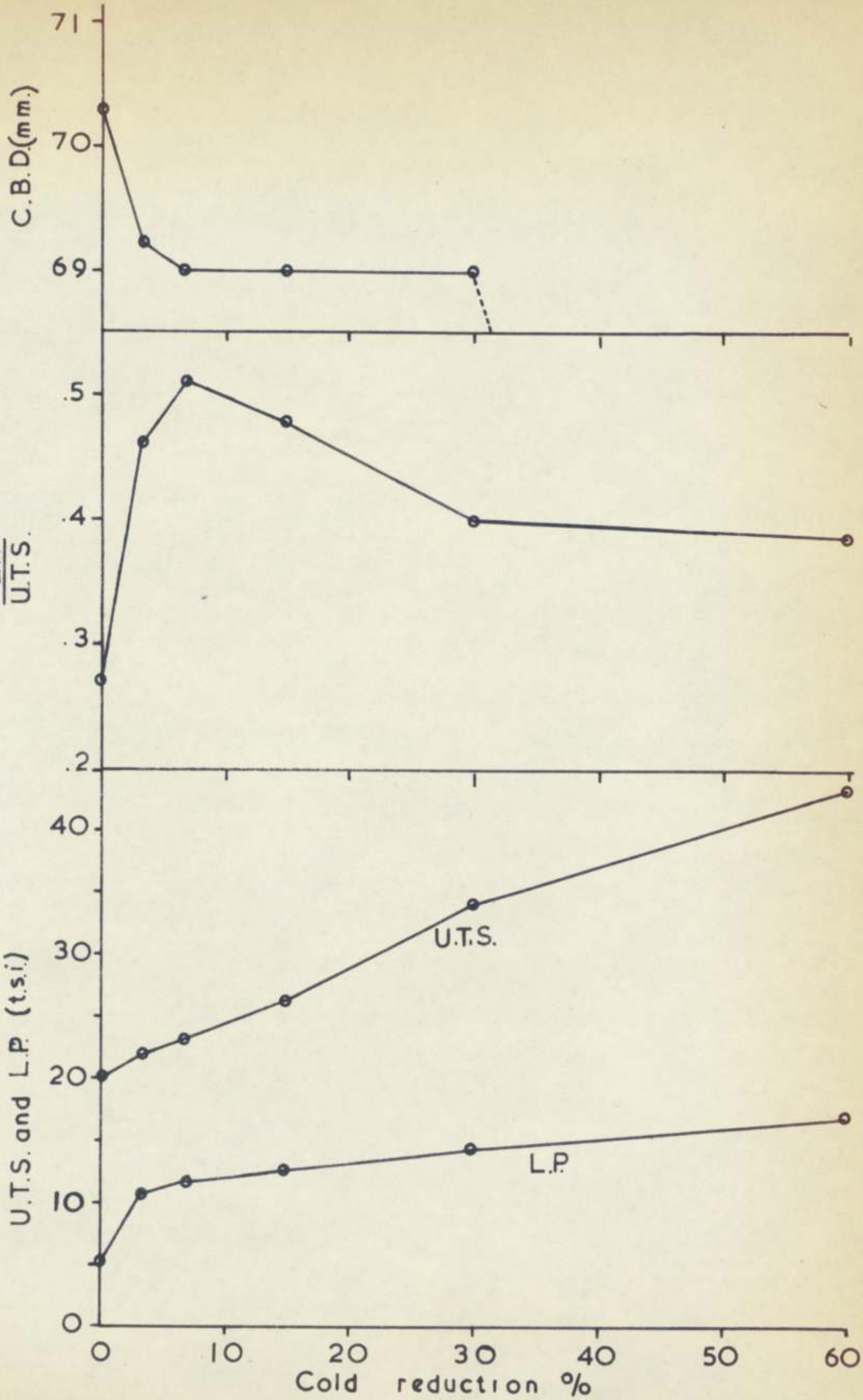


Fig. 8.6 70/30 brass : variations in strength and deep drawing capacity with cold work.

For material in harder tempers the ratio L.P./U.T.S. will therefore be close to the ratio of the initial flow stress to the flow stress at instability in a tensile test. An increase in the ratio L.P./U.T.S. at small rolling reductions has been observed previously in aluminium[10] and copper[62], and is not too surprising a result. The L.P. and the U.T.S., as measured in a tensile test, will both increase with increasing rolling reduction, but the initial rate of increase of the flow stress will be greater than the initial rate of increase of the U.T.S. Consequently there will appear a maximum value of the ratio L.P./U.T.S. after small rolling reductions. The difference in values between the U.T.S. and the L.P. with increasing rolling reduction only reflects the change in the work hardening rate as measured in the tensile test. The fact that the C.B.D. was a maximum after small rolling reductions similar to those at which the L.P./U.T.S. ratio was a maximum is considered to be coincidental. In any case, it is generally believed that a decrease in the ratio L.P./U.T.S. is beneficial to deep drawability.

The variations in C.B.D. and true stress at instability with rolling reduction for copper and 70/30 brass are shown in Fig. 8.7. An initial decrease in true stress corresponding to U.T.S. with cold work has not been widely reported. Grimes[10] found a similar decrease in true stress in aluminium and some of its alloys after rolling reductions within the range 10% - 20%. He associated the decrease in true stress corresponding to U.T.S. with the sub-grain formation which occurs during plastic deformation of aluminium and its alloys. When aluminium is plastically deformed by cold rolling, it has been shown by transmission electron microscopy that after small rolling reductions, a sub-grain structure develops, the minimum size of sub-grain generally being observed at rolling reductions which are similar to those at which a decrease in the true stress corresponding to U.T.S. was observed. However the explanation of why a decrease in sub-grain size leads to a decrease in true stress corresponding to U.T.S. was not made clear.

It is now well established that small amounts of plastic deformation result in the formation of sub-grains in copper, and it is equally well established that sub-grains do not form in 70/30 brass, even after

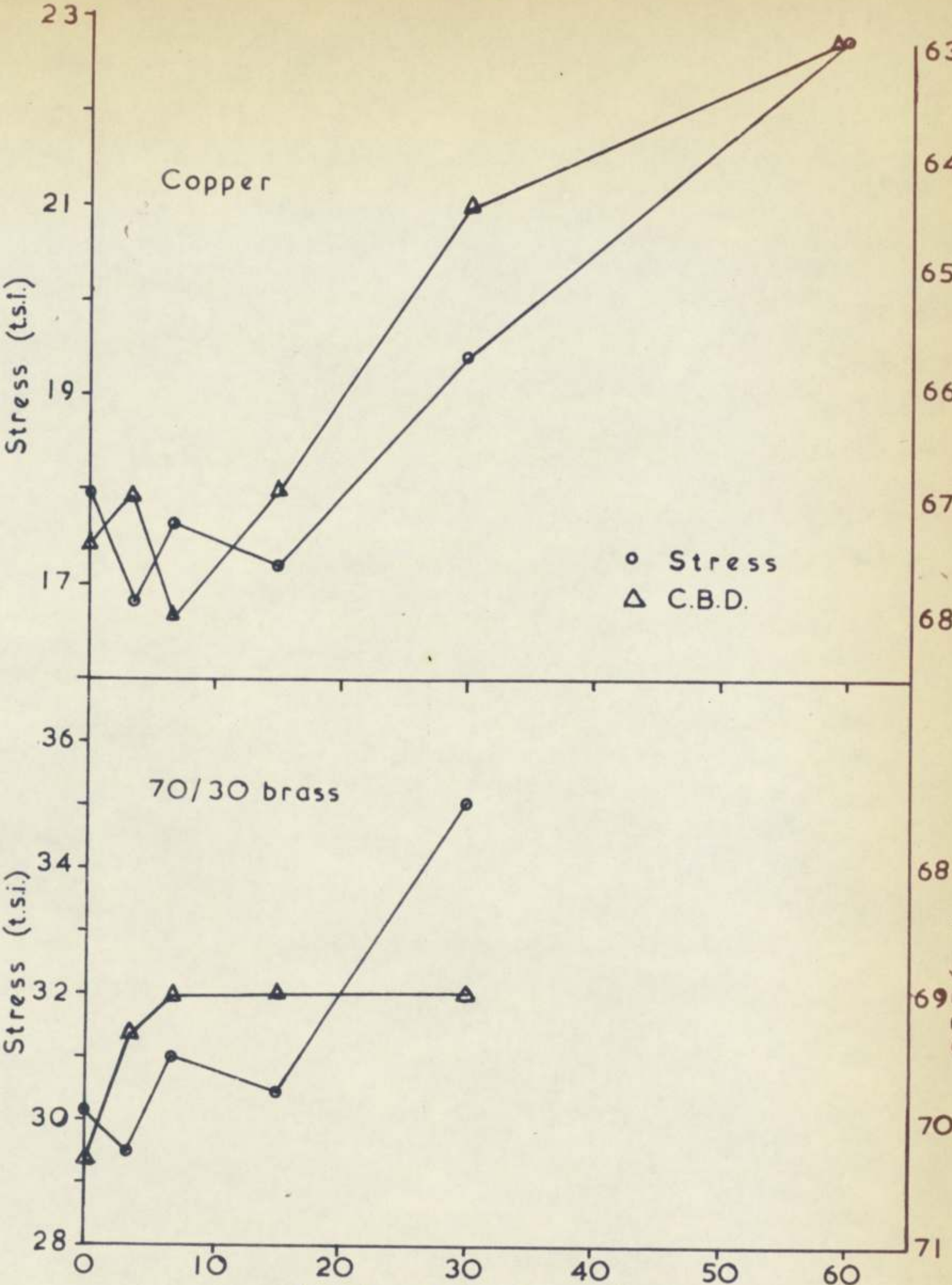


Fig. 8.7 True stress at tensile instability and deep drawing capacity against cold reduction.

substantial amounts of plastic deformation. However, the decrease in true stress corresponding to U.T.S. was observed in both copper and 70/30 brass, although the decrease for 70/30 brass was not so marked as for copper. It therefore seems unlikely that the decrease in true stress at tensile instability is associated with sub-structure formation in the early stages of plastic deformation.

Another possible explanation for the decrease in true stress at instability at small rolling reductions may be that a change in work hardening characteristics occurs due to a change in the preferred orientation. Pole figures for both copper and 70/30 brass, Figs.8.20 and 8.21 respectively, show that although the positions of peak intensities do not alter significantly there is a substantial change in the numerical value of the peak intensities after small rolling reductions, especially for copper.

Since the true stress corresponding to U.T.S. takes into account the uniform elongation during the tensile test, it would appear that variation in the value of the true stress with increasing rolling reduction should be explainable in terms of variation in either the U.T.S. or the uniform elongation with increasing rolling reduction. If the relative rate of change of the U.T.S. and the uniform elongation varies with increasing rolling reduction, one would not necessarily expect a continuous relationship in graphical form between the true stress corresponding to the U.T.S. and the rolling reduction. For example, there is little difference in the value of U.T.S. for copper strips in the annealed condition and after receiving 3% and 6% cold work. However, after 3% cold work there is a considerable reduction in the uniform elongation which is proportionately greater than the increase in U.T.S. Consequently it is not surprising that the true stress at instability of copper strip cold rolled 3% is less than that of annealed strip. Similarly, after 6% cold rolling reduction the uniform elongation is greater than in strip cold rolled 3%, whilst the U.T.S. also increases, and consequently there is an increase in the true stress after 6% rolling reduction. Similar arguments can be applied to the tensile data for 70/30 brass, to explain the discontinuities in the relationship between true stress and percentage cold work.

The important point is why the uniform elongation varies in such a manner. One source of this variation has already been mentioned, that due to changes in the intensity of the components of the textures after small rolling reductions. Another explanation for the observed variation in uniform elongation may be due to errors in the measurement of this property. Some contribution from inaccurate measurement is inevitable in all specimens, and in particular as the total elongation decreases with increase in cold work, measurement of the uniform extension will become more inaccurate. However, in specimens cold rolled 3% and 6%, the inaccuracy introduced is unlikely to be significantly greater than that on a specimen in the annealed condition, especially when the total elongation in all three conditions was greater than 40%. In any case, the copper strip cold rolled 6% which showed an increase in uniform elongation compared with strip cold rolled 3%, also showed an increase in the total elongation. Since all the values of mechanical properties quoted for a strip are the average of six specimens, two in each of the directions at 0°, 45° and 90° to the rolling direction, the final figure is considered to be fairly representative of the whole strip. Therefore although there is inevitably some error in the final value of elongation, the variation in error is not thought to be responsible for the observed variation in the true stress. A more probably explanation is that the effect is due to a majority contribution from the changes in preferred orientation and consequent changes in work hardening characteristics, and a minority contribution from errors in measurement of the tensile properties.

Having accounted for the increases in the ratio L.P./U.T.S. and the true stress at tensile instability with increasing rolling reduction for copper and 70/30 brass, the only other mechanical property which showed a discontinuous change with rolling reduction was the elongation for copper, which increased several percent after 6% rolling reduction compared with strip cold rolled 3%. Possible reasons for this change in terms of changes in work hardening characteristics due to texture changes after small rolling reductions, have been discussed. If this increase in uniform elongation is considered in relation to the increase in C.B.D. of copper strip cold rolled 6%, and in the light of the thickness strains measured, a possible explanation of the increased drawability is suggested.

It is now realised that the stretch formability of a material can influence the deep drawing behaviour. Even in the Erichsen cup drawing test, which sets out to simulate pure deep drawing, the stretch formability of a material is of prime importance, since failure nearly always occurs in the stretch formed zone over the punch-nose radius. An extreme example of this is the case of 70/30 brass cold rolled 60%, strip B.5, whose Erichsen stretch forming value was less than 2 mm., but which failed over the punch nose due to lack of biaxial ductility. The same material would not deep draw at blank diameters as low as 55 mm. and always failed in the early stages of deformation on the punch nose radius due to an inability to stretch rather than an inability to bend. However, 70/30 brass cold rolled 30%, whose Erichsen value was only about 8 mm., had a C.B.D. of 69 mm. Evidently the biaxial ductility, although relatively poor, was high enough to allow sufficient stretch forming over the punch nose radius to occur, and this combined with a high strength enabled the deep drawing of a high blank diameter.

The thickness distribution in cups deep drawn from blanks of the C.B.D. for copper and 70/30 brass are shown in Figs. 8.12 to 8.17. The distance from the pole of the base of the cup is plotted along the abscissa, and the percentage thickness strain on the ordinate. If the thickness strain distribution for cups deep drawn from annealed copper strip and strip having been cold rolled 6% are compared, Fig. 8.12, it can be seen that the maximum thinning over the punch nose radius was about 35% for the annealed material, C.0, whereas for the 6% cold rolled strip, C.2, at a blank diameter 1 mm. greater, the maximum thinning was only about 20%. This is not too surprising since the uniform elongation in the tensile test was greater for annealed copper than for cold rolled copper. However, upon attempting to deep draw a cup from annealed copper using a blank diameter of 68 mm., the remaining area after thinning 35% over the punch nose radius was insufficient to transmit the extra load required to draw in the greater flange diameter, although about two thirds of the draw had been completed at failure. Copper cold rolled 6%, although having an inferior Erichsen value compared with annealed copper, had adequate ductility in this respect to stretch over the punch nose radius without necking down under the greater deep drawing load, since the amount of thinning was less and the strength of the thinned region was greater than that of annealed copper. A simple calculation of maximum punchloads may help to support this argument. For example, assuming that in cups of the critical size drawn from annealed copper the maximum thinning of 35% is constant around the cup circumference, one can calculate the area transmitting the drawing load at this point. If no thinning occurs, the transverse cross-sectional-area of the cup would be approximately:-

= mean circumference of cup x strip thickness

$= \pi \times 34.3 \times 0.048/25.4$ sq. ins.

= <u>0.21 sq. ins</u>.

After 35% thickness strain, the area of the thinned zone is 0.65 x 0.21 sq. ins. Assuming the true stress in the zone of maximum thinning as an instability criteria, one can calculate this true stress from a knowledge of the total strain history of the strip after rolling and deep drawing. For annealed copper the total true strain is only due to thinning in deep drawing:-

$$= \log_n \frac{0.048}{0.031} \quad \begin{pmatrix}\text{initial strip thickness}\\ \text{thickness in thin zone}\end{pmatrix}$$

$= \underline{0.44}$

From the relationship between the true stress of material after rolling and the true strain incurred during rolling the true stress corresponding to a true strain of 0.44 is 19.8 t.s.i. Thus the maximum punch load which can be transmitted by the thinned zone in annealed copper is:-

= true stress x area of thinned zone

= 19.8 x 0.65 x 0.21 tons

= <u>2.71 tons</u>

For cups at the C.B.D. drawn from copper strip cold rolled 6%, and having thinned 20% the total true strain is that due to rolling and deep drawing:-

$$= \log_n \frac{0.051}{0.048} + \log_n \frac{0.048}{0.038}$$

$$= \log_n 1.07 + \log_n 1.20 = 0.0677 + 0.1823$$

$$= \underline{0.2500}$$

The true stress corresponding to a true strain of 0.25 is 18.65 t.s.i. Therefore the maximum punch load is:-

= 18.65 x 0.21 x 0.80 tons

= <u>3.13 tons</u>

Copper strips cold rolled greater than 6%, although possessing adequate strength were insufficiently ductile to accommodate the necessary stretch forming at blank diameters greater than the C.B.D. and failed due to necking over the punch nose radius. Basically therefore, the reason for the maximum value of C.B.D. in copper strip cold rolled 6% is because the particular combination of mechanical properties were optimum for deep drawing. Under different conditions of lubrication which is known to affect stretch formability, this may not have been true.

8.2.3 Annealed to Temper Material

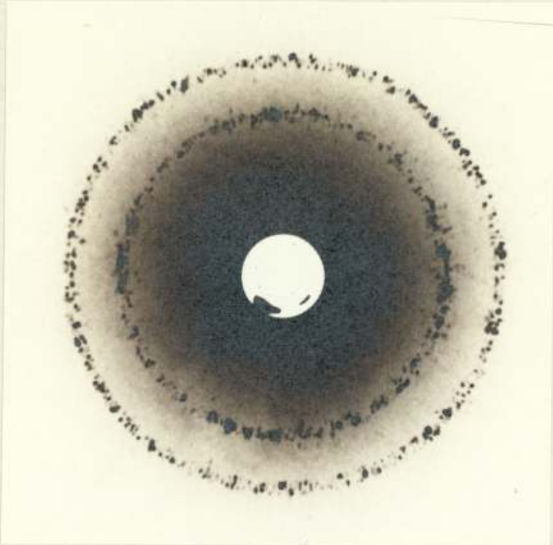
The main object of this part of the investigation was to see if annealing to temper offered any metallurgical advantages over rolling to temper with respect to press formability. In the discussion on stretch forming behaviour, it was shown that there was little difference in Erichsen values between temper rolled and temper annealed strip, and in some cases, temper annealed strip showed a decrease in stretch formability, when compared on a hardness or elongation basis with temper rolled strip.

In general for copper and 70/30 brass there was an increase in C.B.D. when strip was annealed to temper compared with strip rolled to a similar temper. For copper, the increase in C.B.D. was small in most temper annealed strips when partial recrystallisation had occurred. A significant increase in C.B.D. was observed in copper strips which were temper annealed within the recovery range, so that the amount of recrystallisation was small. The C.B.Ds. of recovered strips are shown in Table 7.5 together with the C.B.Ds. of strips cold rolled to similar hardnesses. In every case the C.B.D. is at least 2 mm. greater in the temper annealed condition. Figs. 8.8 to 8.11 show a selection of x-ray back reflection photographs of copper and 70/30 brass after temper rolling and temper annealing. The x-ray photograph of temper annealed copper strip, C.6, Fig. 8.9 (a), whose C.B.D. was at least 2 mm. greater than temper rolled strip of similar hardness, C.4, indicates that very little recrystallisation occurred during the annealing treatment. An comparison of the mechanical properties shows that the U.T.S. of the temper annealed strip, C.6, is slightly lower than that of strip C.4 rolled to a similar hardness, whilst the elongation of strip C.6 is appreciably greater than that of strip C.4. The maximum thinning over the punch nose radius in cups at the C.B.D. deep drawn from strip C.4 was less than 20%, whereas in cups drawn from strip C.6, the maximum thinning exceeded 30%. Using similar calculations to before, it can be shown that the maximum punch load that a cup at the C.B.D. deep drawn from strip C.4 could transmit having thinned 20% in the stretch formed zone is 3.51 tons. However, the maximum load that a cup at the C.B.D. drawn from strip C.6 could transmit having thinned 30% is only 3.14 tons. Therefore on a strength basis, one might expect the deep drawability of strip C.4 to be greater than that of strip C.6, but in fact the C.B.D. of the temper rolled strip was 2 mm. less than that of the temper annealed strip. It is considered that the increased drawability of the temper annealed strip C.6 is a consequence of its greater tensile elongation, which together with the smaller L.P./U.T.S. ratio of this strip, allowed a greater degree of stable necking in the stretch formed zone at blank diameters greater than would have been expected from strength considerations alone. Evidently both strips C.4 and C.6 had adequate strength in the cup wall to transmit the drawing load, but the ductility appeared to be the controlling factor. A similar recovery of ductility is thought to be responsible for the superior deep drawability of copper strips temper annealed after 90% rolling reduction.

The x-ray back reflection photographs, Fig. 8.11(a to c), showed that in all the temper annealed 70/30 brass strips some recrystallisation had occurred.

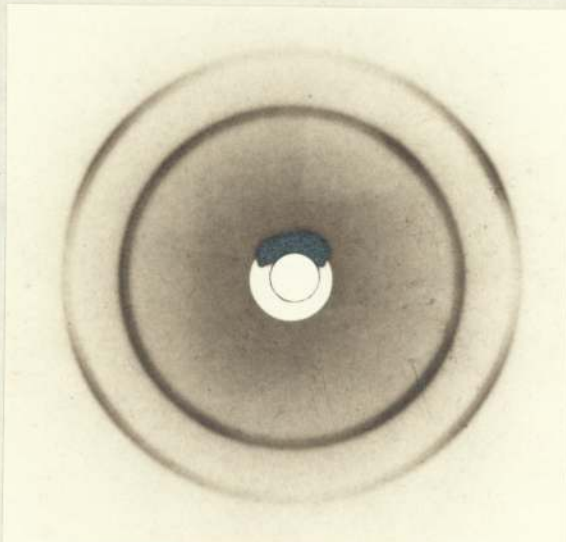
Strips which were annealed to temper generally showed a slight increase in C.B.D. compared with strip rolled to a similar hardness level. Most 70/30 brass strips after temper annealing had a C.B.D. of 70 mm., whereas most temper rolled strips had a C.B.D. of 69 mm. The exception to this trend was that of strip temper annealed at the lowest temperature, B.6, whose C.B.D. was 4 mm. less than that of strip rolled to a similar temper, which is a pattern of behaviour opposite to that observed in copper.

However the mode of failure in cups drawn from strip B.6 differed from that in strip B.4, and the corresponding copper strips C.4 and C.6, in that failure occurred in the cup wall well below the punch nose radius. Evidently, the strip B.6 possessed sufficient ductility to complete the punch nose radius and subsequently to draw in to some extent before the drawing load exceeded the strength of the cup wall and caused failure. Comparison of the mechanical properties of strips B.4 and B.6 shows that after temper annealing the U.T.S. was about 12% lower than that of temper rolled strip, and even though the tensile ductility and Erichsen value were greater in the temper annealed



(a)

(b)

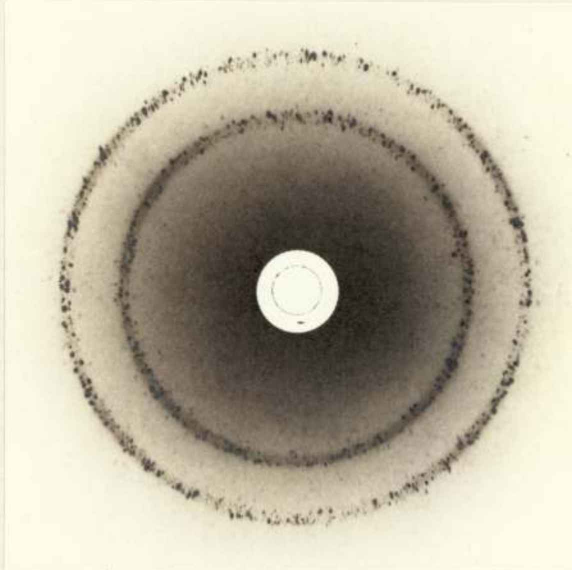


(c)

Fig. 8.8 X-ray back reflection photographs of annealed and cold rolled copper.

(a) annealed
(b) C.W. 6%
(c) " 30%

(a)



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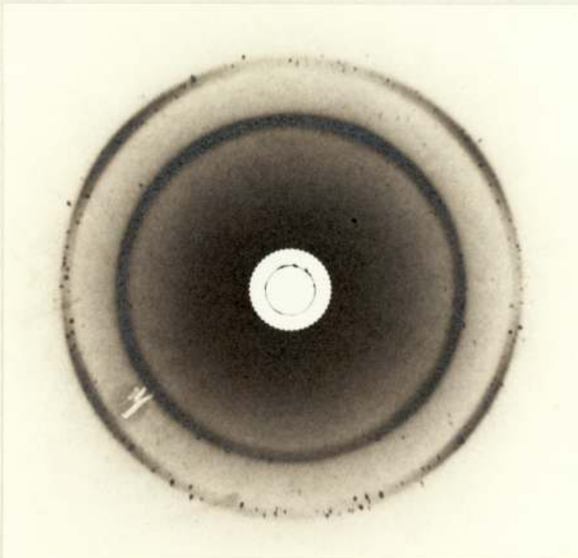
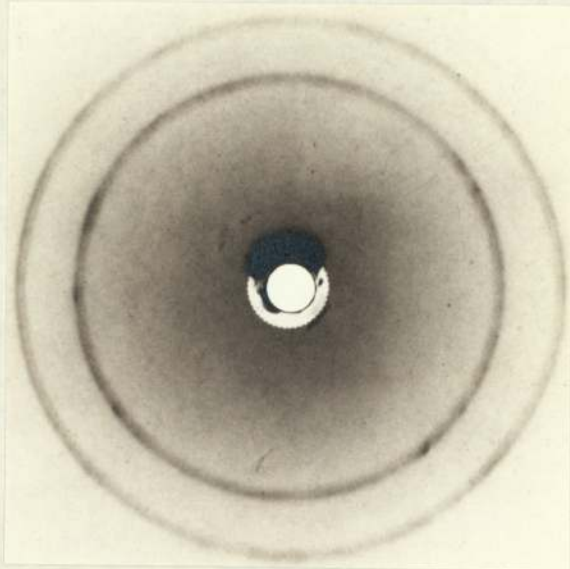


(c)

Fig. 8.9 X-ray back reflection photographs of temper annealed copper.

(a) C.W. 60% + 1hr at 170°C
(b) " 60% + 1hr at 185°C
(c) " 30% + 1hr at 215°C

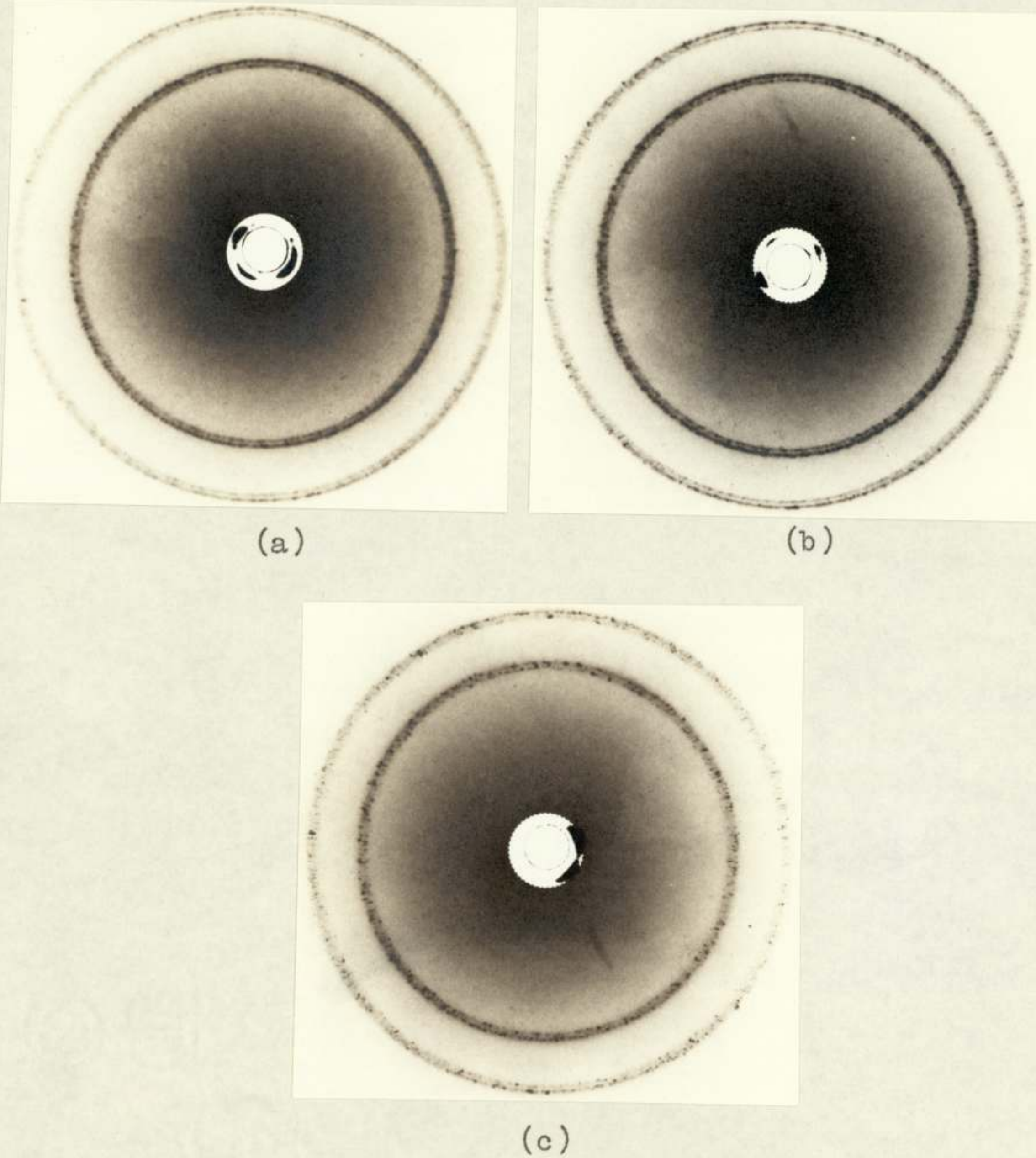
(a)



(b)

(c)

Fig. 8.10 X-ray back reflection photographs of annealed and cold rolled 70/30 brass.
(a) annealed
(b) C.W. 6%
(c) " 30%



(a) (b) (c)

Fig. 8.11 X-ray back reflection photographs of temper annealed 70/30 brass.

(a) C.W. 60% + 1hr at 285°C
(b) " 60% + 1hr at 295°C
(c) " 30% + 1hr at 325°C

condition, the strength of the cup wall was the controlling factor. In addition, the hardness surveys Figs. 8.15 to 8.17, showed that the general hardness of the cups deep drawn from temper annealed strip was higher than in those deep drawn from temper rolled strip, indicating a greater work hardening of the blank, which although strengthening the cup wall also increased the load required to draw in the flange.

In general, temper annealing of copper strip increased the deep drawability compared with temper rolled strip. When the temper annealed strip had been previously heavily cold rolled and the annealing temperature was low, a greater increase in C.B.D. compared with temper rolled strip was found. For 70/30 brass, the reverse trend was observed in that strip temper annealed at the higher temperatures generally showed a slight improvement in C.B.D., whereas the strip cold rolled 60% and temper annealed at the lowest temperature (B.6) suffered a decrease in C.B.D. The differences in C.B.D. between temper annealed and temper rolled copper and 70/30 brass have been explained by comparison of mechanical properties, and thickness strains during deep drawing.

8.3 Thickness and Hardness Surveys on Deep Drawn Cups

The distribution of thickness strain and hardness measured on sectioned cups deep drawn from blanks of the critical size for copper and 70/30 brass are shown in Figs. 8.12 to 8.17. The thickness strain distributions basically showed the same characteristics in that generally two distinct zones of thinning occurred, referred to here as zones A and B, and shown for example in Fig. 8.12 (b). Thinning in zone A was usually greater than in zone B and was caused by a stress combination due to stretch forming and plastic bending under tension over the punch nose radius in the initial stages of the cup drawing operation. Thinning in zone B was due to plastic bending under tension over the die radius, and began at the same time in the drawing sequence as the thinning in zone A. Consequently there was a zone of material in between zone A and zone B which suffered no plastic bending under tension, and very little biaxial stretching, zone C, Fig. 8.12 (b), which as a result was considerably thicker than zones A and B. As drawing in of the flange progressed, the thinning over the die radius became less evident because of the gradual thickening of the blank due to

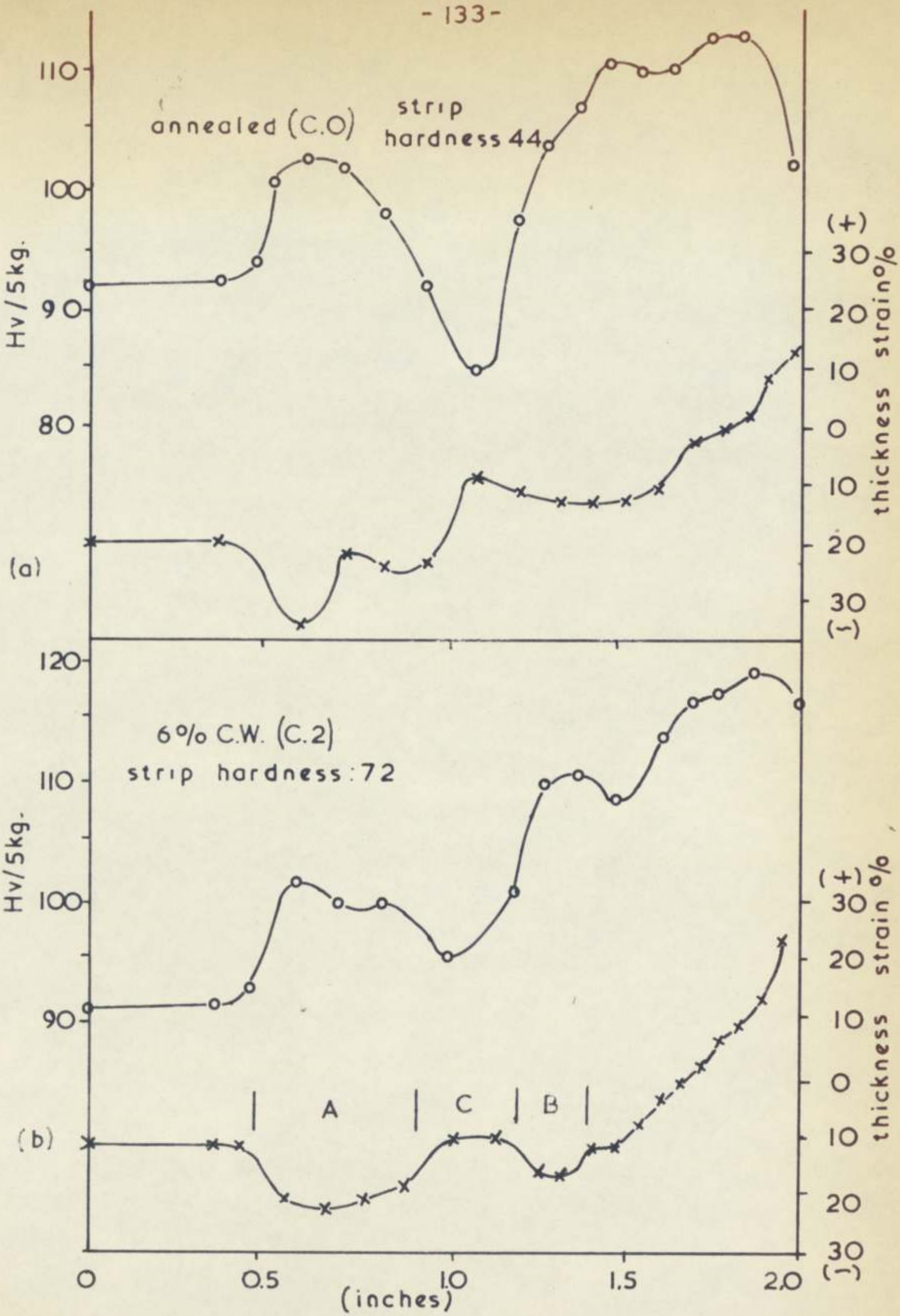


Fig. 8.12 Hardness and thickness strain against distance from cup centre for copper.

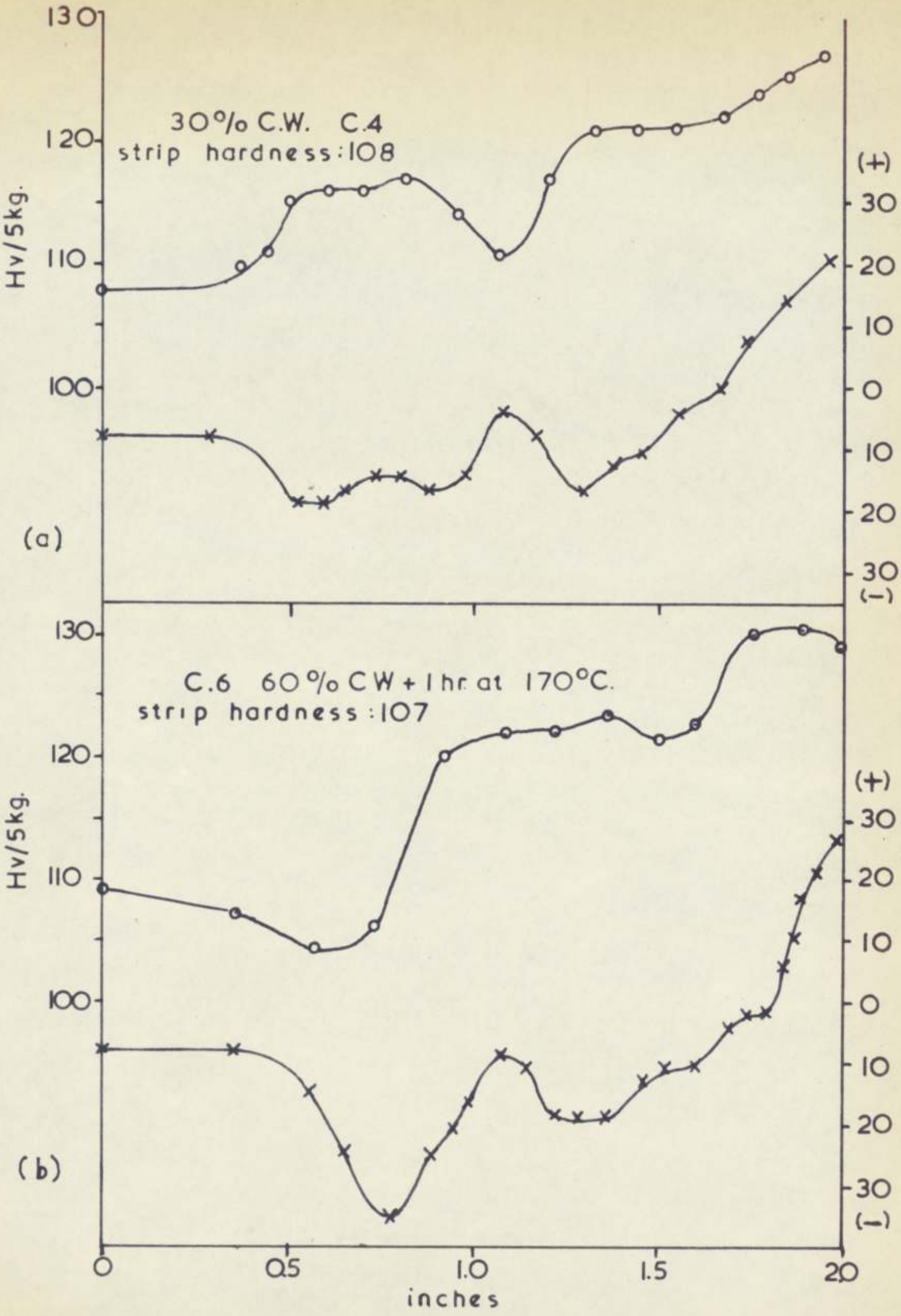


Fig. 8.13 Hardness and thickness strain against distance from cup centre for copper.

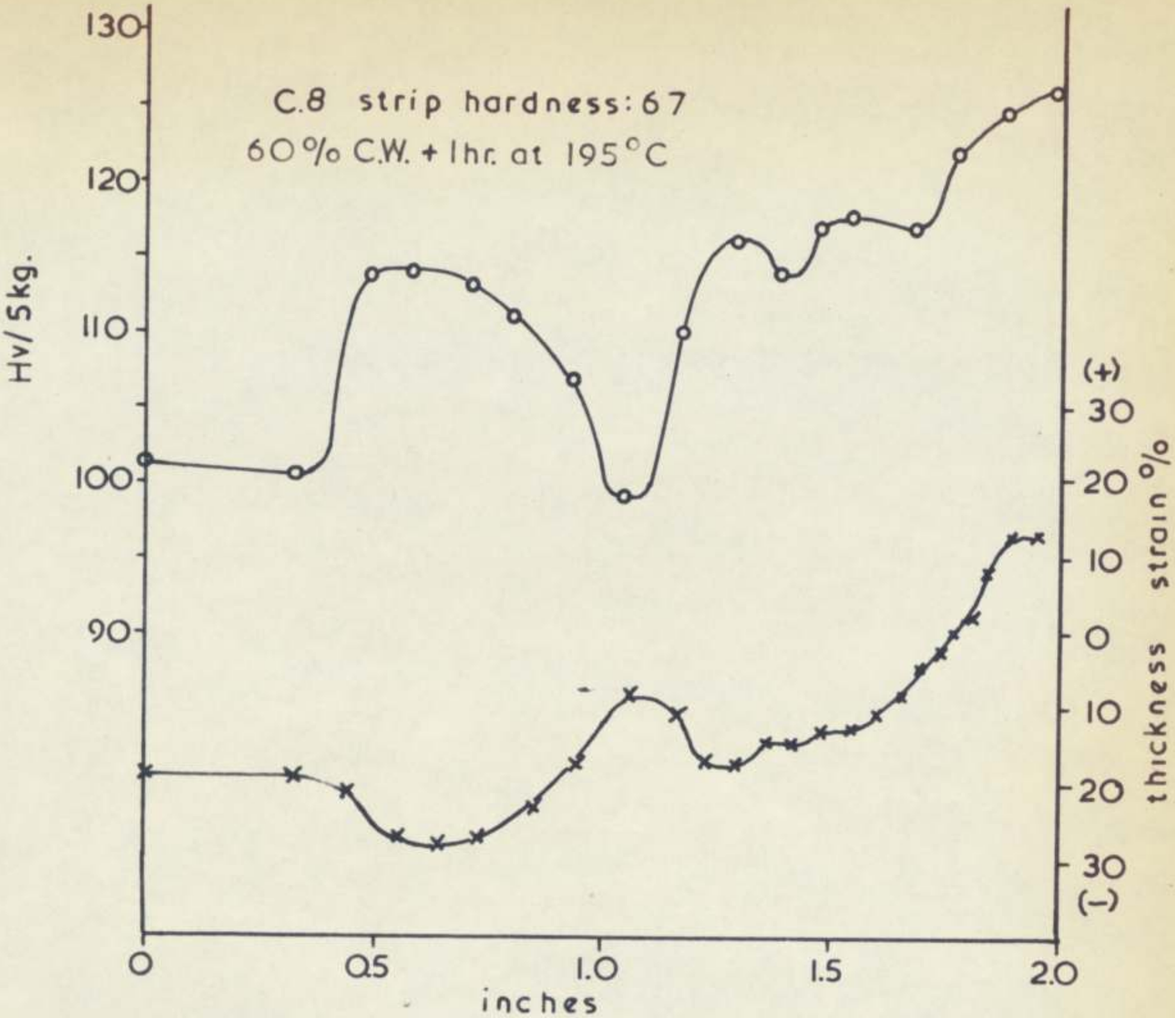


Fig. 8.14 Hardness and thickness strain against distance from cup centre for copper

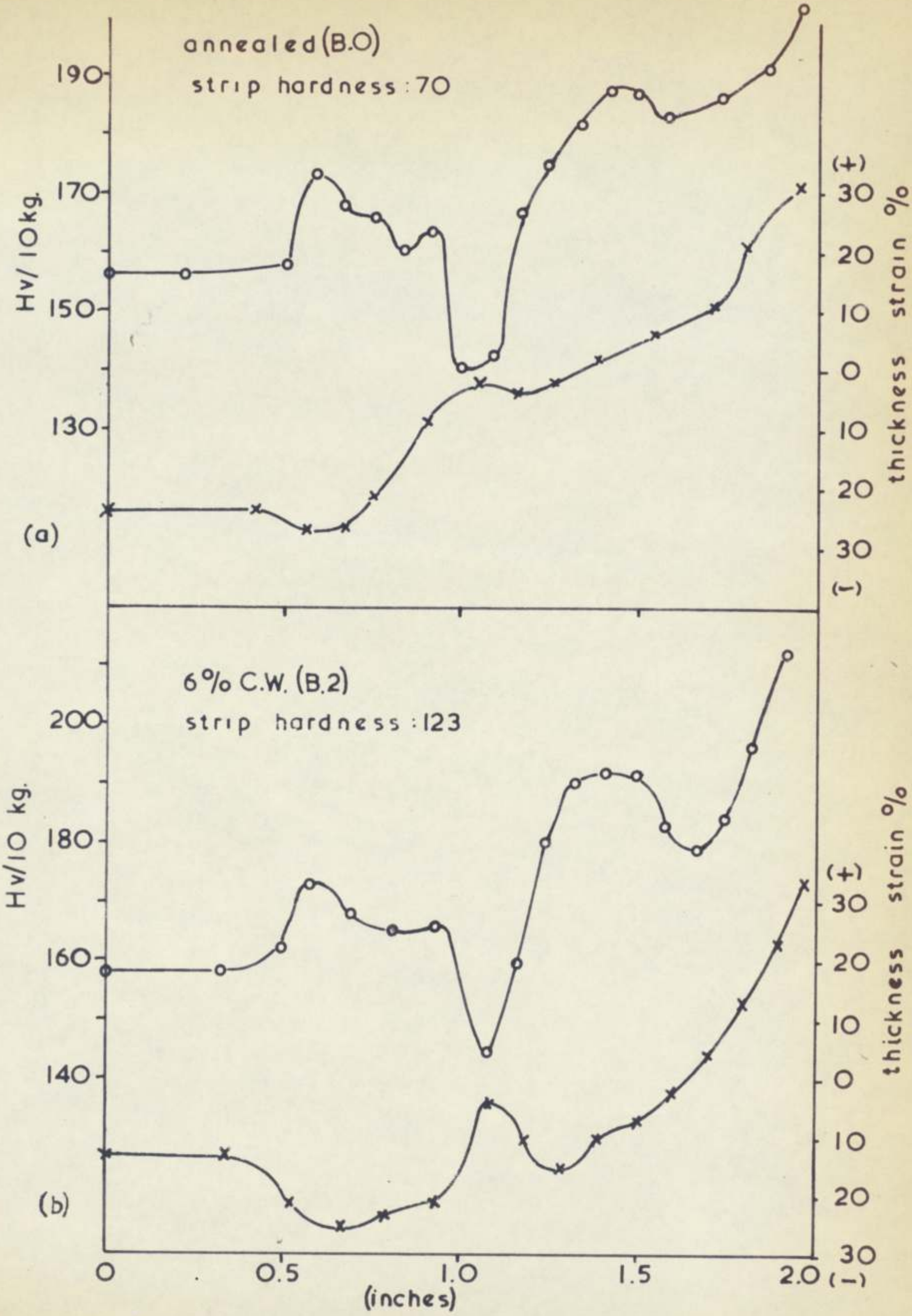


Fig. 8.15 Hardness and thickness strain against distance from cup centre for 70/30 brass

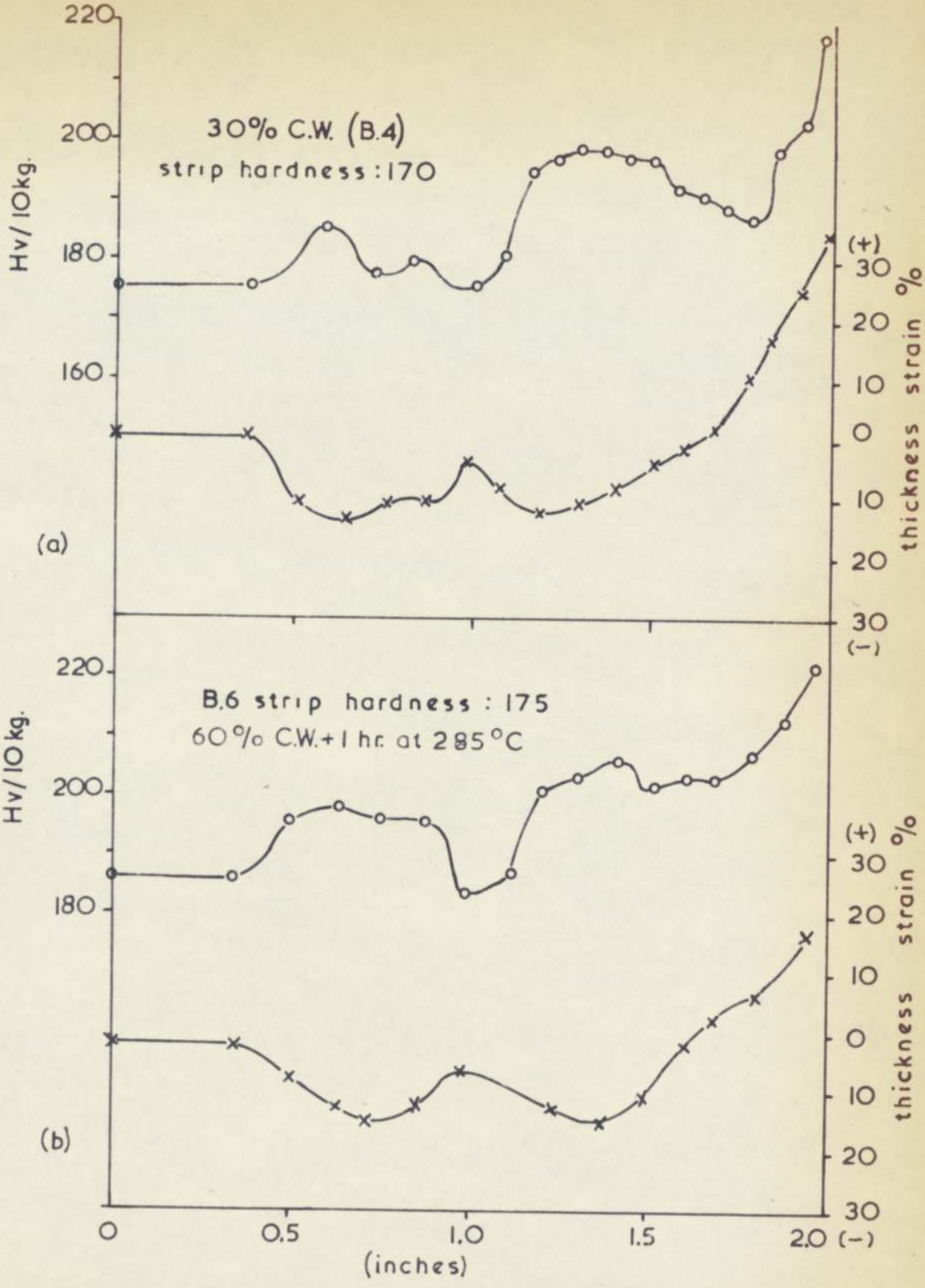


Fig. 8.16 Hardness and thickness strain against distance from cup centre for 70/30 brass.

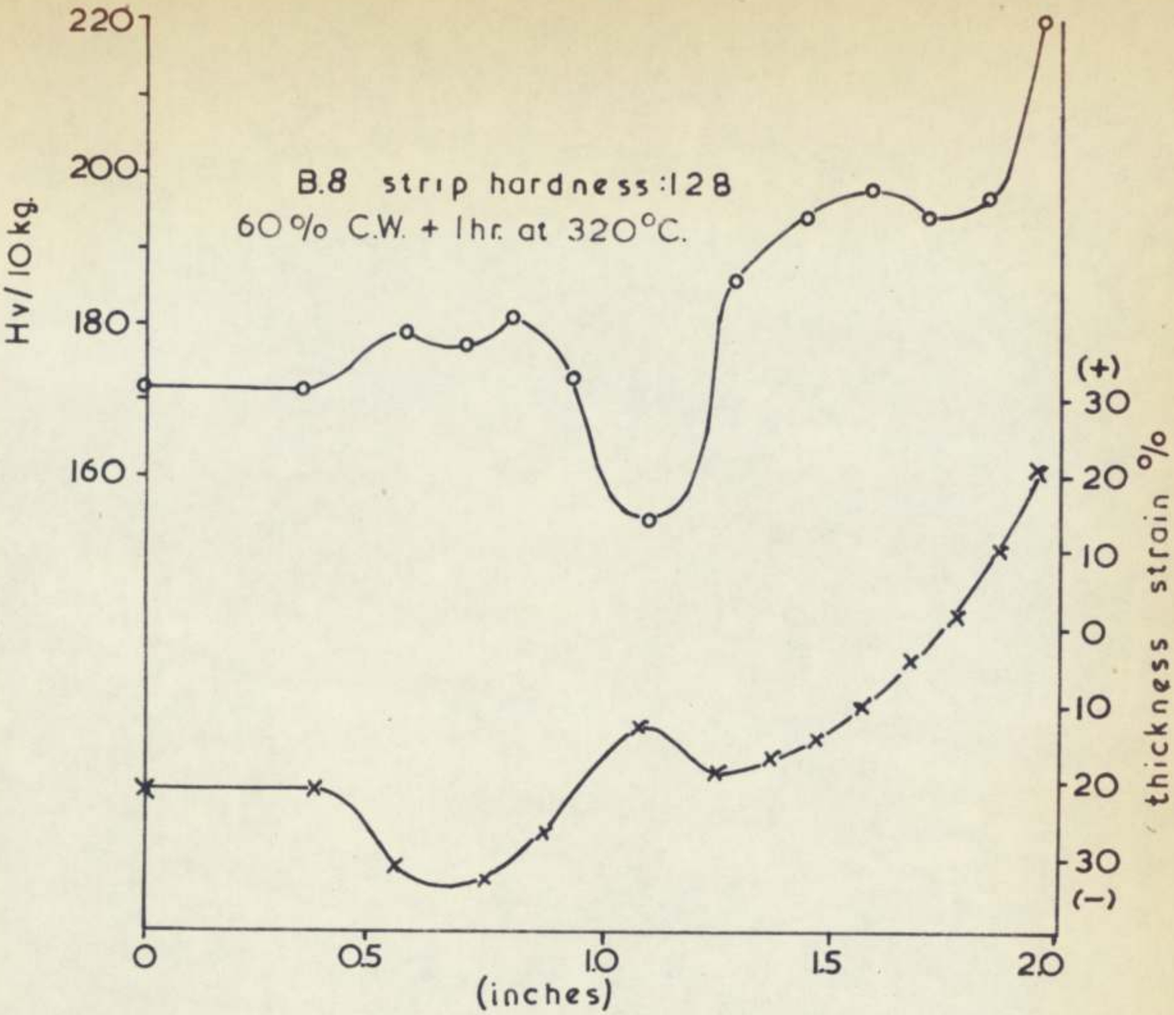


Fig. 8.17 Hardness and thickness strain against distance from cup centre for 70/30 brass.

the hoop compressive stress.

The theoretical thickness strain due to plastic bending under tension over the die profile radius has been shown by Chung and Swift[20] to be of the order of 3 to 4%. If a similar figure is assumed for the thinning due to plastic bending under tension over the punch profile radius, it would appear that most of the thinning observed in this investigation was due to the stretch forming component, since the maximum thinning was as high as 30% in some cases. However, since the analysis of Chung and Swift was specifically for bending over the die entry radius, the adoption of the same analysis for conditions over the punch nose radius is probably an over simplification. In fact, Warwick and Alexander[25] used the analysis of Chung and Swift to predict thickness strains over the punch nose radius for a number of materials, but concluded that the analysis was inadequate because it did not take into account the thinning due to stretch forming. Even a moderately successful analysis of conditions over the punch nose radius has not been achieved, so that it is not yet possible to calculate thickness strains in this region with any accuracy.

Alexander[63] has shown by a simplified analysis that the amount of thinning due to plastic bending under tension is directly proportional to the mean tensile stress and the initial sheet thickness, and inversely proportional to the yield stress and the die or punch profile radius. In the present work, the sheet thickness and tool radii were constant, so that the degree of thinning due to bending over the punch profile radius should decrease with increase in yield stress or with prior work hardening of the blank. This trend was observed in the present investigation, since in annealed copper the thinning exceeded 30% over the punch nose radius, but was somewhat less than 20% in copper strip cold rolled 6%. This was also true of the 70/30 brass. However, since the Erichsen stretch forming value of both copper and 70/30 brass decreased with increase in cold work, it is not clear whether the decrease in thinning over the punch nose radius in cups drawn from cold worked blanks is a result of a decrease in thinning due to bending or stretch forming. It is more likely that the decrease in thinning is due to a decreased contribution from both modes of deformation.

The amount of thinning observed on the base of the cups due to biaxial tension decreased with increase in the yield strength of the initial blank. For example, the base of the cups deep drawn from annealed copper had thinned about 20%, whereas after 60% cold rolling, the cup base thickness was the same as that of the initial strip. A similar trend was observed in 70/30 brass, and in fact the thinning in the base of the cups drawn from annealed material was about 25% (Fig. 8.15a), which is not much less than the thinning over the punch nose radius in region A. The greater thinning of the annealed 70/30 brass compared with annealed copper is mainly attributed to the superior stretch formability of the former. It is also noticeable that zone B is less pronounced in cups drawn from the annealed material compared with cups drawn from temper rolled or temper annealed materials, which all showed a distinct thinner zone at B, This may be due to the more uniform thinning observed in annealed material resulting in more diffuse necking, than in cold rolled material.

There were no systematic differences between the thickness strains in the walls of cups deep drawn from copper and 70/30 brass. This is consistent with the observations of Chung and Swift. The thickness strain progressively decreased to zero with distance from the neck B, and then positively increased to the cup rim due to the compressive stresses under the blankholder. The maximum thinning in the cup wall always occurred at or near the neck B. Since thinning in the cup wall is due to plastically bending and unbending under tension around the die profile radius, it might be expected from Alexander's analysis described earlier that the thinning would be greater in softer strips. In fact, the reverse trend was observed, thinning at neck B tending to be more distinct in the harder materials. An explanation of this in terms of the more diffuse necking of the softer materials has been discussed.

The hardness surveys on sectioned cups of copper and 70/30 brass showed a trend which generally was the reverse of the trend in thickness strain. In all cases there was an increase in hardness over that of the initial strip, but the increase was substantially greater in cups drawn from soft than from hard blanks. In all cups, the increase in hardness was most marked in zone A where stretch forming and bending over the punch nose radius had occurred. Similarly, the hardness of the lightly worked zone C was generally below that of the rest of the cup, especially in annealed and lightly cold rolled strips because of the greater increase in hardness of cups drawn from soft tempers compared with hard tempers. However failure in deep drawing never occurred in the lightly worked zone, presumably because its greater cross-sectional-area could better support the drawing load than the stretch formed zone over the punch nose radius.

An interesting feature which was observed regularly was the slight drop in hardness at a region higher in the cup wall. No explanation can be offered for this phenomenon.

Generally, the results of hardness and thickness strain surveys on deep drawn cups are similar to those presented by Chung and Swift. The surveys have successfully demonstrated the greater overall work hardening and degree of thinning which occurs during drawing a soft blank compared with a hard blank. It was not possible to determine whether thinning due to bending under tension or due to stretch forming was the more important in the region adjacent to the punch nose radius, or whether the relative contribution to thinning changes with blank temper. There was little difference in thickness strain distributions between copper and 70/30 brass, and their hardness surveys differed in magnitude rather than in character due to the greater work hardening of the 70/30 brass.

8.4 Direct Redrawing Behaviour of Copper and 70/30 Brass

The double action form of the direct redrawing process was used, as distinct from the single action type, and the ironing type, neither of which employ a saddle. Initially, the punch advances into the base of the cup and subsequently enters the die with the result that the diameter of the cup is reduced. After this initial deformation over the punch nose, the remainder of the cup is progressively reduced in diameter by plastic deformation of that material which is negotiating the 'S' bend around the saddle and the die, so that little simultaneous deformation occurs in the redrawn portion of the cup or in the remainder of the wall of the first stage cup, Fig. 8.19.

Thus at any stage in the redrawing process, only a narrow zone of material is being deformed, in contrast to cup drawing in which at any stage the whole flange is being deformed. The redrawing load is thus dependent on the flow stress and the volume of material passing

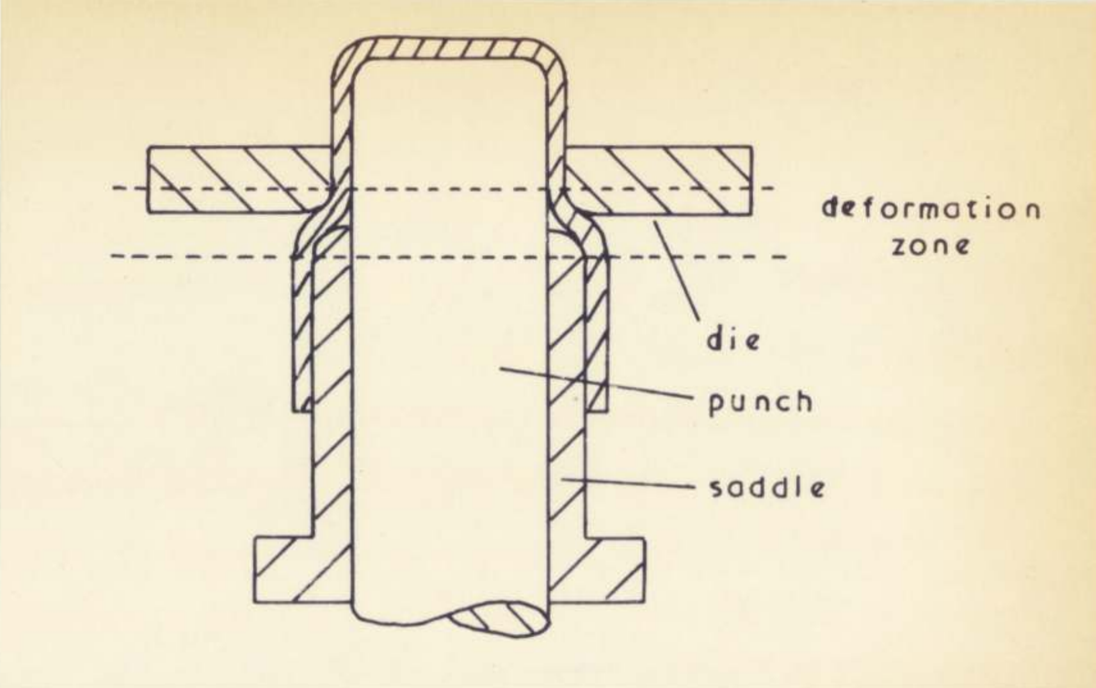


Fig. 8.19

through the deformation zone at any instant, rather than on the properties of the whole flange as in cup drawing. The redrawing load will be affected by hardness and thickness strain distribution in the cup. In cups drawn in the Erichsen test there was always a positive hardness gradient and an increase in thickness from the base to the rim of the cup. Typical hardness and thickness strain distributions for Erichsen cups drawn from copper and 70/30 brass in various conditions are shown in Figs. 8.12 to 8.17. Both hardness gradient and thickening increase the redrawing load.

The results of the redrawing tests showed that the redrawing performance of copper from Erichsen cups of the C.B.D. was vastly superior to that of 70/30 brass. For copper, the redrawing capacity increased with increasing prior cold rolling reductions up to 60%. A similar behaviour was observed in temper annealed copper, in which the harder sheets after temper annealing possessed superior redrawing properties to those sheets which had been annealed to a softer temper. None of the first stage cups of 70/30 brass would redraw without an interstage anneal.

Since only a limited size range of redrawing tools was available, the results of the redrawing tests are only semi-quantitative and so it was not possible to calculate the limiting redrawing ratio for all materials. If larger punch and die sizes had been available, it is possible that the 70/30 brass would have redrawn. Conversely, if smaller tool sizes had been available, it is equally possible that copper sheets in the harder tempers would have limiting redraw ratios larger than the maximum ratios used in this investigation.

In both copper and 70/30 brass failure usually occurred after 70 to 80% of the redrawing operation had been performed, irrespective of the initial sheet temper.

Failure during redrawing copper cups generally occurred over the punch nose radius in a similar manner to that failure most frequently observed in first stage cup drawing. The poor redrawing capacity of cups drawn from the softer copper strips was due to a progressive increase in flow stress during the redrawing operation, since these cups had some capacity for further strain hardening, as can be seen from their hardness gradients, whilst the base of the first stage cup was both thinner and softer than cups drawn from harder tempers, and therefore less able to support the increased redrawing load.

Occasionally when redrawing copper cups from soft strips, failure occurred at the region of greatest thinning in the first stage cup, that is the region adjacent to the punch profile radius, so that on the partially redrawn cup the failure site was in the wall some way from the base. However this was the exception rather than the rule. This suggests that a substantial amount of thinning due to stretch forming and bending under tension occurred during redrawing to further thin the base of the first stage cup to such an extent that the failure site was at the nose radius of the redrawing punch rather than at the thin zone of the first stage cup.

The redrawing capacity of copper increased with increasing rolling reduction because there was less thinning in the first stage cup drawn from a hard blank, so that the material was stronger and better able to support the redrawing load. In addition, the hardness gradient in the walls of cups drawn from harder tempers was considerably smaller than in cups drawn from soft tempers, so that the flow stress would increase only slightly during the redrawing operation. It is suggested that a limit to the amount of cold work which would improve redrawability is reached when the decrease in ductility with work hardening is so great that the material is unable to negotiate the bends involved without cracking.

None of the cups drawn from 70/30 brass would successfully redraw in any condition. However, the failure was different from that observed in copper cups, in that it occurred away from the punch profile radius and was in most cases situated within the 'S' bend in the material. Fracture in these cases was brittle, no necking being observed and was probably associated with a general lack of bend ductility due to the high work hardening rate of 70/30 brass.

One of the objects of the Erichsen cup redrawing test is to differentiate which, of strips showing the same C.B.D. in the cup drawing testhas the greater deep drawability. The results of redrawability tests on copper have shown that the test is not really suitable for differentiating between cups drawn from strips of different temper. The test is useful in providing an assessment of the suitability of strip for redrawing operations, but the results are not directly relevant to deep drawing behaviour. When copper strips rolled to temper had the same C.B.D. as strip annealed to the same temper, there was no difference in redrawing performance.

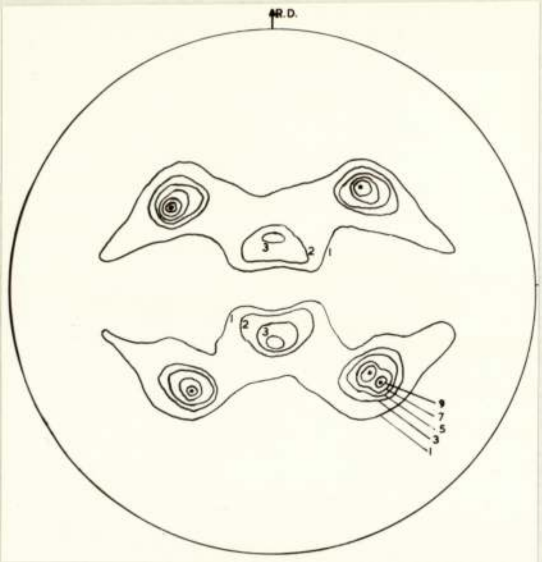
Generally, the best performance in redrawing will be given by the most heavily cold worked material, provided that the first stage cup possesses sufficient ductility to negotiate the bends involved in the redrawing operation.

8.5 Textures Developed in Rolling and Annealing of Copper and 70/30 Brass

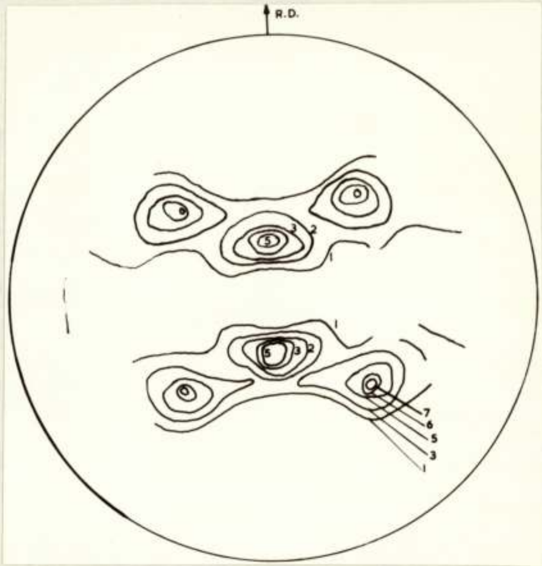
Dillamore and Roberts[58] have considered the development of rolling textures in F.C.C. metals and alloys and have successfully accounted for the different rolling textures observed in aluminium and silver on the basis of the difference in stacking fault energy, and hence on the different modes of deformation which occur during rolling.

The rolling textures observed in copper and 70/30 brass in this work followed similar patterns of development to those observed in aluminium and silver respectively, in that progressive cold rolling resulted in the development of the 'pure metal' type texture in copper, and of the 'alloy' type texture in 70/30 brass.

Fig. 8.20(a) to (c) shows centre pole figures for annealed copper, and for copper after cold reductions of 15% and 60%. Pole figures were obtained for strips after rolling reductions of 3%, 6% and 30%, and conformed with the general pattern of texture development indicated. In the annealed condition, the texture of copper consisted of a mixture of cube texture and rolling texture, the former being present with greater



(a)



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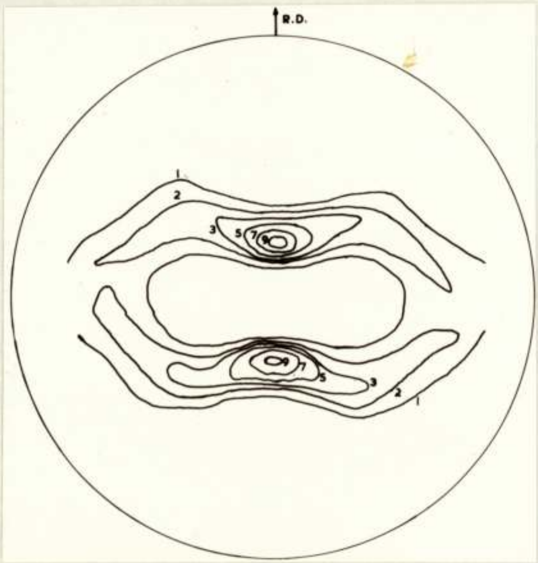
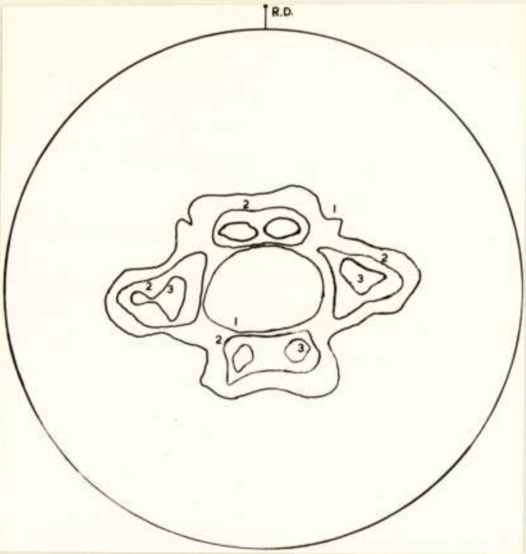
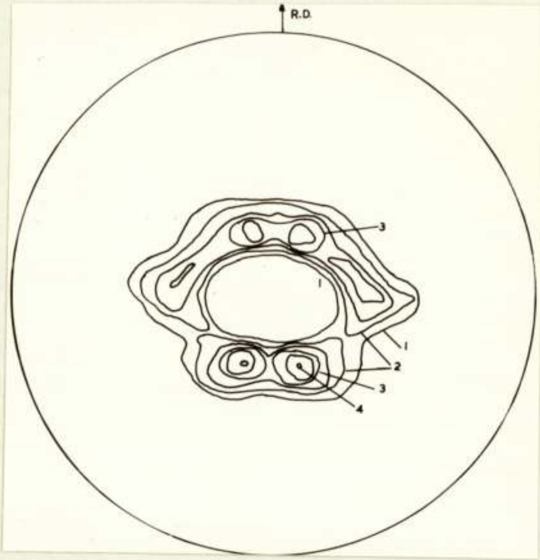


(c)

Fig. 8.20 {111} pole figures for copper
(a) annealed C0
(b) cold rolled 15% C3
(c) cold rolled 60% C5



(a)



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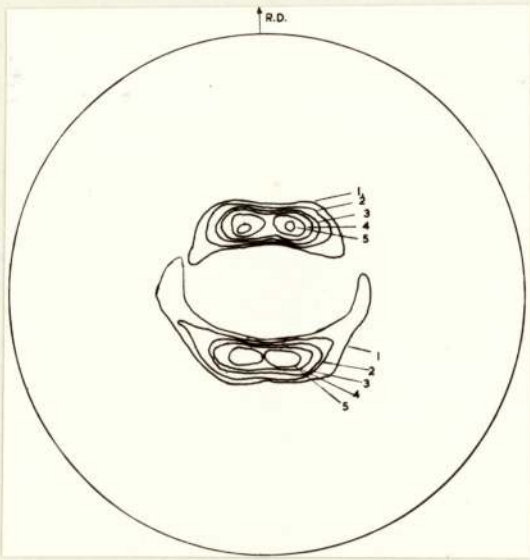
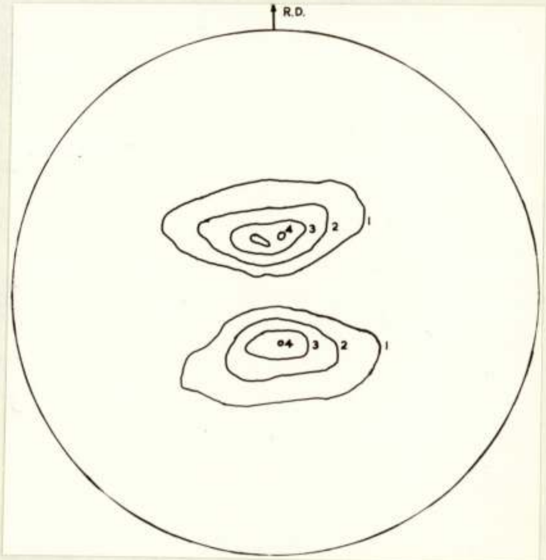


(c)



(d)

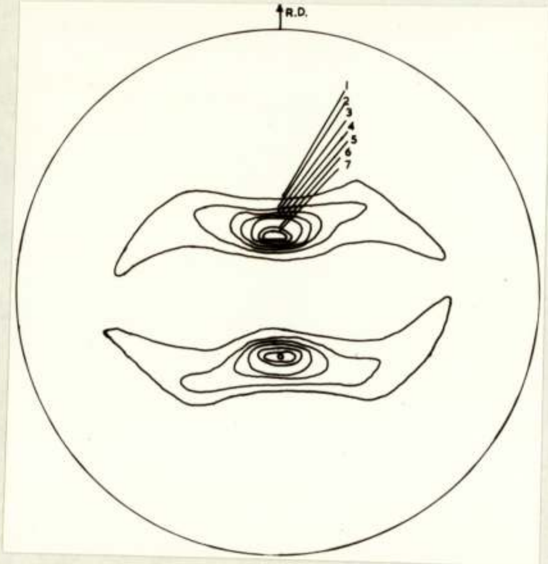
Fig. 8.21 {111} pole figures for 70/30 brass

(a) annealed B0
(b) cold rolled 6% B2
(c) " " 15% B3
(d) " " 60% B5

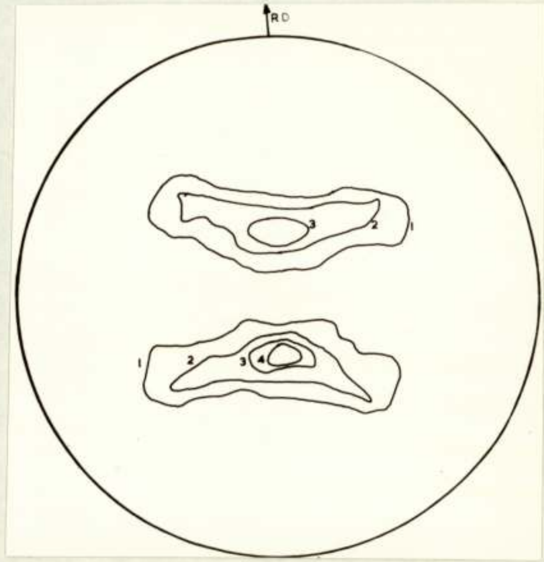
intensity than the latter. With increasing rolling reductions the cube texture component decreased in intensity, whilst the rolling texture component became more intense. After 30% rolling reduction, cube texture peaks were still present at small intensity levels, but they were not evident after 60% rolling reduction, the pole figure being similar to that of the "pure metal" texture described by Dillamore and Roberts. The maximum rolling reduction examined was not sufficient to cause extreme amounts of cross slip, so that the end orientation proposed by Dillamore and Roberts, $\{112\}\langle 111\rangle$, was not approached.

The $\{111\}$ pole figures of 70/30 brass in the annealed condition and for strips after cold reductions of 6%, 15% and 60% are shown in Fig. 8.21(a) to (d). The texture of the annealed material is similar to that reported by Roberts[4], and has been described by the orientation $\{113\}\langle 211\rangle$. Insufficient rotation had occurred after 60% rolling reduction to approach the end orientation $\{110\}\langle 112\rangle$.

The changes in texture on temper annealing copper and 70/30 brass are shown in Figs. 8.22 and 8.23 respectively.



(a)



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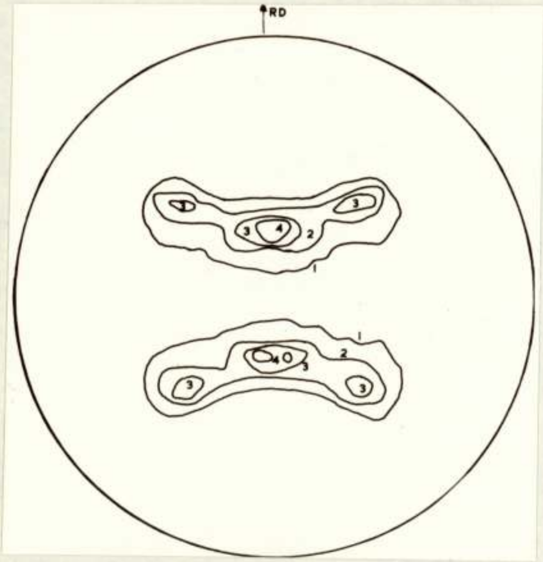

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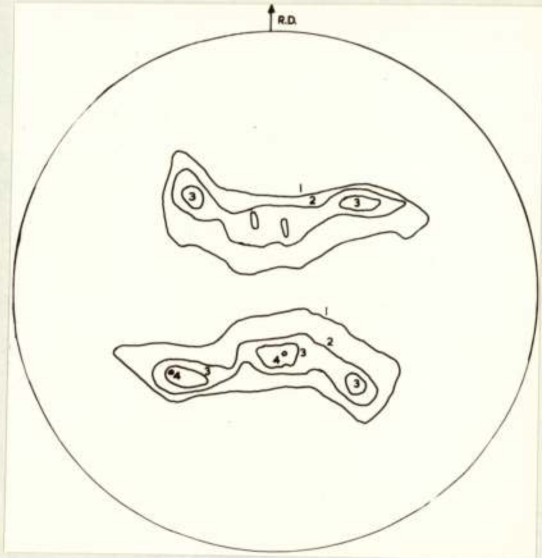
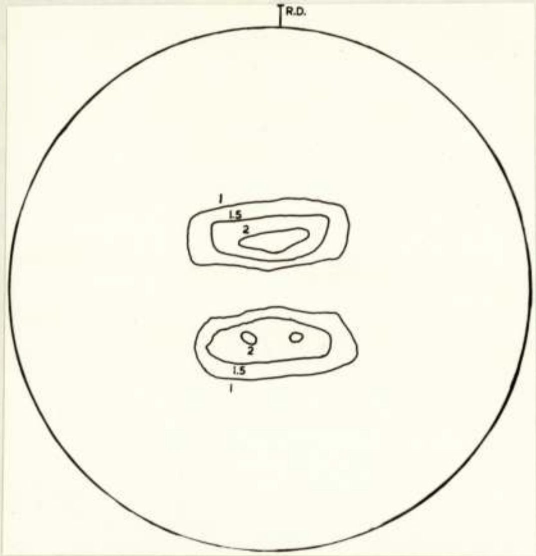
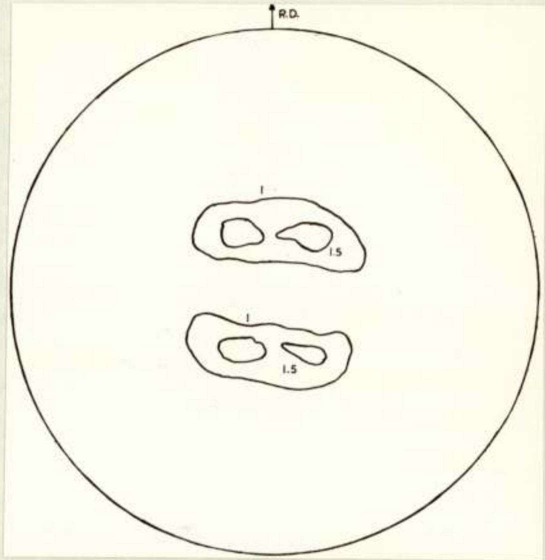


(d)

Fig. 8.22 {111} pole figures for temper annealed copper
(a) C.W. 60% + 1hr at 170°C C6
(b) " 60% + 1hr at 185°C C7
(c) " 60% + 1hr at 195°C C8
(d) " 30% + 1hr at 223°C C10



(a)



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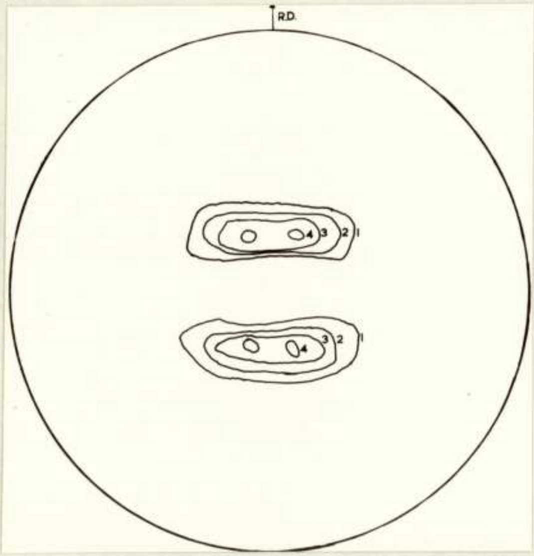
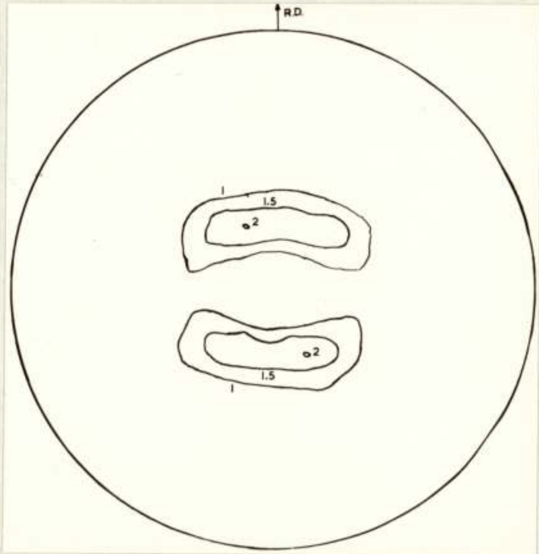


(c)



(d)

Fig. 8.23 {111} pole figures for temper annealing
70/30 brass (a) C.W. 60% + 1hr at 285°C B6
(b) " 60% + 1hr at 295°C B7
(c) " 30% + 1hr at 315°C B9
(d) " 30% + 1hr at 325°C B10

For copper cold rolled 60%, with increasing annealing temperature, there is a gradual decrease in the intensity of the main pole regions on the N - S axis of the pole figures. Similar slight intensity changes were observed after temper annealing treatments on strips cold rolled 15% and 30%, no obvious re-orientation having occurred. Difficulty was found in following the progress of temper annealing treatments from pole figure determinations, and in practice the onset of recrystallisation was much easier to assess from x-ray back reflection photographs. These are shown for a selection of strips which were rolled to temper or annealed to temper in Figs. 8.8 to 8.11.

Temper annealing of cold rolled 70/30 brass again resulted in little re-orientation of the rolling texture components, and in fact there was little difference in texture between any of the temper annealed sheets.

If temper annealing had only caused recovery without any extensive re-orientation of the deformed crystals, one would not expect any major differences between the cold rolling textures and the textures after temper annealing. However, the x-ray back reflection photographs indicate that recrystallisation had begun in many of the temper annealed sheets. Apparently the volume fraction of recrystallised grains was too small to produce any significant effect on the texture after temper annealing.

Summarising, the textures observed in copper and 70/30 brass with increasing rolling reduction are consistent with modern theories of texture development. In copper, sufficient cross slip had occurred after 60% rolling reduction to cause rotation to the "pure metal" texture, whilst the same level of deformation in 70/30 brass resulted in an approach to the "alloy" texture. Temper annealing of cold rolled copper and 70/30 brass produced little re-orientation of the rolling textures even though some recrystallisation had occurred.

8.6 Dependence of Earing on Preferred Orientation

Most of the work on earing in copper and 70/30 brass has been on annealed material with a view to relating the planar variation in mechanical properties to ear formation during deep drawing. Only limited success was obtained by this technique because of the difficulty in accurate measurement of mechanical properties during tensile testing.

More recently, with the introduction of x-ray diffraction counter techniques, attempts have been made to relate the preferred orientation of copper and 70/30 brass from pole figure determinations to the observed earing behaviour.

From theoretical considerations, Tucker[64] has successfully predicted the number, positions and heights of ears formed on cups deep drawn from aluminium single crystal blanks of known crystallographic orientation. He suggests that the method could be extended to polycrystalline materials if the requirements of multiple slip in grain boundary regions are ignored and the texture of the polycrystal is intense so that it can be described in terms of ideal orientations.

Van Horn[42] proposed a graphical method of predicting ear position and relative ear height in aluminium from the type of $\{111\}$ texture observed, and the relative intensities of the components in the texture. Roberts[43] modified Van Horn's method to account for the distribution of slip directions as well as the distribution of slip planes, since a consideration of slip planes alone gave unsatisfactory results for more diffuse textures. Roberts has successfully accounted for the positions of ears in cube textured copper, and annealed copper with a texture similar to the rolling texture, using his method.

A comparison of the pole figures and earing results obtained during this work showed that it was not possible to predict all the earing positions observed either by simple inspection or by using simple graphical methods of the type proposed by Van Horn.

For copper, if the cube texture component predominated, ears were observed at 0° and 90° to the rolling direction, and conversely if the rolling texture component predominated, ears were observed at 45° to the rolling direction. Reasonably flat rimmed cups were produced in some cases, presumably when the contributions to earing from the cube texture and rolling texture components were similar.

In the cold rolled specimens, the transition from earing at 0° and 90° to the rolling direction to earing at 45° to the rolling direction was almost complete after 15% rolling reduction, and after 30% rolling reduction ears were observed at 45° to the rolling direction. However the pole figure for copper strip after 30% rolling reduction still contained a proportion of cube texture, so that it is inferred that the contribution of the cube texture component to earing potential is reduced by increasing amounts of cold rolling, until after 30% rolling reduction any contribution from cube texture is not evident on a deep drawn cup. The actual rolling reduction at which this change of ear position occurs will depend on the volume fraction of grains in the cube orientation present in the strip before rolling. These observations are in agreement with those of Baldwin,[65] who demonstrated an approximately linear relationship between the height of 0° and 90° ears and the percentage of grains in the cube orientation.

In copper strips temper annealed after 60% rolling reduction, cube textured peaks were observed after temper annealing at the highest temperature, Fig. 8.22(c). In these strips, there was also a decrease in the height of ears at 45° to the rolling direction with increasing temper annealing temperature, until at the highest temperature the cup rim was almost flat . It seems that the appearance of cube texture in temper annealed samples was associated with the decrease in ear height. This is certainly true of strip C.10, since the pole figure (Fig. 8.22(d)), contained a proportion of cube texture but cups deep drawn from this strip showed no earing.

It is interesting to observe that strip which was cold rolled by 60% and then temper annealed at the lowest temperature (C.6) developed larger ears than strip which was cold rolled to a similar temper (C.4), even though a comparison of the pole figures indicates a reduction in the intensity of the rolling texture after temper annealing. However this observation is not too surprising since the pole figure of the temper rolled strip C.4 contained some cube texture, but that of the temper annealed strip C.6 did not. The presence of the cube texture component would inhibit the development of high 45° ears in the temper rolled strip.

The earing behaviour of the 70/30 brass strip was generally similar to that of copper strip whose texture is predominantly a rolling texture. Ears were observed at 45° to the rolling direction in almost every strip and the percentage earing was generally smaller than in copper. The annealed strip differed from the rest in that it tended to develop six ears, two at 0° and four at about 60° to the rolling direction. The two ears in the rolling direction were suppressed in that they were lower with respect to the cup base than the troughs between the 60° ears.

The observation of ears at 0° and 90° to the rolling direction in 70/30 brass is uncommon, since this type of earing is associated with the "cube" recrystallisation texture. Cube texture is sometimes obtained after annealing of F.C.C. metals, and has been related by a simple lattice rotation to the "pure metal" type rolling texture. It is unlikely that cube texture would be produced after annealing a material which only develops the "alloy" type texture after cold rolling, and so ear formation at 0° and 90° to the rolling direction in 70/30 brass is unusual. However 70/30 brass cups with six ears have been observed previously[66].

After small rolling reductions, the positions of the ears changed from 0° and 60° to 45° to the rolling direction. With increasing rolling reductions, the heights of the 45° ears observed on 70/30 brass cups increased with the increase in intensity of the rolling texture, although even after 60% rolling reduction the maximum percentage earing observed was only 3.28%. In contrast, the ears on cups deep drawn from strips which had been annealed to temper were generally slightly higher than on cups deep drawn from strips rolled to temper. Since the intensity of the textures was generally lower after temper annealing, this effect is surprising, but since a similar effect was found in copper, it is considered to be a genuine one.

No satisfactory method has yet been proposed to predict the pattern of earing of polycrystalline strip from a knowledge of the strip texture. In particular, when the texture is diffuse, as is generally the case in commercial strip and which is also true of the majority of strip investigated in this work, the methods available for predicting the earing behaviour are inadequate usually due to over simplifications in describing the rolling and annealing textures. In any case, from the point of view of quality control on commercial strip it would appear that the most rapid and reliable method of assessing earing behaviour is by means of a cup drawing test.

Some success has been achieved by attempts to correlate earing behaviour with sheet anisotropy as measured by the R value. Baldwin, Howald and Ross[67] related the heights of ears measured on cups drawn from copper sheets containing varying proportions of cube texture with the planar variations of the R value.

They observed that earing occurred in the radial directions of maximum R value and showed that the ear height was approximately proportional to $R_{0,90} - R_{45}$.

However, Wilson and Butler[36] found that ears developed at positions where the R value was a maximum for uniaxial strain in the circumferential rather than the radial direction. This enabled them to fully relate ear position and height to the planar variation in R value when the earing had a two-fold symmetry. Further, by using the proportionate variation in R value, $\frac{(R_{0,90} - R_{45})}{\overline{R}}$, instead of the simple difference in R values, they demonstrated a linear relationship with percentage ear height which included results from copper, steel and aluminium.

The proportionate variation in R value is shown plotted against percentage earing for the copper and 70/30 brass used in this investigation in Fig. 8.24. The general trend is similar to that observed by previous workers, although there is some scatter in the results particularly those of the heavily cold worked material, which is probably due to errors in the measurement of the R values of these strips. In addition some scatter will

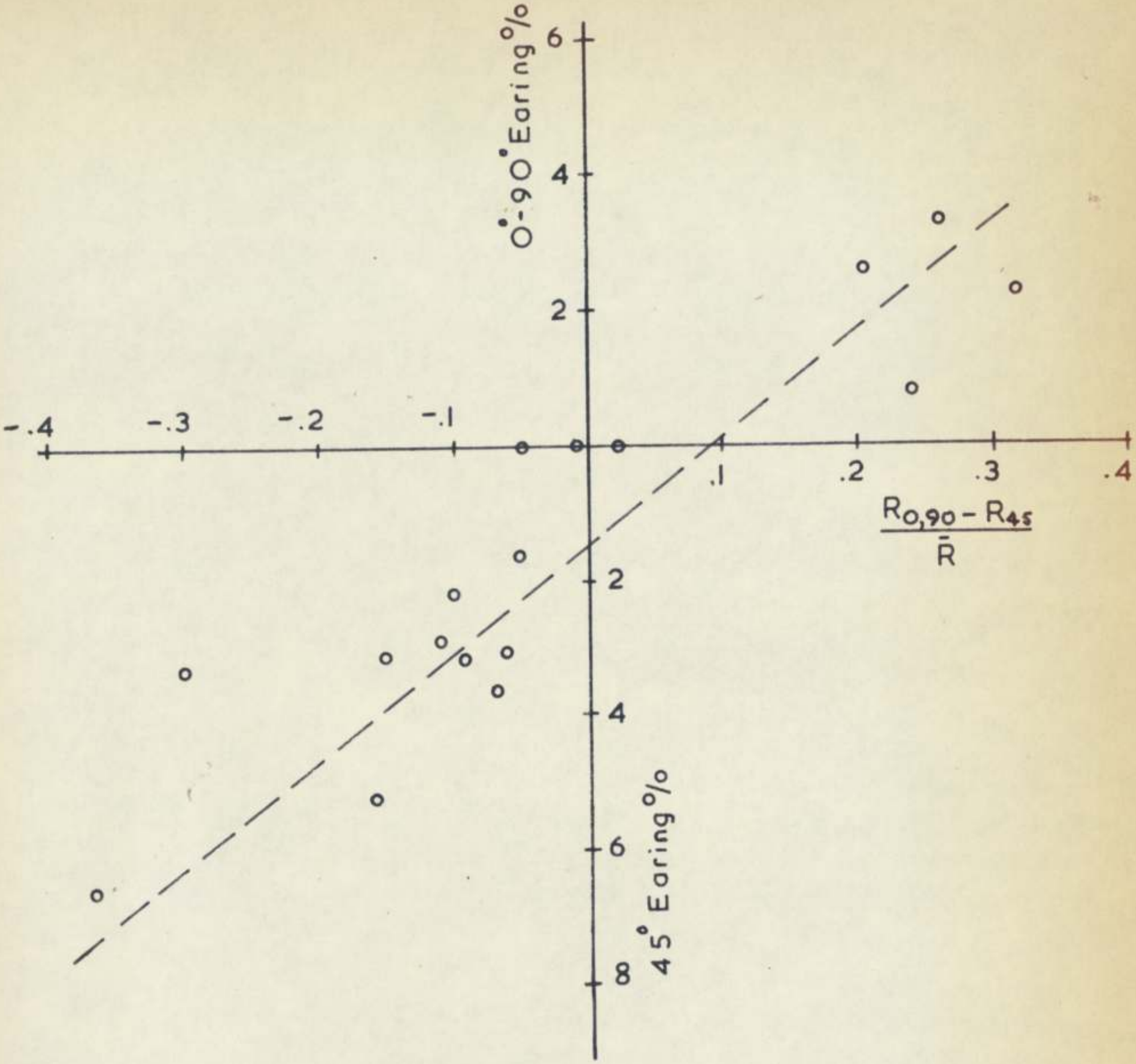


Fig. 8.24 Proportionate variation in R value against % earing for copper and 70/30 brass in all conditions.

arise in the measurement of percentage earing, since the difference in height between ears and troughs was generally only of the order of 1 mm. Nevertheless, it is clear that the heights and positions of ears on copper and 70/30 brass cups can be predicted fairly accurately from a knowledge of the planar variation in normal anisotropy.

In contrast to the observations of Wilson and Butler[36], and Wright[68] it was found that the heights of ears and troughs could apparently only be related to planar variations in R value if the R value is considered in a radial direction in the blank. Some examples of the relationship between cup height and planar variations in R value are shown in Figs. 8.25 to 8.27. These results indicate that a fairly consistent relationship is obtained between R value and ear height when the R value is plotted in a radial direction.

Wright,[68] in considering the phenomenon of earing in deep drawing, examined the earing behaviour of aluminium, zinc, copper and mild steel strip in relation to the planar variation in R value. He showed that earing occurred in all these materials at positions where the anisotropy ratio is a maximum in the

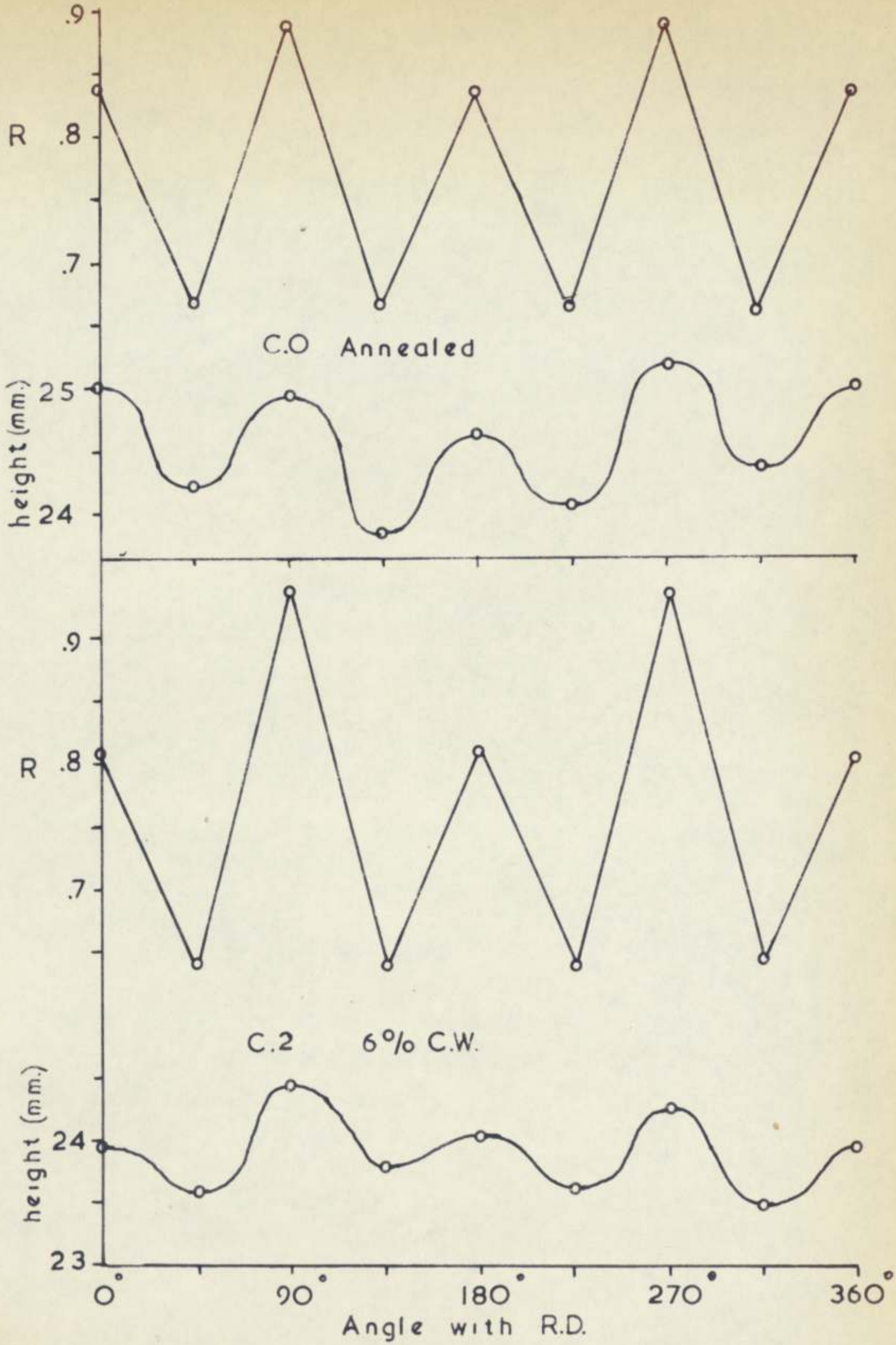


Fig. 8.25 R value and cup height profiles for copper.

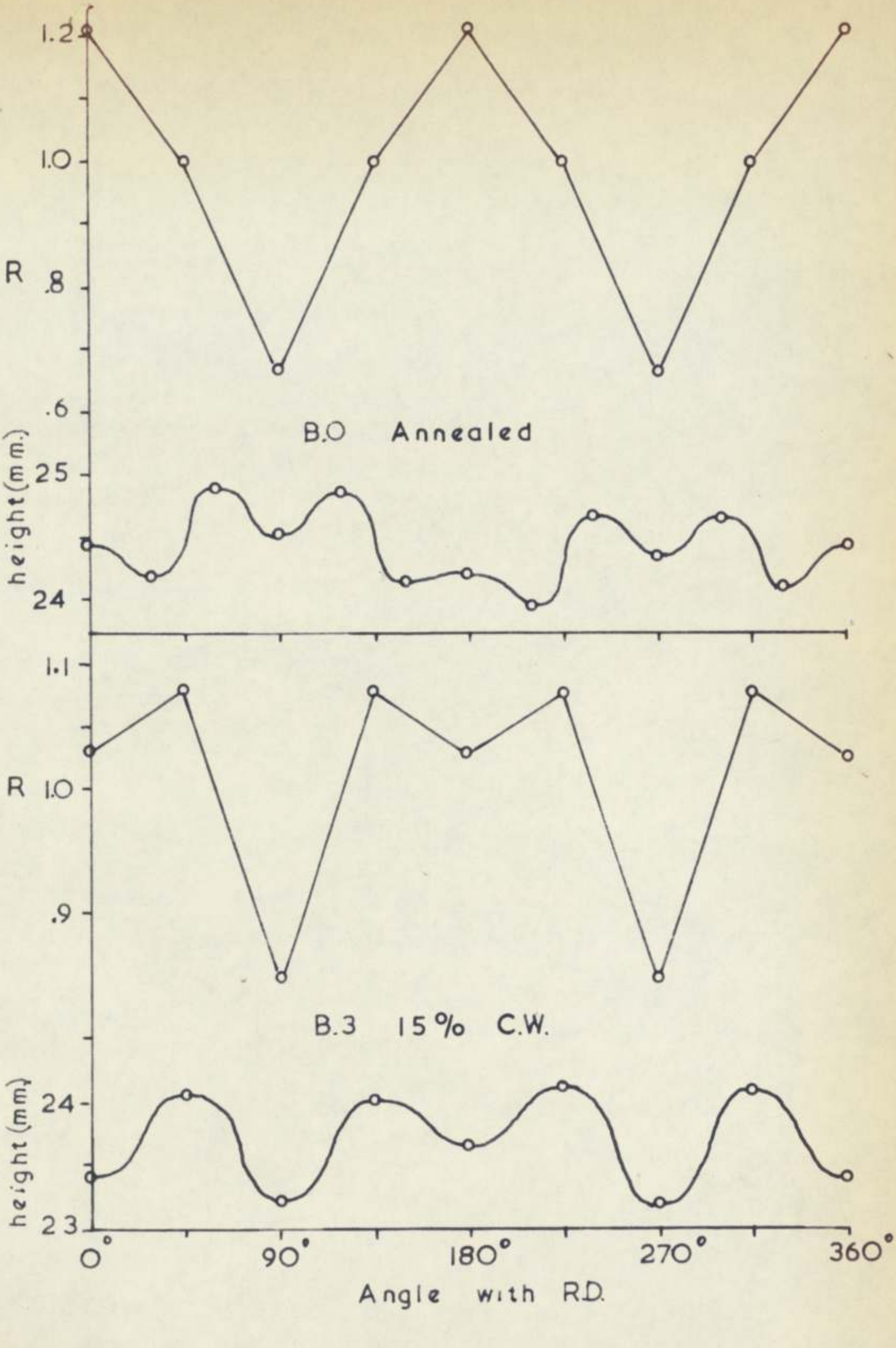


Fig. 8.26 R value and cup height profiles for 70/30 brass.

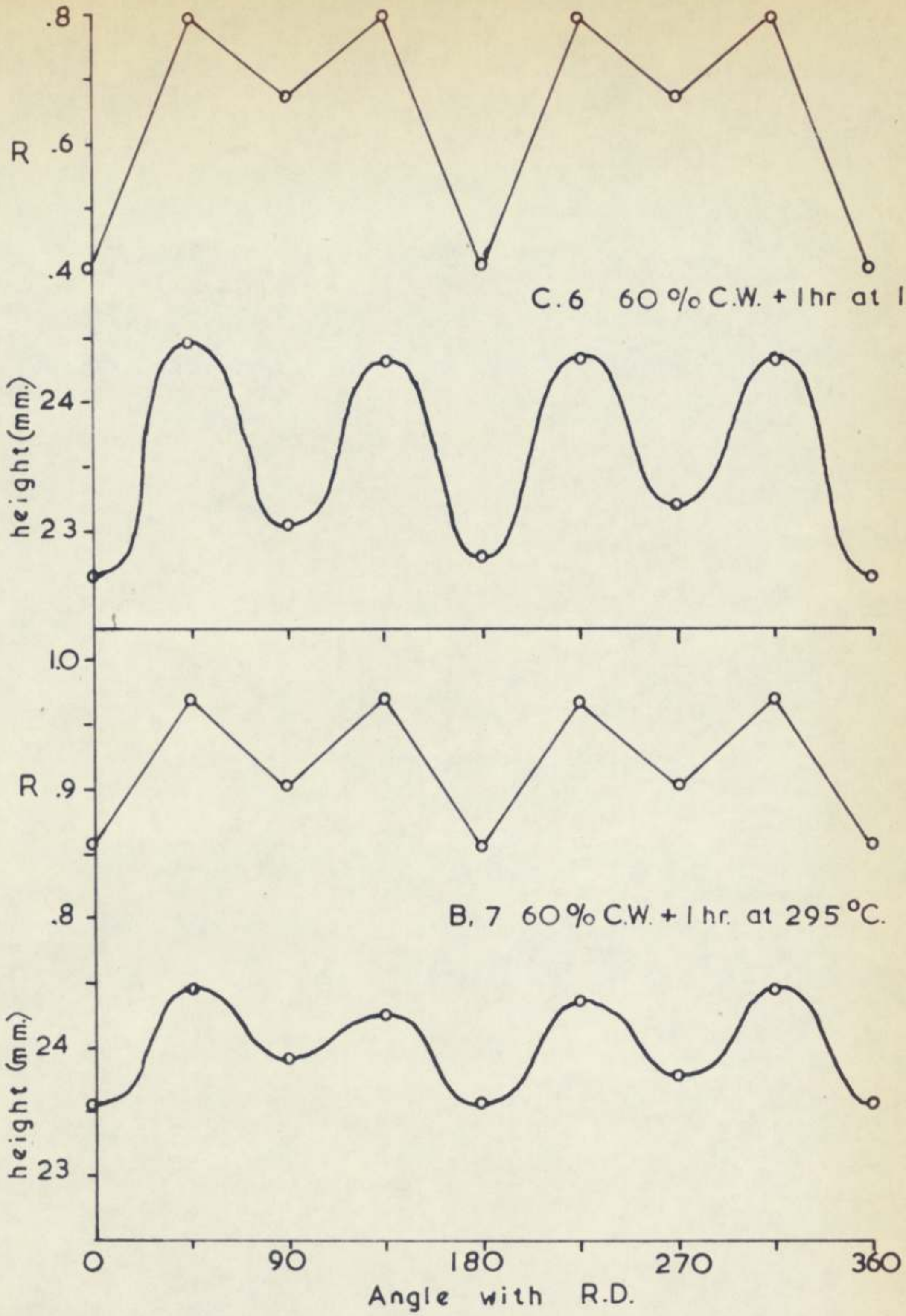


Fig. 8.27 R value and cup height profiles for copper and 70/30 brass

circumferential direction. He further showed that in aluminium the higher pair of troughs were associated with the "higher minimum" R value.

Theoretically, since the dominant stress in the outer part of the blank during deep drawing is circumferential compression, it might be expected that the maximum values of the ratio radial strain:thickness strain, which would allow the development of ears, would occur where R is a maximum for uniaxial strain in the circumferential direction.

The pattern of behaviour observed in this investigation disagrees with those observed by previous workers and with theoretical predictions. However it must be emphasised that the difference in height between ears and troughs was generally less than 1 mm. Consequently it is impossible to be conclusive as to whether the earing profiles best correlate with R value in the radial or circumferential direction. On the other hand, the correlation between earing profile and R value plotted radially is quite marked in certain cases, for example strip C.6, Fig. 8.27, but generally if the R values at 0° and 90° to the rolling direction are transposed the relationship between ear position and maximum R value still holds, although variations in ear and trough heights cannot be accounted for so well in some cases. However it is considered unrealistic to attempt to account precisely for these minor variations when the difference between ear and trough heights is so small.

The earing profile of annealed 70/30 brass, Fig. 8.26, shows a tendency to form six ears at 0^{o} and 60^{o} to the rolling direction. Since R values were only measured at 0^{o}, 45^{o} and 90^{o} to the rolling direction it is not possible to quantitatively relate R value and earing characteristics in this case.

Summarising, the earing position and percentage earing observed in cups deep drawn from copper and 70/30 brass in all conditions have been related to the proportionate variation in R value. With the exception of annealed 70/30 brass, the earing position and relative ear and trough heights appeared to be more accurately related to planar variations in R value when the R value was considered in the radial rather than the circumferential direction, but the evidence is inconclusive because the maximum variation in cup height was small.

9. CONCLUSIONS

1. The stretch forming capacity of both copper and 70/30 brass tested in all conditions correlates best with tensile elongation, but correlates highly with U.T.S. and 'n' values. In addition, there is an almost linear relationship between Erichsen stretch forming value and Vickers hardness for both materials.

2. The best stretch formability is given by fully softened material, but for harder strips there is little difference in performance between material rolled directly to the required temper and material annealed to temper from a greater hardness.

3. There is a high correlation between R value and deep drawability when both materials in all conditions are considered together. When copper and 70/30 brass are considered individually, there is a weak relationship between R value and deep drawing capacity for copper, but the deep drawing capacity of 70/30 brass is independent of R value.

4. No simple correlation exists between deep drawing capacity and such properties as strength or elongation if the materials are considered in

a wide range of conditions.

5. The deep drawing capacity of copper is a maximum after small rolling reductions, whilst that of 70/30 brass is a maximum in the annealed condition; however, the deep drawability of 70/30 brass is only slightly less than the maximum after 30% rolling reduction.

6. Annealing to temper nearly always increases the deep drawing capacity of copper and 70/30 brass compared with strip rolled to a similar temper. For copper, the increase is greatest when temper annealing creates a recovered state, but for 70/30 brass temper annealing always appears to cause partial recrystallisation.

7. Within the limits of tooling and lubrication used, there is little difference in hardness and thickness strain distributions in cups deep drawn from copper and 70/30 brass strip of similar temper. For both materials, the degree of thinning over the punch nose radius generally decreases with increase in hardness of the blank.

8. Within the limits set by the redrawing tools and rolling reductions used, the direct redrawing

capacity of copper increases with strength, or cold work, provided sufficient ductility is retained to negotiate the bends; 70/30 brass will not redraw in any condition without an interstage anneal.

9. The types of textures observed in rolled to temper and annealed to temper copper and 70/30 brass are consistent with recent theories on the development of preferred orientation.

10. There is a linear relationship between the height and position of ears on cups deep drawn from copper and 70/30 brass and the proportionate variation in R value $(R_{0,90} - R_{45})/\bar{R}$.

11. There is some indication that the position and height of ears and troughs in cups deep drawn from copper and 70/30 brass are related to the planar variations in R value when considered in a radial as opposed to a circumferential direction.

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