CHAPTERS 8 - 11 APPENDICES

THE DEVELOPMENT OF NEW ANALYTICAL TECHNIQUES APPLICABLE TO THE POWDER FORGING PROCESS

by


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CHAPTER 3

Preform Design

8.1 South Wales Forgemasters Ltd.

A pilot study of preform design for powder forging was carried out in collaboration with The South Wales Forgemasters Ltd., Cardiff. It arose as a result of a desire by South Wales Forgemasters Ltd., to ascertain the feasibility of producing by powder-forging, a turbine hub which had already been successfully forged from wrought material, see Fig. 6.1(C).

At that point in time i.e. 1973, Forgemasters, together with a number of similar small firms, see Section 2.3, were investigating the possibility of producing some future components by powder forging, and wished to gain a first-hand insight into the problems involved, since they had estimated that the initial capital outlay would be of the order of £2m. Costs included an additional building to house the new equipment and powder, since environmental conditions within their present forging shop were highly unsuitable. (Reference to this risk of contamination from dust in the air has been made more recently by Bittence.)

Since Forgemasters already possessed the means to carry out the actual forging operation e.g. see Fig. 2.12, an alternative possibility was that of purchasing preforms from an established manufacturer of sintered products, thereby reducing the amount of capital to be spent on specialised PM equipment. However, this approach was also beset with problems, since the porous preforms would quite naturally absorb any moisture with which they were likely to come into contact. This would mean that the preforms would have to be stored and transported.
in conditions of low atmospheric humidity, exercising similar care to that undertaken with porous bronze bearings prior to impregnating with oil. This risk of water absorption could, initially, result in the ferrous preforms rusting from the inside out, whilst further harmful effects might arise from the formation of superheated steam during rapid induction heating immediately prior to forging.

Although in the early days it seemed reasonable to expect that a firm such as The South Wales Forgemasters would readily encompass powder forging within the scope of its normal activities, one must not overlook the fact that there is a fundamental difference between powder forging and the conventional forging of wrought material. Whereas the latter is classified as a primary metal-working process, so that it is an accepted fact that the final shape will be produced by subsequent machining and finishing operations, the success of powder forging depends upon it being classified as a secondary process i.e. requiring little or no further machining to produce the finally desired geometry. In order to elevate the status of a forging operation from a primary to that of a secondary process, requires a considerable degree of technical expertise and sophistication, as pointed out in Section 2.14.

The ultimate outcome of the study therefore, was that Forgemasters decided against producing parts by powder forging, since without having received any definite orders or guarantees of future orders, they felt unable to justify the expense involved.

8.2 Initial Considerations in the Design of Preforms

Much of the earlier work on preform design was, of necessity, conducted on a trial and error basis. However, from the outset, the writer realised that such an approach was costly in terms of both time and money, and therefore a graphical technique involving the
use of mass distribution diagrams was devised, for components which resembled solids of revolution, see Section 2.15.

Although the results of the hot tensile tests, Section 7.4, indicate that the preform material possessed very poor ductility it was found that in practice, the sintered material could be successfully forged with more lateral flow than these results suggest.

Even so, it was regarded as futile to attempt to powder forge the turbine-hub, Fig. 2.2 from a featureless slug of material, so that the problem was that of deciding on the best embryonic preform shape to produce a successful forging.

It was obviously necessary to ensure that the distribution of the mass of material in the preform closely followed that of the finished component. However, to make them identical, so that the process becomes that of hot compaction is undesirable, since numerous studies have shown that some lateral flow is necessary to achieve optimum mechanical properties. Furthermore, it is doubtful whether complete consolidation can be obtained by hot compaction alone, since, as already implied in Section 2.4, there is a need for a deviatoric stress component to provide the necessary shear to change the shape of the void and produce complete densification.

Hot compaction and the associated complete lateral restraint would result in a state of isostatic compaction for which, Eudier suggests, the maximum attainable density is only 7.5 Kg/m$^3$ (i.e. $\rho = 0.96$).

From these considerations it follows that the best approach to the problem is to start with the "hot compacted" shape, and refine and simplify this until the desired result is achieved.

The graphical method referred to earlier, was devised for just this purpose, although it is also considered that the technique could
be used to analyse successful preform shapes which have been obtained purely by "trial and error" methods. In this way it may be possible to gain a better understanding of why these shapes are successful, thereby building up a store of knowledge which can be used to help predict preform designs for future products, in the manner outlined in Section 8.5.

8.3 The Forging Operation

The iron powder used for the manufacture of the powder forged turbine hub was Rospol MP32, which was admixed with 1% graphite and 1% zinc stearate, by weight. Details of the manufacture and design of the preform can be found in Sections 5.2, 6.2 and Ref. 28, whilst the three stages involved in the manufacture of the forging are shown illustrated in Fig. 6.1. The powder, the forging tools shown in Fig. 8.1, and the use of the 7.5 MN (750 tonf) AJAX press shown in Fig. 2.12, was kindly supplied by The South Wales Forgemasters Ltd.

With the forging tools mounted in the press the stroke of the latter was set with the aid of small pieces of lead, to produce components \( \approx 0.5 \text{ mm oversize in thickness} \). This was as a precaution from any damage that may result from overweight preforms. Consequently, very few of the forgings completely filled the die, but on average were several millimetres too small in diameter. However, this procedure had the advantage of revealing peripheral cracks more readily than if the material had contacted the die walls.

The preforms were coated with Birkatekt to prevent scaling during preheating which was carried out for 20 minutes at 900°C in a small electric muffle furnace. Attempts were made to preheat the lower die by placing red-hot billets of solid metal on it.

The dies were lubricated with Rocol J166 copper-based lubricant,
Fig. 8.1 FORGING TOOLS USED TO PRODUCE THE TURBINE HUB SHOWN IN FIG. 6.1(c) (Ref. 28)
after which the preforms were forged and then quickly transferred to cooling oil to prevent excessive scaling.

8.4 Brief Review of Some of the Phenomena Peculiar to Powder Forging

Although the principal aim of powder forging is to achieve complete densification, it is acknowledged that this may not always be practicable and that some residual porosity may therefore be inevitable. Fig. 2.11 shows such porosity occurring in the periphery of the forged turbine hubs, whilst Section 3.3 discusses the possible harmful effects arising from such porosity. In this particular instance it is undoubtedly due to a combination of peripheral tensile stress and chilling. The problem of chilling emphasizes the need for adequate preheating of the dies especially the lower one, and shorter contact times between preform and die.

The second major problem encountered with powder forging is that of peripheral cracking. Reference has already been made to this problem in Section 2.4, Section 7.4 and Section 8.2, when referring to the poor ductility of the porous material, and examples of such cracking are illustrated in Figs. 2.11 & 7.5.

Guest et al suggest that the formation of such peripheral cracks may be avoided if the spreading ratio i.e. diameter of forging to diameter of preform is kept less than \( \sim 1.35 \). This value is confirmed by the experimental work described in Section 10.2, where it is also shown that when the peripheral cracking occurs, the apparent Poisson's ratio for the material in the vicinity of the cracks is \( \sim 0.5 \). This coincides with the findings of the tensile tests described in Section 7.6 where it was suspected that before the onset of fracture \( \nu_{ap} \sim 0.5 \). These observations simply indicate that the material is deforming with constant volume i.e. without
densification, and must not be mistaken as indicating true plastic deformation.

It is likely that the problem of chilling will also tend to aggravate peripheral cracking since it reduces the ductility of the preform material. It is possible that the use of larger diameter preforms resulting is less lateral displacement during forging may help to lessen the effects of chilling, and would certainly make peripheral cracking virtually non-existent. Bockstiegel and Olsen comment on the fact that when they produced a certain component the material flowed radially inwards, so that no tensile stresses, and consequently no peripheral cracking occurred. Fig. 8.2 shows that the graphical technique described in Section 8.5 will clearly indicate the regions in which such flow has taken place in forming the component.

Fig. 8.3 shows another fault that can occur with powder forging. In this case it takes the form of blisters on the outer surface of the component and is probably due to air entrapment. Fig. 8.4 shows the backward extrusion of flash due to the combination of an overweight preform and excessive clearance between the punch and die of the forging tools.

Fig. 8.5 shows an extremely interesting phenomena namely that of inverse barrelling, which is contrary to what one normally expects from a compressive loading e.g. see Figs. 9.1 and 9.5. Hirschhorn and Bargainnier suggest that inverse barrelling occurs because the surfaces of the preform in contact with the tooling will densify first and when friction is overcome, these regions will be the first to undergo true plastic deformation. They will therefore flow radially outwards, whilst the material sandwiched between is still in the process of densifying.
Fig. 8.2 Metal-displacement diagram.

(Ref. 28)
Fig. 8.3  BLISTERS ON SURFACE OF COMPONENT DUE TO POSSIBLE AIR ENTRAPMENT
Fig. 8.4  BACKWARD EXTRUSION OF FLASH AROUND CIRCUMFERENCE OF COMPONENT. (THE FLASH APPEARS IRREGULAR DUE TO ACCIDENTAL DAMAGE)

Fig. 8.5  INVERSE BARRELLING OF OUTER PERIPHERAL SURFACE OF COMPONENT
8.5 Use of Mass Distribution Diagrams to Analyse Turbine Hub Preform Design

Since the turbine hub and hence its corresponding preform were both solids of revolution it was possible to represent and compare their mass distributions in the form of diagrams, devised in the following manner.

For a solid of revolution, the mass of an elemental ring of material occurring at radius \( r \) is given by \( 2\pi \rho t \, dr \), where \( \rho \) and \( t \) are the density and thickness respectively. Note that \( \rho, t \) and \( r \) are all variables; \( \rho \) depending upon the method of manufacture, and \( t \) and \( r \) upon the geometry of the component.

Total mass of preform or component = \[ 2\pi \int_0^R \rho t \, dr \] \[ (27) \]

From this it can be seen that the mass may be represented by the area under the mass distribution diagram, in which values of \( 2\pi \rho t \) are plotted on the ordinate scale and \( r \) on the abscissa. The mass of metal contained between any two radii can then be obtained from such a diagram by measuring the area between the radii in question with a planimeter. Furthermore, by superimposing the mass-distribution diagrams for both component and preform on top of one another, it is possible to locate the precise regions from which metal has been displaced, together with the actual quantities involved. Using this technique a preform shape may be gradually refined until it eventually produces the desired result.

Another advantage of this graphical method is that it overcomes the need to sub-divide the component into convenient portions whose volumes can either be determined by formulae, or whose centroids can be accurately located to allow the use of the theorem of Pappus or Guldinus, to determine the volumes. All that is needed is an accurate
drawing of the component from which the thickness may be scaled off at its principal points i.e. changes of section, etc., and these values can then be tabulated against the corresponding radii and density to enable the values \( \frac{\tau}{\rho} \) to be calculated. However, it will of course be necessary, as it would in any event, to obtain density distribution diagrams for both the preform and the component. In the case of the turbine hub preform and forging, these were obtained experimentally by removing annular concentric rings of metal on a lathe, and by drilling and boring out the centres of the parts as shown in Fig. 8.6 and Fig. 8.7. In each case, the volume of metal removed was determined by measurement and calculation, and its weight determined by weighing before and after machining. The mean density of the portion removed was then determined, and taken to act at the mean radius.

The density distribution diagrams obtained in this way are shown in Fig. 8.8, together with the corresponding mass distribution diagrams, Figs. 8.9 and 8.10, whilst the metal displacement diagrams are shown on Fig. 8.2.

8.6 Fatigue Testing of Partly Forged PM Material

Fatigue tests were carried out on partly forged preform material for the reasons outlined in Section 3.3. The powder mix used was as described in Section 8.3 and the compacts were prepared using the tool described in Section 5.3, followed by sintering as described in Section 6.1.

All the compacts were prepared at the same compaction pressure to give an average relative density \( \rho \) of 0.88, and an average thickness of 14.3 mm.

The forging tools used were of similar construction to the compaction tools described in Section 5.3 except that the die aperture and
**Fig. 8.6**

USE OF DRILLING MACHINE TO REMOVE METAL FROM CENTRAL REGION OF TURBINE HUB TO DETERMINE LATERAL DENSITY DISTRIBUTION THROUGHOUT COMPONENT

**Fig. 8.7**

USE OF THE LATHE TO REMOVE METAL FROM PERIPHERY OF TURBINE HUB, TO DETERMINE LATERAL DENSITY DISTRIBUTION THROUGHOUT COMPONENT
Fig. 8.8 Lateral density-distribution diagrams for selected preforms and forgings.
(Ref. 28)
Fig. 8.9 Mass-distribution diagrams: (a) Preform $\rho^* \text{ (mean)} = 0.78$ and corresponding forging; (b) Preform $\rho^* \text{ (mean)} = 0.81$ and corresponding forging.

(Ref. 28)
the corresponding punches were slightly larger measuring 77mm x 20mm overall.

The "Armstrong" Press shown in Fig. 5.1 was used for the actual forging operation, but without the use of the accumulators. The sintered PM compacts were heated to 1000°C in an electrically operated muffle furnace with no protection from the surrounding atmosphere. They were then transferred to the unheated forging tools and the forging operation was carried out without the use of any lubricant. The actual temperature of the sintered compact at the time of forging was between 700°C and 750°C. Table 8.1 shows the four different forging pressures used, together with the corresponding average porosities obtained.

The partly forged specimen were then machined on a copying lathe to provide No. 13 Hounsfield tensile specimen, and also fatigue specimen, see Fig. 8.10, suitable for use on an Avery Fatigue Testing Machine, Type 7305. This machine subjected the specimen to plane bending with a four point loading system, whilst a mean stress of zero was used for all the tests. Although no longitudinal polishing of the fatigue specimen was undertaken, the actual rate of metal removal in producing the specimen was in accordance with the recommendations of B.S. 3518, "Methods of Fatigue Testing". The results of the fatigue tests in the form of "S/N curves" are shown as Figs. 8.11 to 8.14 inclusive. The endurance limit for each set of results was taken as the stress corresponding to 10⁷ reversals without failure. These values are shown in Table 8.1, together with the average U.T.S. values obtained from Hounsfield tests conducted on 3 specimen from each batch.

Fig. 8.15 shows the graphical representation of the overall results. From this it can be seen that the endurance limit increases
Fig. 8.10  
FATIGUE SPECIMEN PARTLY MOUNTED IN AVERY FATIGUE TESTING MACHINE, TYPE 7305
FIG. 8.11  RELATIONSHIP BETWEEN ALTERNATING STRESS AND NUMBER OF CYCLES FOR PARTLY FORGED
MATERIAL OF AVERAGE POROSITY P = 0.208
FIG. 8.12  RELATIONSHIP BETWEEN ALTERNATING STRESS AND NUMBER OF CYCLES FOR PARTLY FORGED
MATERIAL OF AVERAGE POROSITY $P = 0.157$
FIG. 8.13

RELATIONSHIP BETWEEN ALTERNATING STRESS AND NUMBER OF CYCLES FOR PARTLY FORGED
MATERIAL OF AVERAGE POROSITY $P = 0.138$
FIG. 8.14  RELATIONSHIP BETWEEN ALTERNATING STRESS AND NUMBER OF CYCLES FOR PARTLY FORGED MATERIAL OF AVERAGE POROSITY $P = 0.125$
FIG. 8.15 ENDURANCE LIMIT AND ENDURANCE RATIO OF PARTLY FORGED PREFORM MATERIAL AS INFLUENCED BY VARYING AMOUNTS OF RESIDUAL POROSITY

ENDURANCE LIMIT (± MN/m²) vs. POROSITY (P)
As the porosity diminishes which corresponds to the findings of Bockstiegel and Blande. The endurance/U.T.S. ratio is 0.6 which just places it in the generally accepted range of 0.4 - 0.6 for ferrous based materials. However, this value is higher than that obtained by Bockstiegel and Blande, whose values over the same porosity range are \( \sim 0.4 \), which agrees with the value anticipated by Lawley.

The main reason for these differences is probably due to the method of fatigue testing used in this instance. With hindsight it is considered that a truer representation of fatigue properties with less scatter of results would have been obtained if either rotating bending tests with cylindrical specimen or plane bending tests with rectangular specimen had been used. Both of these methods would have subjected a larger proportion of the outer surface of the specimen to the maximum applied stress. This in turn would have considerably increased the chances of crack initiation occurring at the weakest point on the specimen surface, thereby giving lower fatigue strengths than are likely to be obtained from plane bending tests conducted on cylindrical specimen. The reason for this is that the latter arrangement only subjects a small
proportion of the specimen surface to the maximum applied stress, and
the actual amount of porosity contained in this small region is purely
a matter of chance. It is therefore considered that the results of such
fatigue tests can only be regarded as being truly representative, when
applied to high integrity materials.
CHAPTER 9

Analysis of Consolidation During the Upsetting Stage of the Powder Forging Process

9.1 Introduction

Powder forging owes much of its success to the fact that from the outset an enlightened approach was adopted to ensure that proper scientific studies were conducted \(^{34, 43, 126, 127, 128, 129}\) in order to develop quantitative analytical techniques to describe and predict the behaviour of the material during the actual forging operation. This approach has paid handsome dividends in providing a basis for well reasoned scientific approaches to the design of dies and preforms, thereby considerably reducing much of the costly trial and error techniques that were necessary in the early stages of development \(^{13}\).

The writer can claim to have made a small contribution to the sum total of knowledge that exists in this field, by the publication in 1976 of a paper entitled, "Compatibility equations for the powder-forging process" \(^{18}\). These equations were derived by resorting to the artifice of assuming an apparent plastic Poisson's ratio \(v_{\text{ap}}\), which experimental evidence \(^{129, 130}\) suggests is a function of the relative density \(\rho^*\) of the preform i.e.

\[
v_{\text{ap}} = f(\rho^*)
\]

(28)

However, the forms of the empirical equations connecting \(v_{\text{ap}}\) with \(\rho^*\) are merely the result of curve fitting exercises, and as such, are nothing more than a convenient means of linking together a limited range of experimental observations (see Section 3.4), which for the most part only apply to contrived situations. Generally it is
assumed that frictionless homogeneous compression occurs throughout, so that there is no barrelling. However, the type of situation normally encountered in practice is as shown schematically in Fig. 9.1, where the friction at the die/preform interface restricts lateral spreading at the ends of the preform, causing the central portion to bulge outwards as shown. However, if the diameter/length ratio is sufficiently large, see Section 9.4, then these frictional ends effects may tend to prevent barrelling from taking place.

9.2 A Philosophical Approach to an Explanation of the Consolidation Process

Although it is appreciated that the following explanation of the consolidation process may not be strictly accurate, it does help to explain a number of observed phenomena. With reference to the variable morphology model shown in Fig. 9.2, it can be seen that 25% of the porosity, marked (i), is favourably oriented to collapse during the early stages of upsetting. Whilst this collapse is taking place, the spherical pores, marked (ii), will presumably tend to become favourably oriented oblate ellipsoids, so that an additional 25% of the original porosity will then close without too much difficulty. During the closure of this first 50% of the porosity, it must be assumed that the morphology of the bulk of the remaining porosity, marked (iii), will tend to become spherical, and ultimately close up in a manner similar to that already described for the spherical pores. However it will be virtually impossible to completely close all the remaining porosity since Kaufman draws attention to the creation of second generation porosity during consolidation, which together with pores \(< 5 \mu m\) diameter is very difficult to close.

Some of the foregoing predictions have been borne out by the work
FIG. 9.1 SCHEMATIC ILLUSTRATION OF THE CHARACTERISTIC MODES OF DEFORMATION AS SEEN ON A LONGITUDINAL SECTION THROUGH A CYLINDER UPSET BETWEEN FLAT DIES WITH NO LUBRICATION. REGION 1 IS UNDER NEAR-HYDROSTATIC PRESSURE; REGION 2 UNDERGOES HIGH SHEAR; REGION 3 UNDERGOES SMALL AXIAL COMPRESSION AND CIRCUMFERENTIAL TENSION

(Ref. 41)
FIG. 9.2  USE OF VARIABLE MORPHOLOGY MODEL TO HELP EXPLAIN A NUMBER OF PHENOMENA WHICH OCCUR DURING CONSOLIDATION
of Gaigher and Lawley in as much as they observed the rapid closure of favourably oriented pores, and comment on the difficulties experienced in trying to eliminate those pores not favourably oriented. Furthermore, they also found that rapid densification could be achieved from an initial density \( \rho^* \) of \( \sim 0.86 \) to a density of \( \sim 0.94 \), which represents a reduction in porosity of \( \sim 50\% \).

It is obvious that during pore closure, the matrix material of the preform must undergo plastic deformation. This will result in some of the metal being used to fill in the pore volume, whilst the bulk of this metal will flow radially outwards to produce lateral spreading. The apparent plastic Poisson's ratio is the result of the combined effects of pore closure and plastic deformation occurring simultaneously.

9.3 Theoretical Considerations Leading to the Concept of a Coefficient of Consolidation

If PM material is regarded as consisting of two component parts viz. a volume of solid metal, and a volume representing the total porosity, then the solid metal portion will deform in accordance with the ideal plastic Poisson's ration \( \nu = 0.5 \). However, it is necessary to attempt to anticipate the manner in which the porosity is likely to diminish, by making recourse to known qualitative experimental observations. The changing geometry of a PM preform during an actual forging operation can then be ascertained by combining together the individual behaviour patterns of the two component parts.

However, since the equations deduced by this method are based on limited experimental evidence, it is imperative that the effects they predict, other than those already considered, are verified by a suitably planned experimental programme, see Section 3.5.
FIG. 9.3 GEOMETRICAL CHANGES OCCURRING IN A CYLINDRICAL PM PREFORM SUBJECTED TO AXISYMMETRIC UPSETTING

(i) Before upsetting
(ii) After frictionless axisymmetrical upsetting, or, if $h_1$ is infinitesimal, after axisymmetric upsetting of an elemental disc either with or without frictional constraint
With reference to the cylindrical preform shown in Fig. 9.3(i)

\[ V_1 = \frac{\pi}{4} D_1^2 h_1 \]  
\[ V_S = \frac{\pi}{4} D_1^2 h_1 \rho \]  
\[ V_{Cl} = \frac{\pi}{4} D_1^2 h_1 P_1 \]

(29)
(30)
(31)

Although it is fully appreciated that the porosity is completely intermingled with the solid metal, nevertheless, equations (30) and (31) describe two cylinders which represent the proportions of solid metal and porosity, respectively.

The equivalent heights of these cylinders, corresponding to a given diameter \( D \), are \( h_\rho \) for the solid and \( h_P \) for the porosity, so that for a given radial strain \( \varepsilon_\rho \), the corresponding equivalent longitudinal strains will be \( \ln \left( \frac{h_2 \rho}{h_1 \rho} \right) \) and \( \ln \left( \frac{h_2 P}{h_1 P} \right) \) respectively.

Since equation (30) represents a cylinder of solid metal, the volume of which remains constant throughout the upsetting shown in Fig. 9.3, then,

\[ D_1^2 h_1 \rho = D_2^2 h_2 \rho \]
or,
\[ \ln \frac{D_2}{D_1} = \varepsilon_\rho = 0.5 \ln \left( \frac{h_1 \rho}{h_2 \rho} \right) \]

where the constant \( 0.5 = \sqrt{\gamma} \) i.e. Poisson's ratio for a perfectly plastic solid metal, so that,

\[ \varepsilon_\rho = \sqrt{\gamma} \ln \left( \frac{h_1 \rho}{h_2 \rho} \right) \]  

(32)

Alternatively, this can be expressed as the exponential law,
which indicates that the equivalent height of the solid cylinder reduces exponentially with increase in radial strain, as shown by the curve (A), Fig. 9.4.

The volume of the porosity, as given by equation (31) diminishes as upsetting proceeds. Experimental observation suggests that, although initially the porosity diminishes quite rapidly, the rate at which it diminishes becomes progressively less as upsetting continues. Furthermore, the compatibility equations derived by Griffiths et al. suggest that it is impossible to achieve 100% consolidation by upsetting alone. This implies that as \( \varepsilon \rightarrow 0 \), \( \varepsilon \rightarrow 0 \) but \( \varepsilon \rightarrow \infty \) although in practice, \( \varepsilon \) is unlikely to exceed \( 0.4 \), due to peripheral cracking.

In view of this, it would seem reasonable to suppose that the changing geometry of the cylinder representing the porosity could also be expressed by an exponential law of the form,

\[
h_2 P_2 = h_1 P_1 e^{-\frac{\varepsilon \psi}{c}}
\]

(34)

In general, \( C \ll \nu \) (i.e. \( C < 0.5 \)) since the volume will normally be decreasing. Equation (34) is shown as curve (B), Fig. 9.4. It has been deliberately drawn on the underside of the abscissa, so that the overall preform height, corresponding to a given value of \( \varepsilon \), may be conveniently obtained by simply erecting an ordinate from curve (B) to curve (A), to pass through the value of \( \varepsilon \) being considered.

It follows from equation (34) that

\[
\frac{d(hP)}{d\varepsilon} = -\frac{hP}{c}
\]

or

\[
c = -\frac{d\varepsilon}{d \left( \frac{hP}{hP} \right)}
\]

(35)
FIG. 9.4 EXPONENTIAL CURVES SHOWING THE VARIATION IN EQUIVALENT HEIGHTS CORRESPONDING TO RADIAL STRAIN, FOR THE COMPONENT PARTS OF A TYPICAL CYLINDRICAL PH PREFORM SUBJECTED TO AXISYMMETRIC UPSETTING
The constant C, can therefore be defined as the ratio of lateral to longitudinal "strain", and in this respect it can be regarded as being analogous to a Poisson's ratio.

It is proposed to refer to this constant "C", as the "coefficient of consolidation", the actual value of which will depend not only upon the mode of deformation, and prevailing frictional conditions, but also upon the morphology of the porosity present, and the orientation of its principal axes in relation to the direction of the applied compressive load, see Section 9.2. This value of "C" will therefore provide a quantitative measure of the way in which the porosity diminishes for a given set of conditions, hence making it possible to draw direct comparisons between various consolidation processes.

Hitherto it has only been possible to establish empirical fits connecting \( \sqrt{\alpha_p} \) and \( \rho_\ast \), yielding meaningless constants which have not made any worthwhile contribution to a better understanding of the consolidation process.

9.4 Equations to Describe the Changing Geometry of PM Preforms During Powder Forging

Compatibility equations, together with an equation for \( \sqrt{\alpha_p} \), can now be derived in terms of \( \sqrt{\alpha} \), C and \( \rho_\ast \), and the effects predicted by these equations can be examined by means of a suitably planned experimental programme.

From equation (34),

\[
\mathbf{E_p} = C \ln \left( \frac{h_1 P_1}{h_2 P_2} \right)
\]

Substituting for \( \mathbf{E_p} \) from eqn. (32)

\[
\sqrt{\alpha} \ln \left( \frac{h_1 \rho_1}{h_2 \rho_2} \right) = C \ln \left( \frac{h_1 P_1}{h_2 P_2} \right)
\]
Since \( -\varepsilon_z = \ln \left( \frac{h_2}{h_1} \right) \) then it can readily be shown that,

\[
-\varepsilon_z = \left( \frac{c}{\sqrt{\nu - c}} \right) \ln \frac{P_1}{P_2} + \left( \frac{\nu}{\sqrt{\nu - c}} \right) \ln \frac{\rho*2}{\rho*1} \tag{38}
\]

whilst combining equations (32) and (36) gives,

\[
\varepsilon_\perp = \left( \frac{\nu}{\sqrt{\nu - c}} \right) \ln \left( \frac{P_1 / \rho*2}{P_2 / \rho*1} \right) \tag{39}
\]

Although in the strictest sense, equation (38) can only be applied to frictionless axisymmetric upsetting, equation (39) can be used to measure the lateral strain across any section throughout the length of the preform irrespective whether or not barrelling has occurred due to frictional constraint. Although equation (39) may appear to be the consequence of considering frictionless axisymmetric upsetting only, the same underlying reasoning can be applied to elemental discs of material wherever they occur within the test-piece i.e. whether at section X or section Y, Fig. 9.5. The only difference is that the corresponding value for "C" will be less at X, due to friction at the die/preform interface, than it will at section Y, where the effects of friction are considerably smaller.

Furthermore, it can be seen from equation (37) that the graph of

\[ \sqrt{\ln \left( \frac{h_1 / \rho*1}{h_2 / \rho*2} \right)} \text{ vs. } \ln \left( \frac{h_1 P_1}{h_2 P_2} \right) \]

should be linear, with a slope equal to "C".

Although "C" has greater fundamental significance than \( \sqrt{\nu} \) ap, the latter may be readily derived from equations (38) and (39) since by definition,

\[ \sqrt{\nu} \text{ ap} = -\frac{d\varepsilon_\perp}{d\varepsilon_z} \]
FIG. 9.5 CYLINDRICAL PM PREFORM AFTER UMLUBRICATED AXISYMMETRIC UPSETTING
hence, \( \nabla \mathrm{ap} = \frac{\nabla c}{\nabla - (\nabla - c) \rho} \) where \( \nabla = 0.5 \)

\[ \therefore \nabla \mathrm{ap} = \frac{c}{1 - (1 - 2c) \rho} \]  

which is in accordance with equation (28). However, it is interesting to note that unlike previous empirical relationships connecting \( \nabla \mathrm{ap} \) and \( \rho \), equation (40) has universal application, with values for \( c \) ranging from 0 to 0.5, respectively, depending upon whether there is complete lateral restraint as in the case of repressing, or whether there is deformation of porous material without densification i.e. deformation at constant volume. This latter effect appears immediately before the onset of fracture, whether during a tensile test, see Section 7.6, or due to peripheral cracking of an upset preform, see Section 10.2. When these extreme values of 0 and 0.5 are substituted for \( c \) in equation (40), it can be seen that \( \nabla \mathrm{ap} = 0 \) for repressing, and \( \nabla \mathrm{ap} = 0.5 \) for the deformation of a porous material without densification. This latter value remains constant irrespective of preform density, and must not be mistaken for the true plastic Poisson's ratio \( \nabla = 0.5 \), which in the case of porous preforms, will only occur when \( \rho \rightarrow 1 \), and \( c \) has some value intermediate between 0 and 0.5.

9.5 Unlubricated Cold Axisymmetric Compression of Cylindrical Preforms

This programme of experimental work was undertaken for the purpose of gaining a first-hand appreciation of the deformation characteristics of cylindrical porous preforms of different densities and different test-piece geometries. It was also intended as a means of checking the validity of the reasoning applied in Section 9.3 leading to the concept of a "coefficient of consolidation". It was regarded as
justifiable to relate the findings of these cold axisymmetric tests to hot working conditions since Kaufman states that the ultimate degree of densification attainable is independent of forging temperature.

The powder used was Hoganas AHC 100.29 iron powder which was admixed with 1% zinc stearate and 0.5% graphite. Preforms measuring 40 mm diameter x 50 mm long were produced to nominal relative densities (\(\rho_0\)) of 0.7 and 0.83 with the aid of the floating die compaction tool shown in Fig. 5.8. These values of \(\rho_0\) were selected to encompass the practical range of preform densities suggested by Cull.

The compacts were sintered in accordance with the procedure described in Section 6.1, after which test-pieces of both densities were machined, without the use of cutting fluid, to the dimensions shown in Table 9.1, and their end faces were lightly ground. One of the purposes of using different \(D_1/h_1\) ratios was to enable the Cook and Larke method to be used to correct for frictional end effects. This simulation of frictionless conditions will provide results analogous to those corresponding to the frictionless homogeneous compression referred to in Section 9.1.

Each of the specimen was subjected to a series of compression

<table>
<thead>
<tr>
<th>Diameter (D_1) mm</th>
<th>25</th>
<th>40</th>
<th>40</th>
<th>40</th>
<th>40</th>
</tr>
</thead>
<tbody>
<tr>
<td>Length (h_1) mm</td>
<td>50</td>
<td>40</td>
<td>20</td>
<td>13.33</td>
<td>10</td>
</tr>
<tr>
<td>(D_1/h_1) ratio</td>
<td>0.5</td>
<td>1</td>
<td>2</td>
<td>3</td>
<td>4</td>
</tr>
</tbody>
</table>
tests using the axisymmetric compression tool shown in Fig. 9.6. No lubrication was used at any stage, since it was important for the frictional conditions to remain completely unaltered throughout the tests. After each load increment the new overall dimensions of the specimen were checked, and in the case of excessive barrelling, the new volumes were measured by means of an Archimedean technique.

The results of these tests are shown in Figs. 9.7 and 9.8. Unfortunately it was not practical to use $D_1/h_1$ ratios $< 0.5$ due to the slender test-pieces becoming unstable during compression. The densities corresponding to decrements of 5% reduction in height up to a maximum of 30%, were then obtained from Figs. 9.7(i) and Fig. 9.8(i) and (ii), and corrected where necessary, to correspond to initial starting densities ($\rho_*$) of either 0.698 or 0.83. These values are shown replotted as Figs. 9.9 and 9.10 respectively, and the curves extrapolated to $D_1/h_1 = 0$, which corresponds to an infinitely long test-specimen, and hence frictionless homogeneous deformation.

The density values obtained in this way, together with their respective reductions in height were substituted into equation (37) to enable Fig. 9.11 to be plotted. From this it can be seen that the graphs are linear thereby justifying the earlier assumptions leading to the concept of a coefficient of consolidation. The slopes of these graphs i.e. the actual values for "C", the coefficients of consolidation, are as follows:

(i) $C = 0.15$ for density range $\rho_* = 0.698 \rightarrow 0.762$, and

(ii) $C = 0.14$ for density range $\rho_* = 0.83 \rightarrow 0.91$.

Since both the test-pieces having an initial $D_1/h_1$ ratio of 4 exhibited a negligible amount of barrelling, then their experimental results were substituted directly in equation (37) to enable Fig. 9.12 to be
Fig. 9.6  AXISYMMETRIC COMPRESSION TOOL
FIG. 9.7(1)  
VARIATION OF RELATIVE DENSITY WITH REDUCTION IN HEIGHT  
FOR UNLUBRICATED COLD UPSETTING OF CYLINDRICAL PREFORMS  
OF VARYING D/h RATIOS AND INITIAL RELATIVE DENSITY  
\( \rho \) (NOMINAL) = 0.698
FIG. 9.7(ii) VARIATION OF RELATIVE DENSITY WITH REDUCTION IN HEIGHT FOR UNLUBRICATED COLD UPSETTING OF CYLINDRICAL PREFORMS OF VARYING D/h RATIOS AND INITIAL RELATIVE DENSITY $\rho \cdot$ (NOMINAL) = 0.698
FIG. 9.8(1) VARIATION OF RELATIVE DENSITY WITH REDUCTION IN HEIGHT FOR UNLUBRICATED COLD UPSETTING OF CYLINDRICAL PREFORMS OF VARYING D/h RATIOS AND INITIAL RELATIVE DENSITY $\rho_r$ (NOMINAL) = 0.83
FIG. 9.8(ii) VARIATION OF RELATIVE DENSITY WITH REDUCTION IN HEIGHT
FOR UNLUBRICATED COLD UPSETTING OF CYLINDRICAL PREFORMS
OF VARYING D/h RATIOS AND INITIAL RELATIVE DENSITY
\( \rho^* \) (NOMINAL) = 0.83
FIG. 9.9 VARIATION OF RELATIVE DENSITY WITH D/h RATIOS FOR DIFFERENT REDUCTIONS BASED ON DATA OBTAINED FROM FIGS. 9.7(i) AND (ii)
FIG. 9.10 VARIATION OF RELATIVE DENSITY WITH $D_1/h_1$ RATIOS FOR DIFFERENT REDUCTIONS BASED ON DATA OBTAINED FROM FIGS. 9.3(i) and (ii)
Fig. 9.11
Graphs of $0.5 \ln \left( \frac{h_1 \rho_1^*}{h_2 \rho_2^*} \right)$ vs. $\ln \left( \frac{h_1 P_1}{h_2 P_2} \right)$ based on data obtained from Figs. 9.9 and 9.10 to represent simulated frictionless conditions, i.e. Cook and Lark extrapolations.
FIG. 9.12 GRAPHS OF $0.5 \ln \left( \frac{h_1 \rho_{*1}}{h_2 \rho_{*2}} \right)$ vs $\ln \left( \frac{h_1 P_1}{h_2 P_2} \right)$ FOR UNLUBRICATED AXISYMMETRIC UPSETTING
plotted, which now relates to unlubricated axisymmetric compression of a relatively thin disc. Although this condition would not result in a considerable density variation across the section, nevertheless, the deformation would not be completely homogeneous, so that when the circumferential material at the ends of the specimen becomes completely densified, inverse barrelling, see Fig. 8.6, would probably occur. The apparent absence of barrelling in this case is due to the influence that the frictional end effects have on specimen of such large D/h ratios.

With reference to Fig. 9.12 it can again be seen that the graphs are linear, thereby providing further experimental verification of the existence of a coefficient of consolidation "C" the actual values of which in this case are:-

(i) $C = 0.1$ for density range $\rho_\ast = 0.689 \rightarrow 0.841$, and
(ii) $C = 0.121$ for density range $\rho_\ast = 0.822 \rightarrow 0.911$.

9.6 Apparent Plastic Poisson's Ratios, and the Significance of the Coefficient of Consolidation

If the experimentally determined values for "C" quoted in Section 9.5, are substituted into equation (40), then the corresponding values for the apparent plastic Poisson's ratios, are as follows:-

For frictionless conditions:

\[ \sqrt{ap} = \frac{0.154}{1 - 0.692 \rho_\ast} \quad \text{------------------- (41)} \]

for the density range $\rho_\ast = 0.698 \rightarrow 0.962$

and \[ \sqrt{ap} = \frac{0.14}{1 - 0.72 \rho_\ast} \quad \text{------------------- (42)} \]

for the density range $\rho_\ast = 0.83 \rightarrow 0.91$.

For unlubricated conditions:-

93
\[ \nabla \varepsilon_{ap} = \frac{0.1}{1 - 0.3\rho} \]  
(43)

for the density range \[ \rho = 0.689 \rightarrow 0.841 \]

and \[ \nabla \varepsilon_{ap} = \frac{0.121}{1 - 0.758\rho} \]  
(44)

for the density range \[ \rho = 0.822 \rightarrow 0.911. \]

The curves representing the above data are shown in Fig. 9.13.

For the conditions appertaining to equation (42) Kuhn and Downey propose the relationship,

\[ \nabla \varepsilon_{ap} = 0.5 \rho^2 \]  
(45)

which yields almost identical results to those obtained from equation (42).

Although Kuhn states that it is the level of porosity which has the greatest influence on the values for \( \nabla \varepsilon_{ap} \), with initial density having a negligible effect, it would appear that all his published experimental work is confined to the approximate density range \( \rho = 0.82 \rightarrow 1 \). With reference to Fig. 9.13 it can be seen that for a lower density range, the material behaviour as given by equation (41) differs from that given by equation (42), the latter corresponding to a density range similar to that used by Kuhn. The experimental evidence indicates that during frictionless axisymmetric upsetting over the lower density range, there is a greater amount of lateral spreading of the material, accompanied by a smaller degree of densification. Furthermore, it can be seen that the amount of lateral spreading is less, and hence the densification is greater, for both cases of unlubricated compression as compared to frictionless axisymmetric upsetting.

This apparent dependence of \( \nabla \varepsilon_{ap} \) on the initial preform density
FIG. 9.13  CURVES OF APPARENT PLASTIC POISSON'S RATIO VS.
RELATIVE DENSITY FOR FRICTIONLESS AND UNLUBRICATED
CONDITIONS, USING PREFORMS OF DIFFERENT INITIAL DENSITIES
and prevailing frictional conditions, can possibly be explained as follows. As the density of the preform decreases, so does the load bearing area decrease due to an increase in the size and number of the pores. The effect of this is not only to reduce the load at which the matrix material will begin to yield, but also to reduce the load at which the favourably oriented larger pores will now begin to collapse. Although these two events will occur simultaneously, a slight preference will be given to either one or the other depending upon the prevailing conditions. A measure of the manner in which this sequence of events will manifest itself on a macro-scale, can be obtained by determining the value of "C" for the particular process. This value for C may then be substituted in equation (40) to provide a relationship connecting \( \sqrt[\nu]{a_p} \) and \( \rho \), for the density range considered.

With reference to Fig. 9.13, it can be seen that there is a much greater difference in the sequence of these events in the case of the lower density preforms. However, it is interesting to note that for both levels of preform density, the average value for "C" is approximately the same i.e. for the lower density preforms, average value for C = \( \frac{0.154 + 0.1}{2} \) = 0.127, whilst for the higher density preforms, average value for C = \( \frac{0.14 + 0.121}{2} \) = 0.13.

These intermediate values for C will apply to lubricated axisymmetric upsetting, when it would appear that \( \sqrt[\nu]{a_p} \) is no longer dependent upon initial preform density.

9.7 An Appreciation of the Plastic Flow Stress Requirements for Cold Axisymmetric Upsetting

The flow stresses or end stresses occurring during the tests described in Section 9.5, were calculated by dividing the applied load
by the cross-sectional area of the end face of the deformed test-piece. Figs. 9.14, 9.15 and 9.16 show these values plotted to a base of relative density for a selection of different test-piece geometries.

The experimental curves obtained for the lower density preforms are very similar, the notable difference being that as the D/L ratio increases so does the practical range of densification increase, due to a marked absence of the barrelling effect, see Section 9.5. However, for the higher density preforms, the curves tend to become progressively steeper as the D/L ratio increases, indicating an increase in the flow stress required to bring about the same change in density. It is suggested that the flow stress requirements for preforms, with initial densities lying between those indicated in Figs. 9.14, 9.15 and 9.16, may be determined from these diagrams, by interpolation.

The values for flow stress corresponding to different height reductions and D/L ratios were then plotted as shown in Fig. 9.17 and extrapolated back to zero D/L ratio to provide the frictionless flow stress, or "true" stress ($\sigma_t$). Thus the "true" stress VS, natural strain curve shown as Fig. 9.18 was plotted. The empirical fit for such a curve for a ductile material possessing a cubic lattice is generally taken as being

$$\sigma_t = B \varepsilon^\eta$$  \hspace{1cm} (46)

where $B$ is the strength coefficient and $\eta$ is the work hardening exponent. Equation (46) may be linearized as follows:

$$\log \sigma_t = \eta \log \varepsilon + \log B$$  \hspace{1cm} (47)

so that $\eta$ becomes the slope of the graph of $\log \sigma_t$ VS $\log \varepsilon$ and $\log B$ is the intercept on the $\log \sigma_t$ axis corresponding to $\log \varepsilon = 0$.

Using the "method of least squares" described in Section 7.10,
FIG. 9.14  CURVES OF FLOW STRESS VS. RELATIVE DENSITY FOR COLD UNLUBRICATED AXISYMMETRIC COMPRESSION OF CYLINDRICAL PREFORMS OF VARYING INITIAL DENSITIES AND $D_1/h_1$ RATIO = 0.5
FIG. 9.15  CURVES OF FLOW STRESS VS. RELATIVE DENSITY FOR COLD UNLUBRICATED AXISYMMETRIC COMPRESSION OF CYLINDRICAL PREFORMS OF VARYING INITIAL DENSITIES AND $D_1/h_1$ RATIO = 1
FIG. 9.16  CURVES OF FLOW STRESS VS. RELATIVE DENSITY OF COLD UNLUBRICATED AXISYMMETRIC COMPRESSION OF CYLINDRICAL PREFORMS OF VARYING INITIAL DENSITIES AND \( \frac{D_1}{h_1} \) RATIO = 4
FIG. 9.17
VARIATION OF FLOW STRESS WITH REDUCTION IN HEIGHT FOR DIFFERENT $D_1/h_1$ RATIOS, BASED ON DATA OBTAINED FROM FIGS. 9.14 TO 9.16
\[ \sigma = 928 \varepsilon^{0.43} \text{MN/m}^2 \]

(EMPirical CURve FIT)

* RESULTS OBTAINED FROM FIG. 9.17
COOK & LARKE EXTRAPOLATION

INITIAL RELATIVE DENSITY OF PREFORM
= 0.83 (NOMINAL)

**FIG. 9.18** CURVE OF "TRUE" STRESS VS. COMPRESSIVE NATURAL OR LOGARITHMIC STRAIN ($\varepsilon_z$)
it was found that the best empirical fit to the Cook and Larke extrapolated data shown in Fig. 9.1 was,

\[ \sigma_t = 928 \varepsilon_z^{0.443} \text{ MN/m}^2 \]  

(48)

with a coefficient of correlation \( R = 0.976 \). Kuhn states that for pure iron, \( \eta_0 = 0.31 \), and any excess over this value for porous compacts is due to geometric work hardening resulting from a continuous increase in the load bearing area as densification proceeds. Furthermore, he gives the empirical relationship,

\[ \eta_0 = 0.31 \rho_0^{-1.91} \]  

(49)

where \( \rho_0 \) is the initial relative density of the preform. From equation (49) \( \eta_0 = 0.443 \) when \( \rho_0 = 0.83 \). Substituting \( \rho_0 = 0.83 \) in equation (49) gives the value for \( \eta_0 = 0.443 \), thus indicating that the experimental results are in complete agreement with Kuhn's findings.
10.1 **Hot Axisymmetric Upsetting of Cylindrical Preforms**

The powders used for these tests were the "as supplied" Hoganas AHC 100.29 iron powder, and also the coarse fraction i.e. $+150 - 180 \, \mu m$ particle size. In both cases the powders were admixed with 1% zinc stearate and 0.5% graphite and preforms measuring 40 mm diameter x 50 mm long were produced to four different nominal relative densities viz. $\rho_r = 0.7$, 0.75, 0.8 and 0.85, in the manner described in Section 9.5.

One batch of preforms produced from the "as supplied" powder was then coated with Birkatekt and preheated to 900°C in an electric muffle furnace which was occasionally purged with nitrogen. They were then open die forged to 90%, 80%, 70%, 60% and 50% of their original heights, the reductions being controlled by the use of crash rings. The forging was carried out using the 3MN "Armstrong" Press, see Section 5.1, which involved charging the accumulators before each forging operation, and then releasing the charge directly into the press to produce an average ram speed of $\sim 50 \, \text{mm/s}$.

The hot axisymmetric upsetting tool used for these tests is shown in Fig. 10.1, together with one of the crash rings. The tooling consisted very simply of a standard die-set on to which two solid cylindrical anvils were bolted. These were made from "as supplied" W. L. Marisson's, chromium, hot working steel designated WLM 4052, which contained 0.4% C, 5% Cr, 1.4% Mo and 1% V. The crash rings were made from a ductile aluminium alloy. No lubrication was used for these tests, nor were the dies deliberately preheated.
FIG. 10.1  GENERAL ARRANGEMENT OF TOOLING FOR HOT AXISYMMETRIC
UPSET TESTS
Fig. 10.2  PARTLY FORGED TEST PIECES
After forging, the test-pieces shown in Fig. 10.2 were allowed to cool to room temperature. Their minimum and maximum diameters corresponding to Section X and Y, Fig. 9.5 were then measured together with their final lengths and densities, the latter being determined by an Archimedean technique. Although an attempt was made to measure the forging loads, using the load cell arrangement and U.V. recorder described in Appendix E, the results were somewhat erratic, and therefore disappointing. However, such results can be found well documented in other sources. The reason for the erratic traces obtained from the U.V. recorder was due to the fact that the forces involved consist of a steady state hydraulic force, plus an inertia force, due to the initial acceleration of the unrestrained ram, together with the added complication of stress wave propagation in an elastic solid. The latter occurs at the speed of sound for the substance \( \sqrt{E/\rho} \) which for steel is \( \sim 16000 \) km/h.

Nevertheless, it was observed that the initial impact force increased as the amount of free ram movement, prior to the punch making contact with the preform, increased, whereas during the actual free upsetting operation the load tended to gradually diminish until contact was made with the crash rings.

Similar forging tests, conducted with the use of the lubricant Delta-forge 528, were then repeated for another two batches of preforms. One batch was manufactured from the "as supplied" powder, and the other from the "coarse fraction". Delta-forge 528, which was essentially a suspension of graphite in water, was used in the first instance to coat the preforms. This was achieved by preheating the preforms to 175°C followed by immersing them in the lubricant. The second way in which the Delta-forge 528 was used, was by spraying it directly on to
the forging tools immediately prior to forging. The forging test procedure and subsequent measurement of the partly forged preforms was then carried out as for the initial unlubricated series of tests.

10.2 Interpretation of Results of Hot Axisymmetric Upset Tests

Fig. 10.3 shows the experimental results obtained from the unlubricated series of tests conducted on the preforms manufactured from the "as supplied" powder. Fig. 10.4 shows the results obtained from the tests carried out with lubrication, on the preforms manufactured from the "coarse fraction" powder sample. Since these are almost identical to the results obtained from the tests conducted on the "as supplied" powder preforms with lubrication, then the latter have been omitted. In all such tests the densification is non-uniform, and complies with the pattern illustrated schematically in Fig. 9.1 As a matter of interest, this may be compared with the preform density distribution shown in Fig. 5.18 for an initial overall relative density of $\sim 0.85$.

The slopes of the curves shown in Figs. 10.3 and 10.4 provide a quantitative assessment of the instantaneous "densification rate" for the preforms during the upsetting stage of the powder forging process. The initial slopes of these curves were obtained by determining the density corresponding to $\varepsilon_z = 0.1$, and then subtracting the initial density and dividing the difference by 0.1 to give the experimentally determined value for $\frac{d\rho}{d\varepsilon_z}$. This value corresponds to the initial preform density, which as indicated by Fig. 5.18 is virtually uniformly distributed. These values were then compared with those calculated from equation (50), which was derived by Griffiths et al.¹⁸ for homogenous compression, viz.

$$\frac{d\rho}{d\varepsilon_z} = \rho_0 (1 - \rho^2) \quad \text{----------------------} \quad (50)$$
FIG. 10.3 VARIATION OF RELATIVE DENSITY WITH LONGITUDINAL STRAIN FOR UNLUBRICATED HOT AXISYMMETRIC UPSETTING OF CYLINDRICAL PERFORMS, MANUFACTURED FROM "AS SUPPLIED" POWDER
Variation of relative density with longitudinal strain for lubricated hot axisymmetric upsetting of cylindrical preforms manufactured from "coarse fraction" powder.
FIG. 10.5 COMPARISON OF EXPERIMENTAL RESULTS WITH THOSE GAINED
BY THE EQUATION, \( \frac{d \rho_*}{d \varepsilon_z} = \rho_* (1 - \rho_*^2) \)
Fig. 10.5 shows that the experimental results agree favourably with the curve drawn to represent equation (50).

It was found that the critical diameter ratio at which peripheral cracking occurred in some of the preforms subjected to 50% height reductions was √1.35, which agreed well with previously published work. In addition, Figs. 10.6 and 10.7 were drawn, which show the relationship between radial strain and longitudinal strain at Sections X and Y, Fig. 9.5, for preforms of maximum and minimum initial densities. For this purpose, the overall longitudinal strain was regarded as being representative of the longitudinal strain occurring at both Sections X and Y, Fig. 9.5. It can be seen that at Section X, i.e. the die/preform interface, the slopes of the graphs, and hence the values of √ ap, are continuously increasing and tending towards the ideal plastic Poisson's ratio of 0.5, as the peripheral material approaches full densification in this region, see Fig. 9.1. However, at Section Y, which corresponds to the mid-height of the preform, the slopes are constant, and their relatively high values, i.e. approaching 0.5 indicate that in this region, the material is deforming with very little change in overall volume. From this it follows that very little densification takes place in this outer peripheral region, as indicated in Fig. 9.1, and it is likely that the circumferential stresses in this region may worsen the situation, and actually bring about a reduction in the relative density of the material.

A more rigorous and complete analysis of the hot, free upsetting process is difficult, because of the non-uniform density distribution which exists throughout the partly forged preform. However, some analytical work dealing with this problem has been carried out by Dorofeev and Koval'chenko.
FIG. 106  VARIATION OF RADIAL STRAIN WITH LONGITUDINAL STRAIN AT VARIOUS SECTIONS OF PREFORMS OF DIFFERENT DENSITIES DURING HOT UNLUBRICATED AXI_SYMMETRIC COMPRESSION. PREFORMS MANUFACTURED FROM "AS SUPPLIED" POWDER

FOR A PREFORM OF INITIAL DENSITY $\rho_*=0.7$

- $\varepsilon_{ry}$  
- $\varepsilon_{rx}$

FOR A PREFORM OF INITIAL DENSITY $\rho_*=0.832$

- $\triangle$ $\varepsilon_{ry}$  
- $\square$ $\varepsilon_{rx}$

LONGITUDINAL STRAIN $\varepsilon_z$

RADIAL STRAIN $\varepsilon_r$
FIG. 10.7

VARIATION OF RADIAL STRAIN WITH LONGITUDINAL STRAIN AT VARIOUS
SECTIONS OF PREFORMS OF DIFFERENT DENSITIES DURING HOT
LUBRICATED AXISYMMETRIC COMPRESSION. PREFORMS
MANUFACTURED FROM "COARSE FRACTION" POWDER

FOR A PREFORM OF INITIAL DENSITY $\rho_0 = 0.69$
--- $\varepsilon_{ry}$
--- $\varepsilon_{rx}$

FOR A PREFORM OF INITIAL DENSITY $\rho_0 = 0.837$
--- $\Delta - \varepsilon_{ry}$
--- $\square - \varepsilon_{rx}$
10.3 Hot Closed-Die Forging of Cylindrical Preforms

The preforms used for these tests were made from the "as supplied" and "coarse fraction" powders, using the same powder blends and manufacturing techniques as described in Section 10.1. They were produced to the same initial densities as used for the previous forging tests. Although the diameters of all these preforms were kept constant at 40 mm, their lengths varied between approximately 30 mm and 37 mm in order to ensure that the mass of each preform was maintained at a constant 250 g.

The closed-die hot forging tool used for these tests can be seen in Fig. 10.8. It consisted of a standard die-set fitted with new pillars measuring approximately 130 mm longer than the original pillars. The tool was designed to withstand forces of up to 2 MN. The backing plates for both the die and punch were made from "as supplied" WLM 4052, whilst mild steel was used for the shrouds for both the punch and die insert. The material used to make these latter components was heat treated WLM 4052, with the method of heat treatment being as follows. Firstly the parts were hardened by preheating to 800/850°C, holding at this temperature for 2 hours followed by increasing the temperature to 990/1020°C, holding for 1 hour and quenching in oil without agitation. The punch and die insert were then tempered by heating to 550°C, holding at this temperature for 2 hours and then cooling in air. This latter procedure was repeated a second time with the result that the final hardness values were 50 Re. The heat treated parts were then ground to their finished dimensions, which included a 3° draft angle on the die recess and punch, to facilitate ease of removal of the component after forging.

The die insert was shrunk into its shroud with the use of liquid nitrogen.
The forging operation was again carried out using the "Armstrong" 3 MN press in the manner described in Section 10.1. The preforms were coated with Delta-forge 528 in the manner described earlier, and preheated to 900°C prior to forging. The forging tool was also sprayed with Delta-forge 528, to reduce friction. Although the tools were not preheated at the beginning to the tests, they soon acquired a considerable amount of heat from forgings during the dwell period prior to removal of the finished forging. The finished forgings measured ~ 19 mm long with an average diameter of ~ 46 mm. The residual porosity ranged between 0.7% and 1.5% whilst the maximum forging loads involved were ~ 1.5 MN. In this case it was possible to successfully measure these loads with the load cell and U.V. recorder. The corresponding forging pressures were therefore ~ 940 MN/m$^2$, which lies within the range of values quoted in Section 2.13, viz. 800 to 1200 MN/m$^2$. Finally, there were no discernible differences between the forgings produced from the "as supplied" as compared to the "coarse fraction" powders.
CHAPTER 11

Summary of Conclusions and Future Work

11.1 Control of the Process

The success of powder forging as a means of producing high integrity precision forgings depends very much upon the amount of control that one can exercise over the various operations involved. This is very true of the actual forging operation where rapid, uniform heating of the preform and precise temperature control of both the preform and the tool is imperative to the success of the operation. Quite obviously the economic viability of the process depends upon the virtual elimination of any post-forging finishing operations, so that the forging tools have to be designed very carefully to ensure that the forging is produced with sufficient accuracy to obviate the need for subsequent machining. Although some dilatometer tests were conducted in order to provide basic information, the final tool design is very much a matter of experience and trial and error with known, controllable conditions.

Another important factor is that of controlling the mass of the preform. Since powder forging is essentially a closed die, flashless operation, an overweight preform could cause damage to the tooling. Equally so, an underweight preform will lead to regions of residual porosity, which may result in premature failure.

11.2 Advantages of Using "Coarse Fraction" Powders

A point of considerable interest which has emerged from this project has been that of the enhanced mechanical properties obtained with mixes of "coarse fraction" powders as compared to "as supplied" powders, sintered under identical conditions. The reason for this would
seem to be that when sintered in accordance with standard industrial practice, it is easier for the graphite to go into solid solution in the "coarse fraction" powder, hence resulting in better homogenization and less free cementite. It appears that the only advantage of the "as supplied" powder is that of marginally better compactibility, which is only of importance in the manufacture of conventional PM structural parts. It therefore seems likely that powders of coarser particle size could be used to advantage in powder forging, and have the added attraction of being cheaper than conventional powder blends.

11.5 OptimumForging Temperature

The results of the "forgeability" tests viz., hot tensile and hot torsion tests suggest optimum temperatures at which the preform material is in its most amenable state for forging. However, they also indicate that for iron/graphite powder mixes, this temperature depends upon the quantity of carbon contained in the preform material so that the higher the carbon content, the lower the forging temperature.

11.4 Appraisal of Preform Material

An indication of the behaviour of the preform matrix material when subjected to uniaxial stress, was obtained by measuring the corresponding longitudinal and lateral strains. From this one was able to conclude that the matrix material behaves in a very ductile manner.

This important conclusion enabled a mathematical model to be devised to represent porous PM materials subjected to tensile tests to destruction. The model takes into account the strength and toughness of the matrix material as well as the distribution and morphology of the cavities which make up the porosity. Such a model helps to clear up much of the confusion that has persisted for so long in trying to explain
the variations in tensile strengths of porous materials based entirely, but erroneously, on considerations of the amount of porosity only. With the aid of this model it is possible to gain a measure of the strength of the pore free matrix material as given by $\sigma_0$, and also to quantify the effects of the porosity as given by $\lambda$.

11.5 Preform Design

The pilot study of preform design undertaken in collaboration with South WalesForgemasters Ltd., Cardiff, served to indicate the need for careful distribution of the material in the design of the preform. This problem was largely overcome by the mathematical technique described in Appendix D, based on mass distribution diagrams. It is considered that such a technique could overcome much of the costly trial and error methods that have been practised in the past, and could also prove useful to analyse existing successful preform designs. The study also served to reveal some of the problems associated with powder forging viz. peripheral cracking, air entrapment, backward extrusion of flash and inverse barrelling. Additionally it emphasised the need to preheat the lower die and reduce contact times in order to minimize the effects of chilling. Furthermore, there may be some advantages in encouraging radial inward flow of material during forging, to help reduce peripheral cracking.

11.6 Designing Powder Forged Components

In the event of some residual porosity being an inevitability, then every care should be taken to ensure that it is strategically arranged to occur in low stressed regions. The main danger with such partially closed porosity is that it may exhibit mechanical properties that are inferior to the original preform material. The reason for
this is that whereas the porosity of the preform material tends to be fairly well rounded as a result of sintering, partial deformation of this porosity may worsen its morphology thereby increasing the $K_t$ values, and bringing about inferior mechanical properties. Although some fatigue tests were conducted in an attempt to investigate this problem, unfortunately, it was later realised that the method of fatigue testing selected was not entirely suitable.

When designing a component for production by powder forging the important criterion is whether the powder forged material is adequate for the task rather than whether the material is as good as conventionally forged wrought material. In many cases it may be that the quality of the latter is far in excess of the minimum requirement necessary to ensure the satisfactory functioning of the component, see Appendix A.

11.7 Analysis of Consolidation

The mathematical model devised to represent the tensile strength of the preform material was used to provide a philosophical, if not rigorously scientific approach to an explanation of the consolidation process. Furthermore, the introduction of the concept of a "coefficient of consolidation" with universal application proved useful, since it provided a quantitative measure of the manner in which porous materials consolidate, for a given set of conditions. It can also be used to derive compatibility equations to describe the changing geometry of PM preforms during the upsetting stage of the powder forging operation. In addition, these equations have wider general application than previously derived compatibility equations, Appendix B, which simply apply to special cases based on idealised conditions.

It was found that the apparent plastic Poisson's ratio ($\nu_{ap}$)
is dependent upon initial preform density. This is to be expected since, as for tensile strength, $\sqrt[\alpha]{p}$ will be influenced not only by the amount of porosity present, but also by its distribution, morphology and orientation.

11.8 Flow Stress, Density Distributions and Other Considerations

Owing to the problems involved in attempting to provide a rigorous analysis of the plastic flow stress requirements, a method of interpolation is suggested, which could prove useful in a real situation.

Although density distributions are successfully determined by means of hardness tests, Section 5.6 and sectioning and weighing, Section 8.5, another method of assessment is used on the hot forged specimen. This entails the measurement of the apparent plastic Poisson's ratio as given by the slope of the graph of lateral strain vs. longitudinal strain. The results of such tests confirm what is generally known about the non-uniform density distribution which occurs during axisymmetric upsetting of porous preforms. Furthermore, the "densification rates" obtained from the hot upset tests agreed favourably with the theoretically predicted values, whilst the forging pressures observed were within the expected range as reported in other sources. There was no discernible difference in the forging performance of preforms made from "coarse fraction" powder as compared to those produced from "as supplied" powder.

11.9 Suggested Future Work

As with all undertakings of this nature this thesis marks the beginning rather than the end of an on-going programme of work. Since powder forging is essentially a post-sintering operation it has been necessary to briefly investigate many of the operations normally
associated with conventional PM processing, in order to be able to make a worthwhile appraisal of the preform material. As a result of this, much has been learned which is directly applicable to structural PM material as well as powder forging. Suggested areas of work considered suitable for further investigation are therefore as follows.

(i) Mechanical testing of partially deformed preform material to determine the harmful effects that the distorted pore morphology may have on the mechanical properties of the material. In particular to carry out a systematic study of this effect on the fatigue response of partly forged materials in an attempt to simulate the conditions likely to exist in isolated pockets of porosity contained in powder forged components.

(ii) Compare the mechanical properties of powder-forged components made from commercially available powder blends with those produced from coarse particle powders, since there may be definite economic advantages in using the latter type of powder.

(iii) Carry out further consolidation tests involving the use of teflon sheets to minimize friction, and obtain representative values for C (coefficient of consolidation) for different modes of deformation.

(iv) Investigate the effect of powder particle size on the diffusion rate of graphite in elemental iron powder/graphite mixes with a view to establishing the degree of homogenization that can be obtained when sintering in accordance with standard industrial practice.

(v) Carry out a series of tests on PM material of different relative densities, to measure the longitudinal and lateral strain components arising due to uniaxial stress. The purpose of these tests will be to gain a better understanding of the behaviour of the matrix material
during tensile testing.

(vi) Re-appraisal of the results of the transverse rupture test to establish the connection between the "modulus of rupture" and the tensile strength. Since the matrix material has been shown to behave in a ductile manner it is considered that this problem can be approached by assuming that the tensile portion of the beam becomes completely plastic before the onset of fracture. Furthermore, there may even be a slight shift in the neutral axis towards the compressive side of the beam.
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Abstract

This is a report of the impressions gained after undertaking a three month study of powder metallurgy in Europe, with the aid of a Goldsmiths' Travelling Fellowship. The period of the tour was from approximately mid-April to mid-July 1975. Its purpose was to gain a first-hand insight into the "state of the art" of technology and education in the field of powder metallurgy as far as was possible within the limited time available.

An initial approach was to send requests to considerably more establishments than could possibly be visited in the time available. These tactics yielded dividends, since only about 50% of the establishments contacted were actually favourable to a visit. Of the remainder, some were seemingly ignored, inasmuch as no replies were forthcoming, whilst in other instances initially convenient dates were not available. Others were misunderstood, or else firms were too modest about their activities, and felt that to visit them would serve no useful purpose. A number of firms wrote courteous replies expressing regret that they were unable to accept visitors on the grounds that either they considered their processes as being confidential or that it was not company policy.

In formulating the list of establishments for visits, the main sources of reference were "Lists of Delegates", from annual meetings of the Powder Metallurgy Joint Group held over the past few years, lists of members kindly made readily available by the B.S.M.A. and the 1974 H.M.S.O. publication "Scientific Research in British Universities and Colleges".

Itinerary for the three month tour was as follows:

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Not from the obvious benefits derived from the technical information received during visits, the most important benefit was the opportunity for hand discussion with people from almost every aspect of powder metallurgy. Despite the fact that this report is mainly one of impressions, every attempt has been made to present these as objectively as possible.

It must be stated that it is virtually impossible to gain a complete light of PM activity due to complications arising from a multiplicity of commercial interests, which understandably make people guarded about the amount of information that they are prepared to divulge. In the absence of Universities, Polytechnics and research establishments not involved in contract research, this wariness was not in evidence and was quite evident for detailed discussions of their activities. There are always some instances of firms quite openly stating that they welcome visits as a means of advertising. It was felt that this attitude was justified.

During the middle and latter part of the tour it was possible to collect first-hand information regarding what services were available from sources previously visited. In fact, one of the most noticeable features as the tour progressed was that of a greater interchange of information, whereas during the earlier stages this had been predominantly one way.

Discussions were usually regarding the state of the PM industry, its prospects and potential as a route to the manufacture of components, and ways and means of promoting its best interests. In addition, there was a number of newly developed processes and ideas were encountered which the various establishments concerned were anxious to talk about and promote, whilst there was often little mention of other techniques, which were quickly passed over, as they obviously considered that there was some future commercial potential. In these latter discussions and the earlier reluctance of some firms to receive visitors, it soon became apparent that no single person is likely to know the full sum total of knowledge that exists in the field. Even by keeping abreast of all published work, he is still likely many years behind the true frontiers of knowledge in many aspects he subject.

However, it is fully realised that the PM industry is still small and likely new, and despite claims that it has approximately a 15% share of the market, the existing market is probably still too small to enable the industry to utilise its total capacity. The result is that the reputation for orders is extremely keen. Nevertheless it is encouraging to see that the industry is gradually beginning to diversify its activities, so that it is becoming progressively less dependent upon the automotive industry. This would appear to be essential for its future being, since in the present circumstances any recession in the motoring industry would have serious repercussions in the PM industry. This latter approach is obviously aimed at the existing generation of designers and mechanical and production engineers, in an attempt to familiarise these people with the advantages offered by PM.

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The ultimate factor in deciding whether or not a part may be produced by PM depends very much upon the skill and ingenuity of the designer. There can be no doubt that the PM industry is quite firmly established and forms an indispensable part of everyday living. It only requires a few moments' reflection to realise that whether at home, in the office, on the factory floor or driving along in a motor car, use is being made of parts. In addition to the admirable way in which PM manages to cope with these more mundane tasks, it may also be regarded as a technology for the future, being capable of performing functions and producing materials for which as yet there may not be the necessary applications. Situations personally encountered where PM is found on the threshold of technology and science, include proposals to produce superalloy turbine discs by the PM route, production of alloys for nuclear power applications and prosthetic devices for use in orthopaedic surgery. Naturally, in a future where growing emphasis will be placed on conservation of resources, it would seem reasonable to expect that PM will find increasingly greater favour, due to its maximum material utilisation and the resulting energy saving from not having to cycle scrap material.

Acknowledgements
The writer wishes to acknowledge grateful thanks to Dr. D. W. F. James, Director of The Polytechnic of Wales for kindly allowing him three months leave of absence to conduct the study tour and also to the Goldsmiths' Company of London for their confidence and generosity in awarding the financial support for the venture.

The writer also wishes to take this opportunity to express his most sincere gratitude to all those establishments who were kind enough to receive him for a visit. In all cases he was extremely well received and courteously treated, and hopes that the valued contacts and acquaintances made will continue long into the future.

References
2. Fulmer Research Optimiser, Fulmer Research Institute Ltd., Stoke Poges, Slough, Bucks.
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REFERENCES
Compatibility equations for the powder-forging process

T. J. Griffiths, R. Davies, and M. B. Bassett

The three principal modes of deformation encountered during powder forging are (i) axisymmetric compression, (ii) plane-strain compression, and (iii) re-pressing. To establish compatibility or continuity equations to describe the geometry of deformation of porous preforms during each of these processes, the authors resort to the artifice of assuming an apparent plastic Poisson's ratio \( \nu_p \), which experimental observation suggests is a function of the relative density \( \rho_r \) of the preform. However, since the rate of densification is dependent upon the degree of lateral restraint imposed by the mode of deformation, the empirical relationships connecting \( \nu_p \) and \( \rho_r \) will vary accordingly. Compatibility equations are therefore derived for each mode of deformation, and their predictions compared against a selection of previously published experimental data.

Powder forging is currently arousing interest in many parts of the world as an economic method of producing high-strength, high-ductility steel parts from metal powders.1

The process involves the forging of a well-defined preform to produce a component with negligible porosity, and mechanical properties which compare favourably with those of conventional forgings. Careful distribution of the metal in the preform is necessary to minimize unsupported lateral flow and the accompanying risk of cracking during the forging operation. The preform may be produced by either mechanical or isostatic compaction, followed by sintering, generally in an endothermic atmosphere. After sintering, the preforms may be either cooled, stored and then reheated or transferred directly from the sintering furnace to the forging press.

One of the many advantages of the process is that the well-defined preform shape obviates the need for multi-impression dies,2 and enables the component to be produced in a single blow.

The important difference between powder forging and the conventional forging of wrought material is that of consolidation of the PM preform. Powder forging would therefore seem to be a hot-compaction process with some bulk metal plastic flow.3

The three principal modes of deformation that can occur during powder forging are: (i) axisymmetric compression, i.e., no lateral constraint; (ii) plane-strain compression, i.e. restraint in one lateral direction; and (iii) re-pressing, i.e. total constraint.

For the initial part of this study it has been assumed that frictionless, homogeneous compression occurs throughout, so that there is no barrelling, and where applicable, all points on the periphery of the preform make contact with the die walls simultaneously. Although this may not be strictly true in practice, the amount of unsupported lateral flow is generally so small that this assumption will not introduce appreciable errors. As a further consequence of homogeneous compression it seems reasonable to assume an average density for the preforms as given by \( \rho_r \), which is the ratio of actual density to solid density, so that the porosity is given by \( 1 - \rho_r \).

The approximate density changes which take place with a typical commercial iron powder during the various stages in the production of a powder forging are summarized in Table I.

### Table I

<table>
<thead>
<tr>
<th>Condition</th>
<th>( \rho_r )</th>
<th>Remarks</th>
</tr>
</thead>
<tbody>
<tr>
<td>(i) Loose powder as supplied</td>
<td>0.32</td>
<td>Manufacturer's data</td>
</tr>
<tr>
<td>(ii) Preform after compaction</td>
<td>0.72–0.83</td>
<td>Cull4</td>
</tr>
<tr>
<td>(iii) Preform after upsetting</td>
<td>0.95</td>
<td></td>
</tr>
<tr>
<td>(iv) Component</td>
<td>1</td>
<td></td>
</tr>
</tbody>
</table>

### INITIAL CONSIDERATIONS

Eudier5 suggests that a fairly dense porous material, i.e. \( \rho_r = 0.9 \), may be represented as a solid containing spherical pores arranged in a simple cubic pattern. Speculation on what is likely to happen to these pores when subjected to the three principal modes of deformation could possibly explain the fundamental difference in the deformation characteristics brought about by each of these operations.

(i) With axisymmetric compression, i.e. triaxial strain, the predominant reduction in pore diameter takes place in the direction of compression and is accompanied by spreading of the pore laterally in all directions, thereby producing a very low rate of densification.

(ii) In the case of plane-strain compression, i.e. biaxial strain, the predominant reduction of pore diameter will again take place in the direction of compression, but the lateral spread can now take place in one direction only. This partial restriction in lateral spread brings about an enhanced rate of densification as compared with the axisymmetric case.

(iii) With re-pressing, i.e. uniaxial strain, the diameters of the pores decrease in the direction of compression while their lateral diameters remain constant, resulting in relatively rapid rates of densification. This agrees with Antes's experimental conclusions on p. 172 of his paper.
Nomenclature

- \( \nu \): Poisson's ratio for a perfectly plastic solid metal
- \( \nu_a \): apparent plastic Poisson's ratio
- \( \nu_{ap} \): apparent plastic Poisson's ratio for axisymmetric compression
- \( \nu_{ap} \): apparent plastic Poisson’s ratio for plane strain compression
- \( \rho_s \): density of solid metal
- \( \rho \): density of preform material
- \( \rho_* \): relative density of preform material. Subscripts 1, 2, and 3 are used to denote initial, intermediate/final, and final conditions, respectively
- \( m \): preform mass
- \( h \): preform height
- \( d \): preform diameter—axisymmetric compression
- \( l \): preform length—plane-strain compression
- \( V_s \): volume of solid material
- \( \epsilon_z \): principal natural strain in axial direction
- \( \epsilon_r \): principal natural strain in radial direction (axisymmetric compression)
- \( \epsilon_x \): principal natural strain in lateral direction (plane-strain compression)

By definition:

- \( \de_z = \frac{dh}{h} \)
- \( \de_r = \frac{dd}{d} \)
- \( \de_x = \frac{dl}{l} \)

In a real situation involving up to \( \sim 30\% \) porosity, it is obviously difficult to generalize regarding the mode of deformation of the pores, owing to their complexity of shape and orientation. Eudier suggests that the fairly flat pores would tend to collapse the most readily, and states that he has obtained micrographs which seem to confirm this hypothesis. This would certainly account for the higher consolidation rates experienced with the more porous preforms. It is also possible that as these flat voids collapse at their centres, smaller voids may be set up at their peripheries which, depending upon their shape, may then spread until they collapse to form even smaller pores. It can only be assumed that this process is repeated until, at some higher density, all the remaining voids are approximately spherical in shape; whereupon deformation takes place in the manner described earlier, assisted by the shearing effect which accompanies the lateral spread as the strain increases.

AXISYMMETRIC FRICTIONLESS OPEN-DIE COMPRESSION AND RE-PRESSING

Both the above operations occur during the closed-die forging of PM preforms, thereby subdividing the process into two distinct stages. For added simplicity it is proposed to focus attention on cylindrical preforms only.

Figure 1(a) shows the initial closed-die upsetting stage which involves consolidation and unsupported lateral flow of the material to fill the die cavity. Since this stage is assumed to be frictionless, it will be analogous in all respects to an open-die process. Figure 1(b) shows the re-pressing stage, which occurs after the lateral flow of the first stage is checked by the material making contact with the die walls. Compatibility or continuity equations to describe the geometry of deformation of the preforms during these two stages may now be determined from the following considerations.

Since the PM material may be regarded as a two-phase material consisting of the matrix metal and pores, it follows that the axial true strain will comprise two components, one attributable to consolidation, and the other to the lateral spread of the solid material. This may be demonstrated for the axisymmetrically loaded cylindrical preform shown in Fig. 2, as follows. For mass constancy,

\[
m = \pi d^2 h \rho_s = \frac{\pi}{4} (d + \delta d)^2 (h - \delta h) (\rho + \delta \rho)
\]

In the limit, after expanding and ignoring second-order infinitesimals, this becomes

\[
\frac{\de_z}{\rho_*} = \frac{\delta \rho_s}{\rho_*} + 2 \de_r \Rightarrow \frac{\de_z}{\de_r} = 0.5 = \nu
\]

The well-known fact that Poisson's ratio \( (\nu) = 0.5 \) for a perfectly plastic solid material can now be readily shown, since volume constancy prevails when \( \rho_s = 1 \) and \( \delta \rho_s = 0 \), i.e. there is no consolidation and eqn. (1) becomes

\[
\frac{\de_x}{\de_z} = 0.5 = \nu
\]
For re-pressing, the process is that of consolidation only when
\[ -\varepsilon_z = \frac{d\rho_s}{\rho_s} \]  \hspace{1cm} (3)

It would appear from experimental observations\(^1\) that it is the extent of this porosity which dictates the relative rates at which consolidation and plastic flow take place. Such observations suggest that a high degree of porosity favours rapid consolidation, which gradually diminishes in favour of increasing amounts of lateral spread as \( \rho_s \to 1 \). This necessitates having to resort to the artifice of assuming an apparent plastic Poisson's ratio \( \nu_{aa} \) to describe the mode of deformation. By transposition, eqn. (1) becomes
\[ -\varepsilon_z = \frac{d\rho_s}{\rho_s (1 - 2\nu_{aa})} \]  \hspace{1cm} (4)

Clearly, if it can be established that \( \nu_{aa} = f(\rho_s) \), then the above expression can be integrated to provide the compatibility equation for the closed-die upsetting stage. Many experimenters confirm that for axisymmetric conditions such a function exists, and that the value for Poisson's ratio corresponding to a particular density is independent of the means by which that density was achieved.

Marx and Davies\(^9\) suggest that \( \nu_{aa} \) is linearly dependent on the density, being of the general form
\[ \nu_{aa} = kp_s - C \]  \hspace{1cm} (5)

where \( k \) and \( C \) are constants. Unfortunately, these constants cannot be deduced analytically, and the empirical relationship obtained from their experimental observations of an Fe-base powder over the density range \( \rho_s = 0.84 \to 1 \) is
\[ \nu_{aa} = 0.93\rho_s - 0.43 \]  \hspace{1cm} (6)

These results were obtained from a series of Cooke and Larke-type tests, in which frictionless conditions were simulated.

Further experimental data are provided by Kuhn and Downey,\(^10\) who propose that the relationship over what appears to be a similar density range is
\[ \nu_{aa} = 0.5\rho_s^2 \]  \hspace{1cm} (7)

Although essentially intended to refer to an aluminium powder, it is nevertheless stated that in their experience it could apply equally well to ferrous powders. Friction was eliminated in this case by the use of Teflon sheets inserted between the dies and the preforms.

Figure 3 shows that the results of eqns. (6) and (7) compare very favourably over the stated density range; however, it has been necessary to extrapolate to \( \rho_s = 0.72 \) (see Table 1) to cater for the increased density range encountered in a real situation. Coupled to this is the fact that frictionless conditions will not exist in practice, despite the use of die lubrication. Further tests conducted by Marx and Davies, with lubrication, suggest the relationship
\[ \nu_{aa} = 0.25 \rho_s \]  \hspace{1cm} (8)

When plotted on Fig. 3, it shows that the effect of friction is to reduce the instantaneous values for \( \nu_{aa} \) corresponding to any given density. This has arisen from the reduced lateral spread and the accompanying increase in the rate of consolidation.

The compatibility equations needed to describe the geometry of deformation of PM preforms, during the initial closed-die axisymmetrical upsetting stage of the powder-forging process, can now be obtained as follows:

\[ -\varepsilon_z = \frac{dp_s}{\rho_s} + de_x \]  \hspace{1cm} (15)

---

1. Frictionless conditions
(a) Combining eqns. (4) and (6) and integrating gives
\[ -\varepsilon_z = 0.54 \ln \left( \frac{\rho_s}{\rho_s} \right) \]  \hspace{1cm} (9)

Similarly from eqns. (1) and (6)
\[ \varepsilon_f = 0.27 \ln \left( \frac{1 - \rho_s}{1 - \rho_s} \right) \]  \hspace{1cm} (10)

(b) Alternatively, combining eqns. (4) and (7) and integrating gives
\[ -\varepsilon_z = \ln \left( \frac{\rho_s}{\rho_s} \right) + 0.5 \ln \left( \frac{1 - \rho_s}{1 - \rho_s} \right) \]  \hspace{1cm} (11)

where \( \ln \left( \rho_s/\rho_s \right) \) is the strain component attributable to consolidation, and \( 0.5 \ln \left( 1 - \rho_s/1 - \rho_s \right) \) is the strain component attributable to plastic deformation.

Also
\[ \varepsilon_f = 0.25 \ln \left( \frac{1 - \rho_s}{1 - \rho_s} \right) \]  \hspace{1cm} (12)

2. With lubrication
Combining eqns. (4) and (8) and integrating gives
\[ -\varepsilon_z = 0.5 \ln \left( \frac{\rho_s}{\rho_s} \right) \]  \hspace{1cm} (13)

and
\[ \varepsilon_f = 0.25 \ln \left( \frac{\rho_s}{\rho_s} \right) \]  \hspace{1cm} (14)

---

**PLANE-STRAIN COMPRESSION**

It can readily be demonstrated for the plane-strain configuration shown in Fig. 4 that

\[ -\varepsilon_z = \frac{dp_s}{\rho_s} + de_x \]  \hspace{1cm} (15)
whilst the apparent plastic Poisson's ratio is now defined as
\[ \nu_{ap} = \frac{-\frac{d\varepsilon_y}{d\varepsilon_z}}{2d\varepsilon_z} \]

By substitution and transposition eqn. (15) becomes
\[ -d\varepsilon_z = \frac{d\rho_s}{\rho_s(1-2\nu_{ap})} \]

If recourse is made to experimental data previously published by Fischmeister et al., then an empirical relationship can be established between \( \nu_{ap} \) and \( \rho_s \) which can be used to derive the compatibility equation for unlubricated plane-strain compression. Table II is an extract from the experimental data obtained from tests conducted on ferrous powder preforms.

Figure 5 shows that a good correlation between these results is provided by the empirical equation
\[ \nu_{ap} = 0.5 \rho_s^3 \]

Statistical comparison between the experimental data and those calculated from eqn. (17) shows very good agreement at the 0-01 level of significance on the \( \chi^2 \) test.

Combining eqns. (16) and (17) and integrating gives
\[ \varepsilon_z = \ln \left( \frac{\rho_s^3 + \frac{1}{3} \ln \left( \frac{1 - \rho_s^3}{1 - \rho_s^3} \right)}{\rho_s} \right) \]

and hence
\[ \varepsilon_x = \frac{1}{3} \ln \left( \frac{1 - \rho_s^3}{1 - \rho_s^3} \right) \]

APPLICATIONS OF COMPATIBILITY EQUATIONS TO EXPERIMENTAL OBSERVATIONS

From either eqn. (4) or (16) it is now possible to provide a quantitative assessment of densification rates, since
\[ \frac{d\rho_s}{d\varepsilon_z} = \rho_s(1-2\nu_{ap}(1-\rho_s)) \]

(ignoring the negative sign). Hence, for frictionless axisymmetric compression, using eqn. (6),
\[ \frac{d\rho_s}{d\varepsilon_z} = 1.86\rho_s(1-\rho_s) \]

or, using eqn. (7),
\[ \frac{d\rho_s}{d\varepsilon_z} = \rho_s(1-\rho_s^2) \]

and for lubricated axisymmetric compression, using eqn. (8),
\[ \frac{d\rho_s}{d\varepsilon_z} = 2\rho_s(1-\rho_s) \]

For unlubricated plane-strain compression, using eqn. (17),
\[ \frac{d\rho_s}{d\varepsilon_z} = \rho_s(1-\rho_s^3) \]

### Table II

<table>
<thead>
<tr>
<th>Preform porosity</th>
<th>37.4%</th>
<th>32.7%</th>
<th>25.7%</th>
<th>19.8%</th>
<th>0%</th>
</tr>
</thead>
<tbody>
<tr>
<td>Initial apparent plastic Poisson's ratio</td>
<td>0.13</td>
<td>0.16</td>
<td>0.2</td>
<td>0.25</td>
<td>0.5</td>
</tr>
</tbody>
</table>

![Diagram](image.png)

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![Curve-fitting](image.png)

6 Curve-fitting of experimental results of apparent plastic Poisson's ratio vs. relative density for ferrous preforms subjected to unlubricated plane-strain compression.
For re-pressing, transposition of eqn. (3) gives
\[
\frac{dp}{de} = \rho_\ast
\]

For the purpose of comparison, the densification rates arising from the different processes are shown plotted in Fig. 6. The fact that the densification rates for axisymmetrical and plane-strain compression tend to converge at the higher densities (theoretically when \( \rho_\ast = 1 \)) is verified experimentally by Antes.\(^6\) He observed that at a preform density of \( 7.2 \text{ Mg/m}^3 \) the densification rates were approximately the same.

Figure 7 shows how the overall powder-forging process (illustrated in Fig. 1) may be represented by erecting an ordinate from one densification curve to the other, corresponding to the instantaneous density at which the preform material contacts the die walls. If, for example, the end of the upsetting stage occurs at a value of \( \rho_\ast = 0.9 \), for a preform of initial density \( \rho_\ast = 0.8 \), then the complete 'rate of densification' curve is traced out by \( a, b, c, d \) in Fig. 7. However, because of barrelling in a practical situation, it is likely that the 'real' curve is S-shaped, as shown dotted. It is also possible that during this transition period from the axisymmetric to the re-pressing stage, the material in contact with the die wall will tend to behave as if it were subjected to plane-strain conditions.

Before proceeding further it must be pointed out that the compatibility equations imply that as \( \rho_\ast \to 1, \varepsilon_\ast \to \infty \), thereby suggesting that it is impractical to achieve 100% consolidation by upsetting alone. Furthermore, Davies and Dixon\(^1\) show that for frictionless, axisymmetric compression, the situation is further aggravated by the fact that, as the preform spreads, the tensile stresses induced in its periphery have the effect of making it even less dense in this region. Their experimental curve of density vs. percentage reduction of height is shown in Fig. 8, where it is compared with the theoretical curve for frictionless conditions, since from eqn. (9)

\[
\text{Percentage reduction of height} = \left\{ 1 - \frac{\rho_\ast(1 - \rho_\ast)}{\rho_\ast (1 - \rho_\ast)} \right\}^{0.5} \times 100
\]

It can be seen from Fig. 8 that the two curves are in good agreement up to a density of \( \sim 7.5 \text{ Mg/m}^3 \), i.e. \( \rho_\ast = 0.96 \), at which point the phenomenon described by Davies and Dixon becomes significant.

Both stages of the closed-die powder-forging process can now be examined in detail, since for lubricated axisymmetric compression, as represented by eqn. (13),

\[
\text{Percentage reduction of height} = \left\{ 1 - \frac{\rho_\ast(1 - \rho_\ast)}{\rho_\ast (1 - \rho_\ast)} \right\}^{0.5} \times 100
\]
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Typical curves of relative density vs. percentage reduction of height, for powder-forging process.

while for the re-pressing stage, as represented by eqn. (3),

\[
\text{Percentage reduction of height} = \left(1 - \frac{\rho_f}{\rho_s}\right) \times 100
\]

The process as described by eqns. (26) and (28) is shown in Fig. 9, where it can be seen that the densification curves representing the two stages have opposite curvatures, with the point of inflection marking the beginning of the re-pressing stage, i.e. when it is assumed that the metal has made contact with the die walls.

Guest et al.11 have published comparable experimental curves, one of which is shown in Fig. 10, compared against its theoretical counterpart plotted from eqns. (27) and (28). The two curves are in good agreement over most of their lengths, the main discrepancies arising from the fact that some barrelling was inevitable in the experimental situation, despite the use of lubrication. Although Guest et al. state that the initial part of their experimental graph was linear, it should in fact have been curved to correspond with these theoretical predictions. However, the amount of curvature is very slight over such a small density range, i.e. \(\rho_s \approx 0.86-0.94\).

Figure 11 shows the theoretical relationship between axial true strain, \(\varepsilon_z\), and lateral true strain, \(\varepsilon_x\), for unlubricated plane-strain conditions. The curves were drawn using the initial preform porosities shown in Table II, and eqns. (18) and (19). They compare favourably with their equivalent experimental curves which appear as Fig. 4 in Fischmeister et al.8 Again, because of the very slight curvature present, the experimenters have stated that the lower parts of their graphs are linear, for which the values for \(\nu_{ap}\) are quoted as 0.43 and 0.45. These figures can be roughly checked by taking the average slopes of the curves over a suitable density range, in accordance with strains which approximate to those used in the experiments, i.e.

\[
\nu_{ap} \approx \frac{\Delta \varepsilon_x}{2 \Delta \varepsilon_z}
\]

On this basis, irrespective of which preform is selected, the average slopes are 0.428 and 0.452 for the relative density ranges 0.9-0.98 and 0.95-0.98, respectively.

CONCLUSIONS
Frictionless axisymmetric and unlubricated plane-strain compressions produce the least and greatest rates of densification, respectively, during the upsetting stage of the powder-forging process. The actual conditions encountered in practice will lie somewhere between these two extremes.

For instance, in plotting the theoretical curve shown in Fig. 10, eqn. (27) was used, which was derived from lubricated axisymmetric compression. However, no correction was made for barrelling, which was excessive in this case owing to the abnormally high diameter ratio of 1.35 as compared to the critical diameter ratio of \(\sim 1.40\),11 at which peripheral cracking occurs.

It is for this and other reasons, e.g. inefficient densification, etc., that the actual amounts of lateral spread used in practice are generally small, so that the compatibility equations could be applied to real situations with greater accuracy than Fig. 10 suggests. Even when a small amount of barrelling is present, the predictions of the equations will...
not be seriously affected, since, as the preform density increases, so does the rate of densification become progressively less dependent upon the mode of deformation.

In addition to predicting the changing geometry of PM preforms during powder forging, the compatibility equations can also be used for the purpose of summarizing and comparing experimental observations.

REFERENCES
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PILOT STUDY OF PREFORM DESIGN FOR SINTER FORGING*

T. J. Griffiths,† W. Jones,† M. Lundregan,‡ and M. B. Bassett†

An important difference between sinter forging and conventional forging is that of consolidation of the PM preform. It therefore follows that sinter forging would seem to be a hot-compaction process with some bulk metal plastic flow. The permitted amount of plastic flow is considerably less for a PM preform than for a preform produced from wrought material of equivalent composition, since PM preforms exhibit poor ductility and are therefore more prone to cracking. The purpose of this study is to produce some guidelines to assist in designing preforms for sinter forging and to attempt to minimize the 'trial and error' approach. This has been undertaken by the production and study of a component suitable for PM forging, which at present is produced by conventional forging.

The lateral flow of metal was observed and investigations undertaken as to how density variations across the preform might assist the consolidation process. An iron powder containing 1% graphite and 1% zinc stearate was used in the production of components.

Considerably more care must be taken with the design of blanks or preforms for sinter forging than with those used for the conventional flashless closed-die hot forging of wrought materials. One of the main reasons is that the PM preform displays a much smaller degree of plasticity at elevated temperatures than a solid metal of comparable composition. This means that the amount of lateral flow that can be tolerated during the initial stages of sinter forging is critical if peripheral cracking of the upset blank is to be prevented. Despite the fact that the process is termed 'closed-die forging', during the early stages it is essentially one of open-die forging. This results in the upsetting of the blank until lateral flow of the preform is checked by the material making intimate contact with the die walls.1

Another of the main differences between sinter forging and the forging of wrought metal is that of consolidation of the PM preforms, so that the process is really one of hot compaction with some plastic flow.

The purpose of this study was to investigate the phenomena peculiar to the sinter forging process, in order to produce some guidelines to

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assist in the design of PM preforms, since most published work in this
field seems to report the production of preforms by pure 'trial and error'.
For this purpose it was decided to adopt the approach outlined by
Bockstiegel and Olsen, by examining the feasibility of producing, by
sinter forging, a component that had already been successfully forged
from En8 wrought material. The cross-section of this component is
shown in Fig. 1(6).

In addition to an actual component some test-specimens were pro­
duced for a series of laboratory tests to provide additional information
for comparison.

Equipment

The preforms and test-specimens were compacted on a 3 MN (300 tonf)
hydraulic press. A 7.5 MN (750 tonf) Ajax mechanical press, operating
at 80 rev/min was used for the forging operations.

The preforms were sintered in a Carbolite electrically heated muffle
furnace, fitted with control equipment to govern both the rate of heating
and final temperature. Provision was made to introduce nitrogen
periodically into the furnace in an attempt to combat excessive de­
carburization. The test-specimens were sintered in an endothermic
atmosphere in a continuous belt-type sintering furnace under produc­
tion conditions, in accordance with standard industrial practice. The
specimens were tested on a motorized Hounsfield tensometer, fitted with
an automatic recorder. An electrically heated tube furnace was used
in conjunction with the tensometer. Further tests were carried out on
a Griffin and George dilatometer, in which the specimen was contained
in an electrically heated silica tube, surmounted by a dial test indicator
to measure thermal expansion, and fitted with a thermocouple to record
temperature.

Powder

Rospol MP32 iron powder was used throughout the tests, the (advised)
sintering conditions being 20 min at 1120°C in an endothermic atmo­
sphere, dew-point -1°C. It was premixed with 1% zinc stearate, to
provide lubrication during compaction, and 1% graphite, to simulate
the wrought material En8, from which the component had already been
successfully forged. The seemingly high quantity of graphite was con­
sidered necessary, since the improvised sintering technique employed in
the manufacture of the preforms made it uncertain how much associated
carbon would remain in the final microstructure.

It was assumed that the solid density of the powder would be that
of solid iron, viz. 7.8 Mg/m³. The apparent density of the loose powder
was stated as 2.5 Mg/m³, though after mixing it was found to be
3.3 Mg/m³. Values for \( \rho^* \) have been quoted throughout the tests, where \( \rho^* \) is simply the ratio of actual density to solid density, so that the porosity is given by \( 1 - \rho^* \).

**Initial Considerations in Design of Preforms**

With flashless forging of solid metals, the main consideration is that the mass of the blank should equal the mass of the component. In most cases a simple prism of metal is used, this being cut from bar stock. Generally, large amounts of plastic flow occur without the material cracking; complex sections are excluded here. Although in the case of PM forging the mass consideration is still valid, a second criterion occurs, which is a limitation on the amount of plastic flow that may be permitted to minimize tensile stresses and associated cracking. One solution is to produce a preform which is similar in shape to the final product so that the forging process consists of hot compaction only. Complex compacting dies are required to produce such preforms and subsequent material properties are probably not as good as when some plastic flow is permitted to occur.

In practice, a compromise must be sought. The compaction dies must be simplified, while permitting only a limited amount of plastic flow to enhance material properties. It is worth while to compare the two extremes in some detail to demonstrate the futility of endeavouring to evolve a PM preform shape from a typical blank used in conventional forging.

Since the component, blank and preform, are all solids of revolution, it is a relatively simple matter to construct mass-distribution diagrams. Such diagrams are extremely useful at all stages in providing a comparison of the mass distributions of preforms with the finished component.

For a solid of revolution, the mass of an elemental ring of material occurring at radius \( r \) is given by \( 2\pi\rho t r\, dr \), where \( \rho \) and \( t \) are the density and thickness, respectively. Note that \( \rho \), \( t \), and \( r \) are all variables; \( \rho \) depending upon the method of manufacture and \( t \) and \( r \) upon the geometry of the solid.

\[
\text{Total mass of solid} = 2\pi \int_0^r \rho t r\, dr
\]

From this it can be seen that the mass may be represented by the area under the mass-distribution diagram, in which values of \( 2\pi\rho t r\) are plotted on the ordinate scale and \( r \) on the abscissa. The mass of metal contained between any two radii can then be obtained from such a diagram by measuring the area between the radii in question with a planimeter. Furthermore, by superimposing the mass-distribution diagrams for both component and preform on one another, it is possible
FIG. 1. (a) Preform shape. Boss dia. 32 mm for location; dimension A depends upon preform density. (b) Section through component.

to locate the precise regions from which metal has been displaced, together with the actual quantities involved. Using this technique, a preform shape may be gradually refined until it eventually produces the desired result.

Forging from Wrought Materials

The required blank would probably be prepared from a 51 mm (2 in) gothic* section that after heating is upset to ~60 mm dia. to remove scale and facilitate reasonable location within the die. Such a blank would be 19 mm high so that its mass should be the 414 g necessary to produce the component. The mass-distribution diagram for such a cylinder is a triangle as shown by a full line in Fig. 2, where it can be seen superimposed on the mass-distribution diagram for the component, shown ‘chain-dotted’; in each case the density is assumed to be 7.8 Mg/m³. By measuring the difference in area of the two diagrams up to any radius, it is possible to determine the mass flow of metal across that radius; it is shown in Fig. 3 (full line) and is compared with the mass flow of a typical PM preform (‘chain-dotted’ line).

The large amounts of metal displacement necessary for conventional

* A gothic section resembles a lobed square with rounded corners, and is widely used throughout the forging industry. When upset, its cross-section approximates to a circle.
Fig. 2. Comparison of mass distribution of solid cylindrical blank with that of finished component. Metal displaced from region (a) makes up the deficiency at (b), the surplus spreading into (c), which is then displaced to make up the deficiency in region (d).

Fig. 3. Comparison of metal displacement resulting from conventional forging of solid material with that of the same component, sinter forged.

Forging are immediately apparent, and though these are readily achieved with wrought materials, it is extremely doubtful whether a sintered material could withstand such rigorous treatment.

Design of PM Preforms

The shape of a PM preform that will produce the component by hot compaction only can now be obtained, provided that a suitable density
for the preform has been decided. In this connection Cull\(^3\) suggests that the optimum density range is 5-6-6.5 Mg/m\(^3\) (\(\rho^* = 0.72-0.83\)).

However, because of the practical difficulties in producing the multi-level preform that would be required in this case, it would be more expedient to work near to the top end of this range to guarantee reasonable compaction in the least dense regions. As a first approximation it was decided to use a value of \(\rho^* = 0.83\) and to assume that the density distribution would be uniform across the section. Fig. 4 shows the resultant preform shape. This was obtained by dividing the component thickness by \(\rho^*\) to give the new thickness, which was then measured from a common datum to give one flat face, since in this instance the height/dia. ratio of the component was relatively small. The central portion must obviously be modified slightly to form a third concentric cylinder, or boss, of suitable diameter to provide die location. It is interesting to note that Marx \(\text{et al.}\)\(^4\) evolved a similarly shaped preform, by a process of trial and error, for a similarly shaped component.

The foregoing theoretical preform design procedure is based on purely geometrical considerations; its aim is to minimize lateral flow of material to prevent cracking. The object of subsequent compaction is to densify the material completely and thereby improve its mechanical properties. However, the success of a sinter forging ultimately depends upon how satisfactorily it performs in practice,\(^5\) and optimum mechanical properties can best be obtained if there is some metal flow. Since available equipment dictated that the preform would need to be produced by
Griffiths, Jones, Lundregan, and Bassett: Pilot

single-ended compaction, this obviously meant that greater porosity will exist in the central regions, owing to the differences in the compaction ratios possible. With these points in mind, it was decided to modify the 'hot-compacted' shape so that the outer and intermediate cylinders would merge into one and the preform shape would consist of a flange, with a relatively less dense, centrally situated boss (Fig. 1(a)). The tooling was designed to ensure that the resulting blank would be of generous proportions so that a variety of diameters (C) and thickness (B) could be obtained by subsequent machining.

**Preform Manufacture**

The blanks were produced on the hydraulic press at pressures of 210 N/mm² (13.58 tonf/in²), 280 N/mm² (18.11 tonf/in²), 350 N/mm² (22.63 tonf/in²), and 420 N/mm² (27.16 tonf/in²). The approximate weight of the blanks produced was 900 g, though the actual quantities of powder were introduced into the die on a constant-volume basis. The average densities of the blanks are shown in Fig. 5. Difficulty was
experienced with the production of blanks at 210 N/mm², owing to the boss separating from the flange as a result of poor compaction in this region.

The blanks were sintered in a muffle furnace that was occasionally purged with nitrogen. They were heated to 1130°C, soaked for 30 min, and then removed from the furnace and buried in sand to cool. The blanks did not appear to be excessively scaled (Fig. 6(a)), though a decarburized zone to a depth of ~3–5 mm was found. However, it was considered that the decarburization was of little consequence at this stage since machining was to follow. The blanks were machined to the required dimensions, employing their average densities as a basis for the production of preforms of equal mass. Despite the fact that four different compaction pressures were used, the results obtained (by the average density method) led to a mean value of 410 g, which was only 1% below the required value of 414 g, with a variation of ±2% about this mean (acceptable as a first approximation).

The diameter of the flange was varied so that the onset of peripheral cracking could be observed to gain an appreciation of the amount of lateral flow that can be tolerated during the forging operation.

**Forging Operation**

The forging dies were mounted in a 7.5 MN (750 tonf) Ajax press and, as a precaution against any damage that might result from overweight preforms, the press was set to produce components ~0.5 mm oversize in thickness. Consequently, very few of the forgings completely filled the die, but on average were several millimetres too small in diameter. However, this procedure had the advantage of revealing peripheral cracks more readily than if the material had contacted the die walls.

The preforms were coated with Birkatekt to prevent scaling during preheating, which was carried out for 20 min at 900°C in a small electric...
furnace. Attempts were made to preheat the lower die by placing red-hot billets of solid metal on it.

The dies were lubricated with Rocol J166 copper-based lubricant, after which the blanks were forged and then quickly transferred to cooling oil to prevent excessive scaling. Fig. 6 shows the three stages in the manufacture of the forging.

**Associated Tests**

**Hot Tensile Tests**

During the early stages of the forging process the unsupported peripheral surface of the preform is subjected to circumferential tensile stresses that, if allowed to become excessive, result in peripheral cracking. It was not envisaged that any direct correlation between these tests and the actual forging operation would be possible, because of the difference in strain rates involved, the tests being conducted at a strain rate of only $0.046 \text{ s}^{-1}$, compared with $10-100 \text{ s}^{-1}$ for forging. The specimens were machined to size after compaction and sintering. A range of compaction pressures was investigated but the sintering condi-

![Figure 7. Curves showing variation of UTS with test temperature.](image)
Study of Preform Design for Sinter Forging

Dilatometer Tests

Dilatometer tests were conducted to gain an understanding of the thermal expansion properties of the sintered PM material. The tests were carried out on sintered PM specimens compacted at various pressures and on some steel specimens. There was no discernible variation in the results over the range $p^* = 0.77 - 0.84$, as compared with those obtained from 0.6% carbon steel.

Observations

From a visual inspection of the forgings, it became evident that those produced from the denser preforms gave consistently better results, with freedom from cracking and other surface defects. As expected, peripheral cracking was particularly evident in those preforms having flanges of the smallest diameter. An example of this is shown in Fig. 9, where some circular cracking can also be seen in the region of the central
Fig. 9. Typical peripheral and circular cracks.

Fig. 10. Lateral density-distribution diagrams for selected preforms and forgings.
Fig. 11. Mass-distribution diagrams: (a) Preform $\rho^\ast$ (mean) = 0.78 and corresponding forging; (b) Preform $\rho^\ast$ (mean) = 0.81 and corresponding forging.
boss. This latter defect was far more common in forgings produced from low-density preforms, and was undoubtedly due to poor initial compaction in this region, indicating the need to produce these preforms by double-ended compaction. A small amount of blistering on the surface of some of the forgings was also noticeable, possibly due to air entrapment. To determine the variations in lateral density, selected forgings and corresponding preforms were weighed and measured, after which known volumes of material were removed in successive stages. By repeated reweighings, the densities of the removed portions were calculated to give the density distribution curves shown in Fig. 10. These were then used to draw the true mass-distribution diagrams (Fig. 11), which take into account local density variations. The accuracies of these diagrams were checked by comparing their areas, as measured with a planimeter, against the known weights of the preforms and forgings. These were found to agree within a few grammes.

By superimposing the preform and component mass-distribution diagrams, as shown in Fig. 11, it was possible to establish the regions from which metal had been displaced. Furthermore, by means of a planimeter, the quantity of metal displaced was determined to enable Fig. 12 to be drawn, which indicates the quantities and direction of metal flow across any radius. The convention used is such that outward radial flow is regarded as positive, while inward radial flow is negative, the diagram representing the state that exists in the final analysis.

To gain an overall picture of the sinter forging process it is necessary to study the density-distribution diagram (Fig. 10) in conjunction with the lateral displacement diagram (Fig. 12), since consolidation is produced by the effect of compaction and lateral flow occurring simultaneously. Finally, the overall values of $\rho^*$ for the forgings were of the order of 0.94-0.95, while macroexamination revealed a considerable amount of peripheral porosity (Fig. 13).

Discussion

The results of the dilatometer tests suggest that the voids have no influence on the thermal expansion properties of sintered PM materials. This means that the 1 in 60 contraction allowance normally used in the design of conventional hot-forging tools for wrought metals can also be applied to those designed for sinter forging. Furthermore, assuming adequate die preheating plus the contraction in the dimensions of the preform as it undergoes the $\alpha \rightarrow \gamma$ phase change, it should be just possible to make preform diameters equal to component diameters for hot compaction at temperatures up to 1000°C. In practice, however, it
Study of Preform Design for Sinter Forging

Fig. 12. Metal-displacement diagram.

Fig. 13. Peripheral porosity in forgings.
might be more expedient to make a small allowance, with a correspond­
ing thickening of the flanges.

The fall-off in the peripheral density of the forging (Fig. 10) is borne
out by Fig. 13, which provides evidence of marked peripheral porosity.
This is undoubtedly due to a combination of tensile stresses and
chilling. Since chilling will also reduce the ductility of the material, it
must be regarded as another factor contributing to peripheral cracking.

Fig. 10 shows that the lowest density occurs at the centre of the
component and is dependent upon the initial density of the preform in
this region, while Fig. 12 shows that, irrespective of preform density,
approximately the same quantity of metal is displaced laterally from
this region (up to 10 mm radius). A possible explanation is that, since
the forging operation commences in this region, much of the final shape
is developed during initial densification of the metal, when forging
loads are relatively low. As the process continues, the loads increase
to the point where there is a preference to displace metal rather than
densify it further. In this case, the displacement is made possible by
the metal deficiency existing between the preform and the component
in the region of the intermediate flange, see Fig. 11. Obviously, to
courage the material to remain in this central region, the material
in the preform flange must be redistributed to make up the deficiency
in the region immediately adjacent to the boss, thus producing a preform consisting of three concentric cylinders, as arrived at initially
from purely geometrical considerations. The maximum density obtained
in the forgings appears to be independent of initial preform
density, from Figs 10 and 12, occurs at radii across which the mass flow
is apparently zero. This implies that the metal at these points has been
subjected to a predominantly hot-compaction process.

With reference to Fig. 7, the rise in strength which the mild steel
exhibits over the critical range is in keeping with the findings of Keane
et al. However, the graphs obtained from the tests conducted on the
PM specimen appear to be fairly linear over the range considered and do
not exhibit any peak values. This is possibly due to premature failure
of the PM specimen, resulting from the pores acting as stress-concen-
trators. Fig. 8 shows that the percent elongations for the sintered PM
specimens tend to reach a peak value at the same temperature, and in
this respect are analogous to wrought material, the main differences
being that the percent elongations obtained for the mild steel specimen
were often >100%. It can also be seen that the percent elongation
improves as the density increases. Despite the apparent low ductility
of the sintered material it can still be successfully forged with more
lateral flow than these results suggest. This is no doubt due to the
greater amount of support afforded the material as a result of the com-
plex system of stresses that occur during the forging operation. Another factor assisting in the success of the forging operation is the strain rate to which the material is subjected. The UTS obtained from tensile tests carried out at room temperatures on specimens machined from forged components tended to be somewhat low, e.g. only 355 N/mm² (23 tonf/in²), as compared with 390 N/mm² (25 tonf/in²) quoted by the powder manufacturer for sintered compacts containing 1% graphite. The cause is no doubt lack of a suitably controlled atmosphere during sintering, resulting in possible oxidization along the grain boundaries.4

Conclusions

It seems that in this instance a more uniformly dense component would have resulted if the preform shape had more closely followed the shape originally evolved by calculation, rather than the simplified shape derived from this. The advantages gained would have been:

1. A denser central region, caused by prevention of lateral displacement of metal from this region.
2. Reduced peripheral porosity, since use of larger diameters results in less lateral displacement, and hence less chilling.

The problem of chilling also emphasizes the need for adequate preheating of the dies, especially the lower one, and shorter contact times between preform and die. The increased outer diameter will, of course, make peripheral cracking virtually non-existent.

Additionally, if more rapid working is adopted to reduce contact times, the increased strain rates may improve the ductility of the material so that it can more adequately cope with the small amounts of metal displacement that take place.

Acknowledgements

The authors wish to express their gratitude to the South Wales Forgemasters, Cardiff, for kindly providing powder, tooling, and the use of their 7-5 MN (750 tonf) Ajax press; also to Firth Cleveland Sintered Products, Treforest, for sintering the hot tensile and dilatometer specimens. Finally, they are indebted to the Science Research Council, as the 3 MN (300 tonf) hydraulic press used in the tests was provided by the Council for an associated research programme currently being undertaken at the Glamorgan Polytechnic.

References


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Graphical Method for Determining Volumes of Solids of Revolution

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After reviewing the normal methods for finding the volumes of solids of revolution, the author introduces the graphical method and applies it to determining not only the volume of a forging, but also the lateral flow of material from blank to forged shape.

**Introduction**

A SOLID of revolution is generated when an area, which does not cut an axis, rotates about that axis. Common examples are cylinders, cones and spheres, the rotating areas being rectangles, triangles and semicircles, respectively.

The importance of these shapes is that they appear in some form or other in a variety of engineering components, and this frequently results in having to determine their volumes and/or masses by the most suitable means. The methods used include calculation from formulae, or the use of the Pappus or Guldinus Theorem. Additionally there is a graphical method, which has useful engineering applications, especially when applied to forging operations as will be demonstrated later.

**Basic Considerations**

A shaded rectangular strip, shown in Fig. 1, is rotated about the XX axis, the volume swept out is that of an elemental ring, and is given by 

\[ 2\pi rt \, dr \]

If the relationship between \( r \) and \( t \) is known, i.e. \( t = f(r) \), then the total volume of the solid of revolution is

\[ V = 2\pi \int_0^R rt \, dr \]

where \( R \) is the maximum radius.

The integration provides a formula which can be used subsequently to calculate volume. When the relationship between \( r \) and \( t \) is difficult to establish, recourse must be made to an alternative method of calculating volume.

**Review and Comparison of Methods**

USE will now be made of the cylinder, cone and sphere to review and compare three methods by which volumes of solids of revolution may be determined. Quite obviously, Method 1, calculation by formula, would normally be used when dealing with such simple solids.

**Method 1—CALCULATION BY FORMULA**

(a) Cylinder

![Fig. 2]

Since \( t \), the thickness of the cylinder, is constant,

\[ V = 2\pi \int_0^R t \, dr \]

\[ = 2\pi t \left[ \frac{r^2}{2} \right]_0^R = \pi R^2 t \]

(b) Cone

![Fig. 3]

By similar triangles:

\[
\frac{t}{h} = \frac{r}{R}, \quad \frac{h}{R - r} = \frac{h}{R}
\]

\[
R = \frac{R^2}{R - r}
\]
Substituting for \( t \) in equation (i), total volume
\[
2\pi \int_0^R t(R - r) \, dr
\]
\[
= R \int_0^R \frac{R^2}{2} - \frac{r^3}{3} \, dr
\]
\[
= \frac{2\pi R}{6} \left( \frac{R^3}{3} - \frac{r^3}{3} \right) \bigg|_0^R
\]
\[
= \frac{4\pi R^3}{3}
\]

NOTE that the usual textbook proof of this formula is obtained by considering elemental discs rather than rings.

Method 2—THEOREM OF PAPPUS OR GULDINUS

If an area, which does not cut an axis, rotates about that axis, the volume swept out is equal to the product of the area and the distance moved by its centroid.

The proof is as follows: from Fig. 1 area of elemental strip is \( t \, dr = dA \), so that total volume is \( 2\pi \Sigma dr \)

where \( \Sigma dr \) is the first moment of area about axis XX, i.e. \( \Sigma drA = r \times A \)

where \( r = \) distance from XX axis to centroid of area

and \( A = \) total area.

Consequently, total volume
\[
= A \times 2\pi r
\]

area \( \times \) distance travelled by its centroid.

To use this method, the positions of the centroids of the areas must be known which, in the case of complicated shapes, may either involve lengthy calculations or resorting to suitable graphical or experimental techniques.

For the three solids under consideration, the procedure would be as follows.

(a) Cylinder
Area of rectangle shown in Fig. 2 = \( Rt \)
Position of centroid from XX axis = \( r = R/2 \)
Total volume = \( \frac{2\pi R^3}{3} \)

(b) Cone
Area of triangle shown in Fig. 3 = \( \frac{\pi Rh}{2} \)
Position of centroid from XX axis = \( r = R/3 \)
Total volume = \( \frac{2\pi R^3}{3} \)

(c) Sphere
Area of semi-circle shown in Fig. 4 = \( \pi R^2/2 \)
Position of centroid from XX axis = \( r = R/2 \)
Total volume = \( \frac{2\pi R^3}{3} \)

Method 3—GRAPHICAL METHOD

This method can be applied directly to any solid of revolution, irrespective of section and, despite the need for accurate drawing, the calculations are straightforward. The method is intended for use with a planimeter to measure the area obtained; however, Simpson’s Rule will be used in the first instance to demonstrate the validity of the method.

The total volume of a solid of revolution
\[
= \int_0^R r \, dr
\]

which may be represented by the area under the curve obtained by plotting values of \( 2\pi r t \) on the ordinate against values of radius \( r \) along the abscissa.

Since \( 2\pi \) is constant, it is only necessary to calculate the product \( rt \) and plot this to a base of radius \( r \), as shown below.

(a) Cylinder
Since the thickness \( t \) is a constant, the diagram will take the form of a triangle, as shown in Fig. 5.

Area under curve
\[
= \frac{(2\pi Rt \times R/2) = \pi R^3t}{\pi} \quad \text{volume of cylinder}
\]
TABLE II

<table>
<thead>
<tr>
<th>r</th>
<th>0</th>
<th>0.1R</th>
<th>0.2R</th>
<th>0.3R</th>
<th>0.4R</th>
<th>0.5R</th>
<th>0.6R</th>
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<td>0.96R</td>
<td>0.78R</td>
<td>0</td>
</tr>
</tbody>
</table>

(b) Cone

In this case \( t = h(R - r)/R \), though in the general case the actual thickness will be measured directly off an accurate drawing of the section being considered.

The resulting curve is shown in Fig. 6. Whilst in the general case the area under the curve should be measured with a planimeter, for our present purposes it will be evaluated using Simpson’s Rule, which states that the area of the figure is equal to one third the width of strip selected multiplied by the first and last ordinates, plus four times the sum of the even ordinates, plus twice the sum of the remaining odd ordinates. The area under this curve is therefore

\[
\frac{0.1R \times 2\pi \times Rh}{3} = \frac{4 \times (0.09 + 0.21 + 0.25 + 0.21 + 0.09) + 2 \times (0.24 + 0.24 + 0.16)}{3} = \frac{7.2}{3} \times 4 = \text{volume of cone.}
\]

 application of the Graphical Method to a Simple Forging Operation

FIG. 8 shows a section through a component which was hot forged in a closed die from a solid cylindrical blank. The graphical method can be applied to such a forging operation in the following manner.

The first requirement is to ascertain the weight of material needed, and then make appropriate allowances for flash, etc. As the component is a solid of revolution, any of the three methods discussed could be used for this purpose, since weight is merely volume multiplied by density. Both Methods 1 and 2, however, require sub-dividing the component into portions which either conveniently lend themselves to solution by formulae, or whose centroids can be accurately located.

With the graphical method the problem can be approached directly. The thicknesses are scaled off an accurate drawing of the component at its principal points, i.e. changes of section, etc., and these values are tabulated against the corresponding radii, as shown below.

Application of the Graphical Method to a Simple Forging Operation

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With the graphical method the problem can be approached directly. The thicknesses are scaled off an accurate drawing of the component at its principal points, i.e. changes of section, etc., and these values are tabulated against the corresponding radii, as shown below.

Values of radius \( r \times \) thickness \( x \) density are then calculated (bottom row of table) and plotted to a base of radius to give a mass distribution diagram, as shown by the chain-dotted line in Fig. 9. The area under this line represents the mass of the component, and when measured by a planimeter this was found to be 414g.

For present purposes it will be assumed that flashless forging is employed so that no allowance has to be added to this weight, hence the mass of the blank will also be 414g.

The dimensions of this blank can now be determined as follows. Volume of metal required is mass/density = (414 \times 10^3)/7.8 = (53 \times 10^3) mm^3.

Assuming the diameter of the blank to be 60 mm after heating and upsetting to remove scale prior to forging, then \( \pi/4 \times 60^2 \times (\text{thickness of blank}) = 53 \times 10^3 \) from which the blank thickness is 19 mm.

Using this information it is now possible to draw the mass distribution diagram for the blank, which takes the form of a triangle, shown superimposed on the mass distribution diagram of the forging, Fig. 9.

In plotting this diagram the maximum value of 414g at a radius of 30mm is used to determine the maximum value of 414g at a radius of 30mm.
TABLE C

<table>
<thead>
<tr>
<th>Radius r (mm)</th>
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<th>15</th>
<th>17-5</th>
<th>20</th>
<th>22-5</th>
<th>32-5</th>
<th>47-5</th>
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<tbody>
<tr>
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<td>10</td>
<td>19-1</td>
<td>16</td>
<td>10</td>
<td>8-64</td>
<td>8-64</td>
<td>4-83</td>
<td>4-83</td>
</tr>
<tr>
<td>Density ρ (Mg/m³)</td>
<td>7-8</td>
<td>7-8</td>
<td>7-8</td>
<td>7-8</td>
<td>7-8</td>
<td>7-8</td>
<td>7-8</td>
<td>7-8</td>
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<tr>
<td>rt ρ (kg m)</td>
<td>0</td>
<td>0.39</td>
<td>1.49</td>
<td>1.872</td>
<td>1.365</td>
<td>1.248</td>
<td>2.19</td>
<td>1.224</td>
<td>1.79</td>
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</tbody>
</table>

With reference to Fig. 9, it can be seen that metal is displaced from region (a) to make up the deficiency at (b), the surplus spreading into (c), which is then displaced to make up the deficiency in region (d). In fact, the differences in the areas of the two diagrams up to any radius represents the mass of metal which has been displaced across that radius in forming the component.

These differences were measured with a planimeter to provide the following table of values, which were then used to construct Fig. 10, in which sufficient accuracy prevails if the principal points are merely joined by straight lines.

Hence, this graphical technique, in addition to determining the weight of the forging, also provides a better understanding of the lateral flow which takes place during the forging operation. Without this method, such information could only be obtained as a result of tedious calculation.

Conclusion

It is suggested that the technique described above may prove useful in the design of preforms for closed die forging, since it can be used to quantify experience previously gained with successful preform shapes obtained by trial and error methods.

The method has already been used to provide information regarding the lateral flow of metal during sinter forging of P/M preforms, where the problem is aggravated by the non-uniform density distribution arising from the powder compaction process used to produce the preforms.

Finally, I would like to thank the South Wales Forge-masters Ltd., Cardiff, for permission to make reference to the forging shown in Fig. 8.


Fig. 9 — The triangle represents the volume of the blank and the chain-dotted line that of the forging.

Fig. 10 — Shows the lateral flow of material from the blank to the forged shape.
Details and Calibration of Load Cell

Details of the load cell appear in Fig. 1E, where it can be seen that the working region consists of a hollow cylinder with eight equi-spaced bi-axial strain gauges attached around the circumference. These are connected together to form a temperature compensated Wheatstone bridge, so arranged that one pair of opposing arms measures longitudinal strains, whilst an opposing pair measures circumferential strains. In this way, a compressive load causes an increase in the resistance of the circumferential strain gauges, and a decrease in the resistance of the longitudinal strain gauges. The resulting out of balance in the circuit causes a current to flow through the central galvanometer, which in turn produces a deflection on the Ultra-Violet (U.V.) recorder. The measuring equipment was energised with a 12 V d.c. supply, and either one of two currents could be selected viz. 4.4 mA or 10 mA, to provide a high and low load range respectively.

The deflection of the U.V. recorder was related to the ram force by the following calibration tests, which involved the use of an inductive differential pressure transducer, a transducer/ converter Type 905 and a digital voltmeter (DVM). Before this equipment could be used on the press, it was necessary to calibrate the DVM with the aid of a dead weight tester, to provide the results shown plotted in Fig. 2E. The equipment was then connected into the ram inlet pipe of the press as shown in Fig. 3E, and the load on the press was increased in increments of √ 10 tonf. This load was monitored with the aid of a Bourdon type gauge, which was already fitted to the press and had been previously
8 Off, Bi-axial strain gauges equally spaced around circumference at X.
FROM SLOPE OF GRAPH:

\[ 1\text{V} = 1400 \text{ bpf/in}^2 \]

FIG. 2E CALIBRATION OF DIGITAL VOLTMETER

DEAD WEIGHT PRESSURE (lf/lb)
Fig. 3E  INDUCTIVE DIFFERENTIAL PRESSURE TRANSDUCER, TOGETHER WITH TRANSDUCER/CONVERTER AND DIGITAL VOLTMETER SHOWN CONNECTED INTO RAM INLET PIPE OF "ARMSTRONG" PRESS
FIG. 4E: ERROR IN BOURDON GAUGE READINGS
FIG. 5E LOAD CELL CALIBRATION CHART
calibrated in tonf, making use of the fact that the working area of the ram was 148.44 in².

The DVM reading and the UV recorder deflections were noted for each load increment for both the low and high load ranges. The results of these tests are shown plotted as Figs. 4E and 5E, the latter being virtually identical to that obtained by Holloway⁹².
APPENDIX F

Design Details of Compaction Tool Die to
Produce Cylindrical Preforms

The stresses occurring in the die-insert and the shroud, were calculated using the standard compound cylinder theory\(^\text{93}\), based on thick-walled cylinders. The dimensions of the die components are shown in Fig. 1F. Since the tool was designed to withstand a maximum compaction force of 600 kN then the corresponding maximum compaction pressure

\[
\frac{\text{maximum load}}{\text{cross sectional area of compact}} = \frac{600 \times 10^3}{\pi \times 40^2} = 477 \text{ MN/m}^2
\]

It was assumed that this pressure was transmitted to the walls of the die as a radially acting hydrostatic pressure, and furthermore that it acted over the total length of the die whereas in practice it would only act over the length of the compact i.e. 50 mm or \(\sqrt[4]{40}\%\) of the total die length.

Hence the known conditions were:

- Young's modulus for both materials = 207 GN/m\(^2\)
- For the die insert, at \(\rho = 20 \text{ mm}\), \(\sigma_+ = -477 \text{ MN/m}^2\)
- For the shroud, at \(\rho = 125 \text{ mm}\), \(\sigma_+ = 0\)
- At the mating interface, \(\rho = 48 \text{ mm}\) nominally, and the radial stresses in the die insert and shroud were equal i.e. \(\sigma_{+i} = \sigma_{+o}\)

The radial displacement \(\mu_i\), of the insert inwards plus the radial displacement, \(\mu_o\), of the outer cylinder outwards due to the shrink fit must equal the interference value. In deciding this interference value it was important to ensure that the stresses would not exceed the
FIG. 1F STRESS DISTRIBUTION IN DIE AND SHROUD
elastic limit of the metals concerned. It was therefore necessary to perform several calculations before arriving at an interference value of 0.1 mm,

\[ -\mu_i + \mu_0 = 0.1 \text{ mm} \]

hence,

Using the above conditions and the constants A, B and C, N for the insert and shroud respectively, the following four equations were obtained:

\[ -477 = A - \frac{B}{0.0004} \]

\[ 0 = C - \frac{N}{0.0156} \]

\[ A - \frac{B}{0.0023} = C - \frac{N}{0.0023} \]

and \[ \frac{-\mu_i}{0.048} + \frac{\mu_s}{0.048} = \frac{0.0001}{0.048} \]

or \[ \varepsilon_{\theta i} + \varepsilon_{\theta s} = \frac{0.0001}{0.048} \]

Substituting for the strain in terms of stresses in (4F):

\[ -\sigma_{\theta i} + \sqrt{\sigma_{\theta i} + \sigma_{\theta s}} - \sqrt{\sigma_{\theta s}} = \frac{0.0001}{0.048} \times 207 \times 10^3 \]

and since \( \sigma_{\theta i} = \sigma_{\theta s} \)

\[ \sigma_{\theta s} - \sigma_{\theta i} = 431.25 \text{ MN/m}^2 \]

\[ \therefore C + \frac{D}{0.0023} - (A + \frac{B}{0.0023}) = 431.25 \]

From equation (2F), \( C = \frac{N}{0.0156} \) substituting into equations (3F) and (5F):

\[ A - \frac{B}{0.0023} = \frac{N}{0.0156} - \frac{N}{0.0023} \]

and \[ \frac{N}{0.0156} + \frac{N}{0.0023} - A - \frac{B}{0.0023} = 431.25 \]

From equation (6F):

\[ N \left( \frac{0.0023 - 0.0156}{0.0156 \times 0.0023} \right) = A - \frac{B}{0.0023} \]
\[ N = 0.0027 \left( \frac{B}{0.0023} - A \right) \]  \hfill (8F)

Subst. for \( D \) from equation (8F) into equation (7F)

\[ 1.347 \left( \frac{B}{0.0023} - A \right) - A - \frac{B}{0.0023} = 431.25 \]

\[ 585.65B - 1.347A - A - 434.78B = 431.25 \]

\[ 150.87B - 2.347A = 431.25 \]

hence \( A = 64.28B - 183.75 \) \hfill (9F)

Subst. for \( A \) from equation (9F) into equation (1F)

\[ -477 = 64.28B - 183.75 - 2500B \]

\[ 2435.72B = 293.25 \]

\[ \therefore B = 0.1204 \text{ MN} \]

From equation (9F)

\( A = -176 \text{ MN} \)

From equation (8F)

\[ N = 0.0027 \left( \frac{0.1204}{0.0023} + 176 \right) \]

\[ N = 0.616 \text{ MN} \]

From equation (2F)

\[ C = \frac{0.616}{0.0156} = 39.52 \text{ MN} \]

Hence the stresses likely to occur in the die insert and the shroud can now be calculated as follows:

**The Die Insert**

**Radial stresses:**

at \( \theta = 20 \text{ mm} \)

\[ \sigma_r = -176 - \frac{0.1204}{0.0023} = -477 \text{ MN/m}^2 \]

and at \( \theta = 48 \text{ mm} \)

\[ \sigma_r = -176 - \frac{0.1204}{0.0023} = -228.35 \text{ MN/m}^2 \]
Circumferential stresses:

at \( r = 20 \text{ mm} \), \( \sigma_\theta = -176 + \frac{0.1204}{0.0004} = 125 \text{ MN/m}^2 \)

and at \( r = 48 \text{ mm} \), \( \sigma_\theta = -176 + \frac{0.1204}{0.0023} = -123.65 \text{ MN/m}^2 \)

The Shroud

Radial stress:

at \( r = 48 \text{ mm} \)

\[ \sigma_r = 39.52 - \frac{0.616}{0.0023} = -228.3 \text{ MN/m}^2 \]

and at \( r = 125 \text{ mm} \)

\[ \sigma_r = 39.52 - \frac{0.616}{0.0156} = 0 \]

Circumferential stresses:

at \( r = 48 \text{ mm} \)

\[ \sigma_\theta = 39.52 + \frac{0.616}{0.0023} = 307.35 \text{ MN/m}^2 \]

and at \( r = 125 \text{ mm} \)

\[ \sigma_\theta = 39.52 + \frac{0.616}{0.0156} = 79 \text{ MN/m}^2 \]

The resulting distribution of radial and hoop stresses occurring in the die insert and shroud, are shown in Fig. 1F.
APPENDIX G

Admix Mark II Gas Mixing Unit

This unit was designed and supplied by BOC Ltd. It can be seen in Fig. 1G, mounted on a trolley together with its own nitrogen and methane cylinders, although a natural gas supply can be substituted for the latter. The trolley was designed and constructed at the Polytechnic in order to make the Unit portable. The circuit diagram, Fig. 2G, shows the arrangement of the component parts of the Unit, whilst the following explanatory notes are based on those supplied by BOC Ltd.

The Unit was designed to supply the sintering furnace with nitrogen containing up to 5% natural gas (or methane), at flow rates of up to 3.4 m$^3$/h. Once the gas mixture enters the furnace the methane becomes the reducing medium since it cracks into its constituent components at 1050°C, as follows,

$$\text{CH}_4 + T^0 \rightarrow \text{C} + 2\text{H}_2$$

From this it can be seen that one volume of methane produces two volumes of hydrogen so that when a 96% N$_2$/4% H$_2$ mixture is required during sintering, the methane addition need only be 2%.

It was recommended that during the "burn off" of the zinc stearate, the flow rate should be $\sim 0.7$ m$^3$/h, to ensure the complete removal of the volatiles from the furnace chamber. This flow could then be reduced to $\sim 0.3$ m$^3$/h during the actual sintering cycle. Furthermore, to ensure complete removal of the air from the furnace prior to sintering the volume of nitrogen to be used to purge the furnace should be 6 times the volume of the furnace chamber.

The Admix II system was designed to be self-correcting and fail
Fig. 1G ADMIX II GAS MIXING SYSTEM MOUNTED ON TROLLEY TOGETHER WITH ITS OWN NITROGEN AND METHANE SUPPLY
MIXED GAS 2" wg

PRESSURE TEST NIPPLE

PRESSURE LINE TO SAFETY SHUT-OFF VALVE

24" wg

NITROGEN 20-40 psig

NATURAL GAS 7½" wg

FIG. 2G “ADMIX II” CIRCUIT DIAGRAM

(See following sheet for schedule of equipment)
ADMIX II SYSTEM

Schedule of Equipment


5. Nitrogen metering orifice. 17/64" orifice in suitable plate and holder. 120 scfh nitrogen passing through this orifice will be reduced in pressure from 6" wg to 4" wg.


7. Furnace pressure regulator. Bryan Donkin ½" Fig. 226 regulator, fitted spring No. 397 for 1½" to 3" wg.


9. Natural gas non return valve. ½" Amal "Otan" non return valve.

10. Safety shut off valve (to shut off natural gas supply if nitrogen pressure falls). Bryan Donkin fig. 999R pressure operated valve connected by ½" OD copper and "Enots" adaptors to nitrogen line just upstream of second stage regulator. Regulator fitted with 15 to 30" wg spring.

11. Natural gas flowmeter. Gapmeter GUH Series, size 6CD. 0.6 to 6.0 scfh.


13. Natural gas metering orifice. ½" "Ballofix" adjustable metering orifice. An adjustable orifice is specified to allow the natural gas proportion in the mixture to be adjusted as and when necessary.

14. Nitrogen bypass bleed. Amal metering jet in holder. This bleed ensures that a predominance of nitrogen will always be present in the gas even at the very lowest flow rates.

P Pressure test nipples.
safe, so that the mixture would remain at a set value for the complete range of flows used, whilst a safety shut-off valve will automatically stop the natural gas or methane flow, if the nitrogen pressure failed.
APPENDIX H

Horizontal Tube Sintering Furnace

The partly complete horizontal tube sintering furnace referred to in Section 6.3 can be seen in Fig. 1H. Although the furnace was constructed at the Polytechnic, the basic design was the work of Dr. J. H. Buddery of Severn Science Ltd., Bristol.

The furnace consists essentially of an Inconel 600 Schedule 10 work tube measuring approximately 2 m long and having a nominal bore size of 75 mm diameter. This tube carries water cooled aluminium-alloy end caps fitted with sealing rings to prevent the air from entering the system during operation. The caps were fitted with easily removeable end plates which carried the gas fittings necessary to introduce and exhaust the controlled atmosphere.

With reference to Fig. 1H, it can be seen that the furnace is sub-divided into two electrically heated zones represented by two independent furnaces. The first of these, which appears as the large cylindrical shape in the foreground of the photograph, is a Kanthal wire wound furnace. This single phase 2.3 kW furnace was intended to operate at $\sim 700^\circ C$ in order to preheat and remove the lubricant from the green compacts before transferring them into the sintering zone, which appears as the large rectangular shape in the background of the photograph. Since the sintering zone was designed to operate at temperatures of 1130 - 1150°C it was necessary to use 4 Crusilite (silicon carbide) rods for this purpose. These rods, each with a nominal electrical resistance of $\sim 2.1 \Omega$ were connected in series, and arranged around the Inconel tube in a square configuration.
Fig. 1H  HORIZONTAL TUBE SINTERING FURNACE
The electrical supply to the Crusilite rods was 3-phase, whilst control was effected by means of a suitable Eurotherm proportional band temperature controller, governed by a thermocouple positioned near the centre of the sintering zone. The bulk of the rectangular box shape which represents the sintering zone is made up of heat insulating brickwork, enclosed in a galvanised sheet and welded angle iron frame. Since it was necessary to keep the ends of the heating elements cool, then these were allowed to protrude from the ends of the furnace. However, for electrical safety reasons, it was necessary to enclose these live ends of the heating elements with a suitable perforated metal cover.

The complete furnace assembly was mounted on an all welded angle iron frame as shown in Fig. 1H, which was fitted with shelves to carry the control equipment, just visible underneath the sintering zone, and any other ancillary equipment.
Experimental Results Obtained from Direct Tensile Tests

These tests were performed on sintered specimens either produced from "as supplied" Hoganas ARC 100.29 iron powder, or from one of its fractions. In all cases the powders were admixed with 0.5% graphite and 1% zinc stearate.

(i) Sub-sieve powder fraction

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(ii) **Powder particle size (+90 -125 μm)**

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(iii) Powder particles size (+150 - 180 µm)

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(iv) "As supplied" powder

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