HOT WORKING OF POWDERS, PROCEDURE AND PRODUCTS

BJÖRN ARÉN
Division of Metal Working

UNIVERSITY OF LULEÅ
HOT WORKING OF POWDERS,
PROCEDURE AND PRODUCTS

av

Björn Arén

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Abstract

This dissertation can be divided into two main parts. The first deals with the hot working conditions of mild steels in powder forging, the second with the deformation behaviour of powder forged products; both monolithic and compound parts are studied.

Hot forging of low alloy steel powders makes it possible to produce complicated parts with properties comparable with homogeneous forged parts, whilst reducing the machining operations necessary to a minimum. This powder forging technique is a high production rate method, and can only be applied to mass produced components. It also gives the possibility of complete reconstruction of the shape of the part. Forging of porous preforms however require certain precautions to be made to prevent working faults and to optimize the densification process. These things are therefore studied with some interest.

In order to compete with other high production rate methods such as hot forging, the powder forged parts must be economically advantageous. Therefore the most attractive parts to be produced are those in which substantial material is saved, where most machining operations can be excluded and in which material of substantial structure is required.

A method to produce compound structural parts of high strength hardened steel and ductile mild steel from alloyed powders is presented. It involves the manufacture of compound preforms by forging and their heat treatment to produce an optimal structure.

On examining the deformation behaviour of compound parts produced, a substantial rise in the fracture toughness is seen, which is associated with the interface between hard and ductile material zones.

KEY WORDS:
HOT WORKING, POWDER FORGING, DENSIFICATION, COMPOUND STEEL.
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This dissertation comprises work contained in the following studies:

I  Fischmeister, Arén and Easterling:
   Deformation and Densification of Porous Preforms in Hot Forging

II  Arén, Olsson and Fischmeister:
   The Influence of Presintering and Forging Temperature in Powder Forging

III  Arén:
   Material Flow in the Powder Forging of a Gear-Profile as a function of Preform Shape

IV  Arén:
   FEM-Simulation of the Densification Process in the Powder Forging of a Linear Gear Profile
   Linköping University, Scientific report IKP R - 40, 1974.

V  Arén:
   Optimizing the Preform Shape in Powder Forging of a Linear Gear Profile

VI  Arén:
   Powder Forging of Compound Steels, Part I
   University of Luleå Proceedings, TULEA 1982:16
   Submitted to Powder Met. Int. for publishing.

VII  Arén:
   Powder Forging of Compound Steels, Part II
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SUMMARY

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Abstract

This dissertation can be divided into two main parts. The first deals with the hot working conditions of mild steels in powder forging, the second with the deformation behaviour of powder forged products; both monolithic and compound parts are studied.

Hot forging of low alloy steel powders makes it possible to produce complicated parts with properties comparable with homogeneous forged parts, whilst reducing the machining operations necessary to a minimum. This powder forging technique is a high production rate method, and can only be applied to mass produced components. It also gives the possibility of complete reconstruction of the shape of the part. Forging of porous preforms however require certain precautions to be made to prevent working faults and to optimize the densification process. These things are therefore studied with some interest.

In order to compete with other high production rate methods such as hot forging, the powder forged parts must be economically advantageous. Therefore the most attractive parts to be produced are those in which substantial material is saved, where most machining operations can be excluded and in which material of substantial structure is required.

A method to produce compound structural parts of high strength hardened steel and ductile mild steel from alloyed powders is presented. It involves the manufacture of compound preforms by forging and their heat treatment to produce an optimal structure.

On examining the deformation behaviour of compound parts produced, a substantial rise in the fracture toughness is seen, which is associated with the interface between hard and ductile material zones.
Introduction

A combination of the advantages of both conventional powder metallurgy and hot forging makes it possible to produce parts complicated in shape, not only with close tolerances and high quality surfaces, but also with properties comparable with homogeneous forged parts. Many components, mainly for the automotive industry, have been produced in pilot plants and many of these have performed equally as well as corresponding hot forged parts, and in some cases even better (1). However, there are still few components in commercial production despite the surge of interest, research, and development during recent years. The commercial viability is still in dispute (2).

After a period when early expectations were not realized, and it was accepted that quick savings would be prevented by amortization of equipment and factory alterations it is now generally believed that powder forging will slowly find its own niche beside other high-precision, high production-rate methods.

Practical experience however shows that economic production of components by powder-forging might not be viable for the production of complicated parts in which a high degree of plastic flow is involved (3). The reason is to be found in the die wear caused by hot preform material under high pressure sliding against the die walls. In contrast with the part shape expected initially to be of value, e.g. the well known GM pinion-gear, modern parts in production (4) can be described as fully dense "sinter-shaped" parts in which the forging operation involved is principally pure upsetting.

With this background, modern powder forged parts can be characterized as sintered parts with improved strength and thoughness, a virtue of the pore-frees structure. Such effects however have also been achieved by improvements in the powder and procedure involved in the production of sintered parts. The market for powder forgings has therefore been reduced for two reasons:
1. Complicated shapes cause intolerably expensive die wear and therefore cannot be powder forged.

2. Powder forgings of "sinter type" design could in many cases be replaced by sintered parts made of modern high-strength sinter-alloys.

The remaining market for powder forging can therefore be described as: "sinter-shaped" parts which require such strength and toughness that a porous structure is not acceptable. There is also a market for structural parts which can only be produced as powder forgings. An example of these are compound parts (5).

Several investigations have been made concerning the problem of pore closure in hot compaction of powder material. Very often the pores themselves have been primarily studied and the surrounding material where the activities of densification in reality take place is often seen as a continuum. Although the pore size, shape and distribution are very important factors for the compaction, the solid phase will control the pore closure. In 1948 Torre (6) for instance studied the closure of a spherical pore as a function of external isostatic pressure. Later, other investigators has shown that an addition of shear will be of importance (7). Even in the late sixties and during the seventies, however, the closure investigations presented mainly discussed the pores and not the material in itself.

Indeed, no important attention has been given to the base material until Arzt, Ashby and Easterling 1982 (8) presented sintering maps revealing the main densification mechanisms as a function of temperature, pressure, sintering time and particle size.

Compared to methods such as hot forging and hot pressing, the Hot-Isostatic Pressing (HIP)-process is associated with substantially lower strain rates. The difference in densification time is 3 to 4 orders of magnitude. Thus an extrapolation of
the HIP-diagram to the application of the rapid densification methods appears to be unrealistic. Nevertheless the principles of the HIP-diagrams are very useful and may also be applied to forging and pressing.

In hot forging the pressure and time parameters, however, are functions of the geometry, size and design of the press, the die-tool and the work-piece. The consequence of this is that the diagrams for hot forging therefore can be simplified but will only be valid for the given conditions.

An effect clearly illustrated by the HIP-diagram is the importance of pressing time necessary to reach full density. Another important factor is the temperature. By combining these two factors one can find the key for optimizing the hot working processes for powders.

One conclusion of this is that hot pressing in a hydraulic press where the pressure can be held for some time should be advantageous in comparison with a rapid forging operation performed in a mechanical press without facilities for maintaining the pressure. When applying this conclusion to powder forging, however, it should be remembered that cooling of the hot preform in the die tool will start at the surface, thus causing a substantial reduction in surface densification achieved. This is well documented in the litterature (e.g. 4, 9). With a long pressing time, however, the die surface temperature will increase and thus cause severe die wear.

As a result of the investigations undertaken in this field it can be noted that hot working of powder preforms today is standing on a more stable foundation that when this work once started.
Hot forging procedure

The forging method used in this work has been developed from the simple upsetting of sponge iron compacts to the forging of high speed steel compound bars under vacuum. This change in forging method is a result not only of the change in powder material investigated, but also of development of the equipment available.

In the first study (I) two types of deformation were investigated, namely (a) pure upsetting between flat dies and (b) shear deformation. The dimensions of the specimens were 10x10x100 mm. The bars were made of Höganäs HC 100.25 sponge iron powder without a lubricant addition.

Four series of bars were made with densities between 4.885 and 6.250 g/cm^3; expressed as relative densities: 62.6 - 80.2%.

The bars were heated to 1200°C in the hot zone of a tube furnace with a hydrogen atmosphere and held for 6 min. No prior sintering treatment was employed. Transferring the specimens from the furnace to the die took between 2 and 3 s, during which time the temperature dropped by ~40°C, as recorded by a thermocouple embedded in a test-bar. During this time the specimen was protected from oxidation by the hydrogen diffusing from its pores.

Forging was carried out using a 250 tons toggle-press with planeground dies. After the stroke, the specimens were allowed to cool on the die platen. The press was equipped with force and displacement transducers and signals from these were recorded. During the upsetting a plane strain condition was realized in the bar.

The upsetting of elongated bars is a well-established technique for plane-strain experiments in conventional forging research. In the case of porous preforms, the validity of the plane strain condition in the centre section was verified by the observation that the increase in length of the bars during upsetting was always much less (of the order of a few per cent only) than the increase in width up to the highest strain investigated.
No lubricant was used between the dies and the preforms.

In the second study (II) the effects of presintering and forging temperature were studied. The methods and material used were the same.

Part of the material was presintered in batches in a hydrogen tube furnace, where the bars were held for either 4 or 24 hrs at 1100°C, and subsequently cooled in hydrogen. As expected, the effects of presintering were small, and long sintering times were necessary to demonstrate them.

In figure 5 of paper II a study of the forging pressure required to reach a given strain is presented. There the original data points are fitted with a straight line through the transition temperature, indicating that we did not at that time find any marked change in deformability at this transition. We have later found however, that the data points can also be fitted by a curve showing a slight effect of transition (10) which is in better agreement to the results of other investigators (e.g. 11).

The low transformation effect on the forging pressure found in paper II is explained by our forging to a very high density level (98.2 - 99.7%) which greatly affect the forging pressure, as the forged density asymptotically approaches the theoretical density with increasing forging force. Studying our forged specimens in detail it was found that the centre density vary with the forging temperature, and we mean that this variation can compensate for the influence of the phase transformation on the forging pressure.

The yield strength and ultimate tensile strength was slightly diminished by raising of the forging temperature. The temperature dependence can be explained as a result of grain growth.
Presintering was found to increase the forging pressure, but did not affect the strength of the material as forged. Presintering however, is later found to diminish the tendency of stress cracking in powder forging, but the effect is very small (12, 13).

In the third paper (III) the aim was to study the optimization of the preform geometry for a given die tool.

In this case the preforms were machined from presintered (1 hour at 900°C in an atmosphere of cracked ammonia) bars (16x10x100 mm) of Höganäs AHC 100.29 iron powder with a mean relative density of 82.4%. The bars were preheated to 900°C (1173 K) in a tube furnace with a hydrogen atmosphere with a holding time of 6 min. The temperature difference between the furnace and the centre of the sample was measured to be about 5°C.

Forging equipment was developed so that the forging operations were performed in a screw press with a closed preheated die tool lubricated with a suspension of graphite in mineral oil. Figure 1.

Figure 1. Section through forging equipment.
The transfer of the specimens from the furnace to the forging die took between 1 and 2s and was performed by pulling a transport wire through the die tool as illustrated in figure 1. The die was preheated to 200°C (473 K). This operation caused a central temperature drop of approx 15°C (corresponding to a calculated drop of 50°C at the surface).

In comparison with the previous forging method two differences were apparent. The first was that the force was limited by the construction of the tool, limiting the final mean density to around 95%. The second was a slight variation in the preform density after cold compaction, caused by the large height-to-width ratio (1.6:1) of the green compacts from which the preforms were machined. The result presented have been corrected for these systematic variations.

Deformation behaviour of porous structures

As a foundation for the study, some aspects of the fundamental behaviour of porous materials will be discussed.

Because of the compressibility of pores, plastic deformation of the material is accompanied by a volume change, the result of tensile stresses expanding, and compressive stresses compressing the material.

An interesting factor in the process is the proportions of shear and uniaxial flow. A very rough idea of densification produced by shear flow was developed by forging specimens with inlaid square patterns as shown in fig. 2 (I). No pure shear flow was found in this experiment, but at the centre, the degree of shear was demonstrated with reasonable accuracy by the distortion of the raster net. The densification can be related to the shear by the deformation of the grid elements.
Figure 2. Densification as measured by the change of grid element area ($A_r/A_o$) versus shear ($\tan \theta$) (a) for a shear type forging (b).

In this experiment it was not possible to separate the contributions to densification of shear and compressive flow, but although the importance of shear in the densification process was not fully demonstrated, it can be shown that a lack of shear may have important consequences with some deformation geometries.

Notice for example the material flow against a wall that gives "dead" zones of high porosity.

A similar effect occurs in the forging of parts containing planes of symmetry where two material flows impinge as if against an imaginary wall, figure 3.

In this case two material streams converge forming one vertical stream, giving a zone of low density in the centre. With large deformations the parts were found to crack in the "fold zone" in addition to tensile stress cracking at the top.
Figure 3. Impingement flow in extrusion.

Tensile stresses tend to extend the material (I), leading to rapid pore growth and finally to the formation of cracks. The cracks close again towards the end of the forging stroke (figure 3). This is in agreement with the results of Bockstiegel and Björk (14).

For the geometry in fig 3, however, the stress cracks were not closed significantly even with very high forces due to a less favourable geometry.

Surface cracking can be reduced slightly by presintering (12) but at the expense of higher flow stress (III). Another cause of cracking is folding that may occur during the forging of undercut sections (14) as well as in extrusion (fig. 3). The occurrence of strain cracking and folding therefore sets the limits for the geometrical relationship between preform and die.

Finally it is worth noting that the surface zones of powder forged samples often attain lower density than the central material. There appear to be two reasons for this effect; absence of shear flow discussed above, and cooling which raises the flow stress of the surface material.
The searching for a constitutive equation for the deformation of porous materials has been the topic of a number of investigations and a review of proposed laws of deformation has been given e.g. by Corapigogly and Uz (15). In their review, however, the problems of describing porous preform deformation is well illustrated in that sense that not less than five different proposed laws of deformation are presented, all of which roughly fit experimental data. In conclusion it has to be accepted that laws governing porous preform deformation maintain to be established.

Densification studies

An important subject of the first part of the work is the criteria of densification which must be understood to optimize the geometry in powder forging.

Initially (I) the effects of pure upsetting and shear deformation, figure 2, were investigated. Differences in densification were revealed easily by systematic Brinell hardness testing of actual square sections.

Often however, true forging cannot be simplified to these forging principles. In the third study (III) therefore, the influence of the geometrical relationship between the preform and the forming die was studied. The techniques chosen were the forging of wave profiles from prismatic preforms, and gear rods forged from preshaped preforms.

The resulting structures were studied by Brinell hardness testing of a section taken 1 mm from the lateral surface of the specimens. Mirror symmetry was utilized in the derivation of hardness maps from the indentation grids (typically 150 indentations per sample). An example of a Brinell-hardness map is given in figure 4.
Brinell hardness was chosen to characterise the material rather than quantitative metallography because the method saves time and has a depth-integrating effect. To eliminate residual effects of work hardening, the specimens were annealed at 550°C for 30 min before testing.

The forging of wave-profiles showed that the densification behaviour of the porous structure is analogous to shovelling wet snow; material piles up and densifies in front of the shovel, figure 4.

In the investigation of gear rod forging, experiments were made with five groups of preshaped preforms designed to give fully dense forgings of equivalent form. The dimensions are defined in figure 5 and tabulated below.
Figure 5. Main dimensions of preform (a); and fully dense part (b) \( l_B = L_B \) and \( l_T = L_T \)

The deformation is principally plane, but the clearance used for the preform introduction caused a width increase of 5 \( \pm \) 1/2\% \((W > w)\). Dimensioning of the preform by scaling-up the final shape in proportion to the densities, based on width flow, gives:

\[
\frac{h_B}{H_B} = \frac{h_T}{H_T} = \frac{h_O}{H_O} = \frac{\rho_M}{\rho_p} \cdot \frac{W}{w} \quad \text{[1]}
\]

the middle equality can be rewritten:

\[
\frac{h_T}{h_O} = \frac{H_T}{H_O} \quad \text{[2]}
\]

<table>
<thead>
<tr>
<th>Group</th>
<th>Preform dimensions (mm)</th>
<th>Fully dense dimensions (mm)</th>
<th>Parameter</th>
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<tr>
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<td>H_o H_T H_a H/a H_o/H_o</td>
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<td></td>
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<tr>
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<td></td>
<td></td>
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<tr>
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<tr>
<td>D</td>
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<td></td>
<td></td>
</tr>
<tr>
<td>E</td>
<td>10.00 5.00 7.50 0.50 7.85 6.00 4.85 0.765 0.65</td>
<td></td>
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Dimensions of preforms in mm:s versus dimensions of fully dense shapes.
The top of the tooth is contacted by a rigid punch in the scaled up preform, and material flows to the base region before full die contact is attained (figure 6a).

Figure 6. Initial contact situations for different preform height ratios teeth/base.

When the preform tooth height is smaller than that of the final part, material flows into the tooth from the base (figure 6b). The study showed that the height ratio giving the most homogeneous densification lies within these limits:

\[ \frac{h_T}{h_B} < \frac{h_T}{h_B} < \frac{H_T}{H_B} \]  

The right hand limit is defined by the absence of horizontal material flow during deformation but with a risk of folding at the tooth base corners. At the left hand limit the preformed teeth have the same dimensions as the tool, therefore no folding occurs but there is a substantial need for material flow from the base to the tooth to reach full density in the whole piece.

On the basis of this limit philosophy and the deformation densification pattern shown in figure 7 of a forging from the B-group, one can predict that a rounded design of toothbase corner should be used to diminish the risk of folding.
As illustrated by the upper right micrograph in the figure, a high stress level involving shear is favourable for pore closure. With this geometry however the risk of folding is apparent and one would therefore expect that a rounded tooth base corner should be advantageous in preventing the problem of folding. The optimization of such a rounded design however has not been studied in this work.

A point of interest is the importance of slip and porosity variation. For the preforms in group E, (figure 8) the initial contact results in a higher stress in the $\alpha$- than in the $\beta$-direction, confining primary densification to the $\alpha$-oriented region, figure 8a.
Subsequently, contact extends along the gear flank and results in surface densification. Material forced into the tooth is now predensified.

Densification results in a higher flow stress and a reduction in the remaining deformation necessary in the $\alpha$-direction. Coinciding with an increase in force due to extended die contact, there is an increase in the densification in the $\beta$-direction, figure 8c. The resulting structure is thus dense in the $\beta$-direction and contains "dead zones" at the tops of the teeth and bases adjacent to the indenter.

A clear effect of shear slip was found after forging when the preforms had been introduced into the die in an off-centred manner. Figure 9 shows the results of a very slight deformation of a specimen where the eccentricity was 0.5 mm (4% of tooth interval). The result indicates the importance of reasonable precision when introducing the preform into the die tool.
Die tool design

The search for design criteria applicable to powder preforms in hot forging has been the topic of interest of a number of investigators. Of these, Kuhn and Downey have published a number of papers in which, based on earlier experiments, present certain guidelines concerning preform design (16).

The conclusions of the present study show that the rules for preform design are determined by the risk of cracking and folding, and the localization of the remaining porosity.

Despite gently rounded die engraving, forging of wave profiles from prismatic preforms led to stress cracking. Due to favourable die geometry however the cracks were closed. Although apparently homogeneous the material must be tested for strength (e.g. in fatigue) before closed cracks should be accepted. In another case, crack closure was found to be more difficult. The risk of bad crack closure has later been confirmed in studies by Sjöberg and Fischmeister (17) and Awano et. al. (18).

Experiments with preshaped green bodies indicated damaging folding with a large base/height ratio, a tendency that diminishes and finally disappears when the ratio is decreased.

The structure development was strongly influenced by the preform height ratio with principally different patterns for initial die contact at the tops or bottoms of the preform teeth.
The latter is less favourable due to a high proportion of porosity in the central plane of the tooth which must be eliminated during the last stage of forging when the predensified base has become rigid.

Concentration of compressive stress as well as shear flow under hydrostatic pressure are found to be useful instruments in density localization. Wall effects due to friction or mirror symmetry however reduce the densification rate.

The most uniform densification is found with a moderate preform height ratio $|\frac{3}{j}|$. Moderate preform shapes might also cause the least die wear while other geometries might wear the tool heavily by the large flow of material under high pressure past the corner of the tool profile.

**Plastic flow in manufacture and FEM-analysis**

Numerical analysis of a powder-forging process is extremely complicated, and in practice impossible. Even with the simplifications commonly used in structural mechanics, the porosity complicates the calculations by introducing an unmanageable number of equations necessary for such an analysis. The forming process would therefore imply that empirical or computer-methods must be used.

Industrial powder-forgers (19) confirm that current empirical knowledge allows dimensional change of samples in production without great difficulty, but samples with new shapes still require certain development.

In order to study the use of a computer for optimizing the powder-forging process, an attempt to simulate plastic flow by FEM was undertaken in the fourth study (IV). Since no FEM-program for analysing plastic deformation of a compressible material was available, the calculations were performed using
an iterative method as follows:

Straining induces elastic stresses →
  → local plastic flow when the stresses
  exceed the actual yieldpoint →
  → local densification and thus local
  hardening where material is flowing →
  → input of new material data in the
  actual FEM-elements →
  → new set of calculations etc.

This investigation is based on a FEM-program for stress-strain
analysis of solid bodies. This program has been modified for
this particular use. The influence of porosity, pore size,
shape and distribution on Young's modulus, Poisson's ratio,
yield criterion and yield strength have been extracted from
different papers (I, 20 - 23).

In order to simplify the study, forging of the "D"-type was
chosen in which the shape of the preform teeth coincides with
the shape of the upper die. The geometry change under forging
can therefore be described as an upward transference of the
base line. The mirror-symmetrical shape of the body is divided
into FEM-elements and compression of material is represented
in the FEM-grid by densification of the different elements
supplied by volume reduction of the two basal element rows.
A picture of the densification development calculated is shown
in figure 10, where the density is expressed by shading.
Figure 10. Densification development simulated by computer calculations.
Though there are clear imperfections in the calculated density maps the results are mainly consistent with the true densification; compare e.g. figures 10e and 10g with figures 11c and 11d.

Figure 11. True densification development revealed by forging experiments.

Refined results by FEM-studies can be expected however when certain fundamental weaknesses of the method can be avoided. An example of this is transfer of material from dense elements to nearby less dense ones which occurs in practice. Lack of such an assumption is the reason for the development of the structure-zones where dense elements alternate with undensified ones.

Another refinement to be carried out is to allow the element nodes to slide along the die-surface lines. The analysis has clearly shown that the calculated shear stresses along the surface exceed the true forces of friction, and sliding might therefore take place in a true forging process.

The use of computer methods for powder preform design has also been a topic of interest for other investigators. Pillay (24) for instance has studied a CAD-method in which a complicated part in divided into sub-units, each of which can be managed by known methods.

However, a computer assisted design method, corresponding to the one presented in the present work, has not been found.
Monolithic and compound products

By powder forging, monolithic materials of full or almost full density can be produced. The technique involved also makes it possible to produce parts from combinations of materials, so called composites in which two or more materials are combined to give the product specific properties. The method even allows the production of parts in which different materials are macroscopically combined in the same body. The latter type of products, defined here as compounds, are designed to give specific properties to different sites in the parts produced. Advantages asked for are e.g. wear resistance, hardness and strength in combination with plastic deformability, toughness, machinability and weldability.

The difference between an ordinary composite and a compound is shown in figure 12.

![Figure 12. Examples of composite, left, and compound, right.](image)

An important point concerning the compound body is the influence of the interface boundary on the overall behaviour, and in the latter section of the work the influence on the mechanical behaviour of the interface is studied. The papers VI and VII deals with transversely and longitudinally oriented compound geometries respectively. As a reference for these investigations, the materials are also studied in monolithic form.

From experiments on hard-ductile laminates (e.g.25 - 27 ) it is known that interface boundaries can act as crack arresters and rise the impact strength to very high levels. In an attempt to achieve the lamellar effect in a transverse connection a distributed interface was choosen. Both sharp and diffuse interfaces have been examined in order to study the importance of controlling the interfacial structure.
Several patent applications e.g. 28 - 30 show the commercial interest for the method, with applications like gripping, cutting and forming tools, transmission parts and similar load-bearing components. For successful performance of such compounds it is important to ensure that the interface should not be the cause of failure.

Strength of dense materials

The main advantage in densifying powder products by hot forging is a substantial improvement in material behaviour. Whilst a porous sintered body is somewhat weak and brittle as a result of the crack nucleating effect of the pores, a hot worked dense material possesses substantial strength and ductility. These properties can, for a fully dense hot worked powder material, be equivalent to or very often be even better than the corresponding properties of a conventionally wrought material with the same composition. When apparent the resulting improvement in comparison with a normal wrought material is mainly a virtue of the fine grained and roughly isotropic structure of powder forged materials.

The materials here studied range from soft, ductile mild steel to highly hardened high speed steel in a very brittle condition. As a result of hot working however, all the structures investigated are essentially pore-free and the performance of the materials is therefore comparable with the same materials in as wrought condition.

For the design of compound products it seems to be important to avoid skew forces to occur, thus therefore the transverse and lengthwise geometry types of combination seems to be most interesting. The transverse type of combination is least affected by skew forces and is also advantageous in the sense that it is easily produced. A sharp interface is found to be advantageous in comparison with a disordered one due to a better impact strength. The explanation to this is to be found by that a mixed structure will neither be fully ductile
nor fully hard. Therefore the structure produced will not reach the impact strength of a pure ductile or the strength of a fully hard material.

In a longitudinally combined compound loaded in tension the fracture development will take place perpendicularly to the interface. An important advantage of this type of combination therefore is the crack arresting effect, which effect can be used in several applications. E.g. in tensile straining the crack arresting effect will act as an assurance against brittle fracture.

When using the hard phase as a load bearing surface, the fracture blocking effect of the mild phase can be used as another design facility. In bending for instance the hard material facing the applied load will not only be advantageous by sustaining plastic deformation of the loaded surface but also by the fact that the hard phase material will influence the localisation of the neutral stress level in that sense that it under bending will be moved towards the loaded surface. As a result of this the ductile material zone will be strained more uniformly which gives an improved total strength against bending.

**Plastic deformation of compound parts and FEM-analysis**

Compound parts divided by both transverse and longitudinal planes were studied in order to examine the interface behaviour.

Transversely divided parts were studied using tensile testing and impact testing. Test pieces with both sharp and diffuse interfaces were investigated.

The parts examined under tensile testing showed that the combination of a soft steel with a hardened steel raised the total strength of the part. Sharply and diffusely interfaced parts was found to be roughly equivalent in tensile testing, and the fracture consequently occurred in the weak zone of the compound bar.
In impact testing it was found that compound bodies typically exhibited an impact strength between that of the monolithic materials used. Diffusely combined bars were found to give lower impact values than bars with sharp interfacial planes. From these experiments it was evident that a sharp interface boundary should be advantageous in comparison with a diffuse one.

An FEM-study was undertaken in order to reveal the stress and strain pattern developed under the tension. The analysis confirmed that no serious stress concentrations as result of the interface plane between the adjacent structures should be expected. The FEM-study also confirmed that fracture occurs in the weak zone at some distance from the boundary.

Longitudinally divided parts were also tested in tension; one type in pure tension where the development of a transverse crack was studied, and one type in bend testing. The tensile test part took the form of a flat bar with hardened high speed steel on each side of the bar and a section of mild steel in the centre. The bar was loaded in tension until brittle fractures occurred in the hard zones of the test section. After these primary fractures, a secondary ductile fracture took place in the mild steel zone and the bar finally failed.

Although there was extremely large local deformation in the ductile zone close to the interface, no delamination between the two zones was found. One can therefore expect that laminated bodies of highly hardened and mild steel may be successfully developed for uses in which the load direction is the same as in the test piece.

An FEM-study was undertaken in order to obtain a rough idea of local stress and strain at the crack tip. The results of this study naturally show a local concentration of stress and strain in this region. No shear stress however is found that would affect the strength of the interfacial zone. In general the results of the FEM-study coincided very well with the test results.
Double layer bars of hardened steel and mild steel were also tested in bending. The mild steel was located at the elongating side of the bar and was elongated in plastic deformation during the test. The hard zone was strained elastically to fracture which was initiated in the hard phase of the boundary zone. The development of the hard phase fracture was then associated with a plastic strain of the ductile phase, figure 13.

![Figure 13](image13.jpg)

**Figure 13.** Bend test bar of the 2-layer material. The ductile part is stretched in the strained zone while the hard part is almost completely broken. 3X.

From figure 13 one can see that the interfacial zone is not separated even with a large local tension in the ductile zone.

To balance the tension in the extended zone also a corresponding compression is developed at the top surface. As a result of the compressive character this stress does not lead to fracture. Near the centre section of the hard phase below the point of load application a local stress minimum therefore occurs.

On studying the results of the bend test some interesting intervals of the tensile stress distribution could be found, figure 14.
From figure 14, it can be seen that the hard phase is principally stressed in compression but that tensile stresses occur locally after plasticising of the mild steel phase. Tensile stresses in the hard phase are then set up at the interface and the stressed area expands during continued deformation. The maximum value is reached just before the crack initiation, corresponding to figure 14d.

When once started the crack propagates very fast perpendicularly to the stress direction until force balance is achieved for the actual geometry. Later on the crack opening is slowly extended during the continued bend test.

During continued bending the portion of hard phase loaded under tension is continually diminished towards zero. Up to the point of final fracture however, only the hard phase carries compressive forces.

On the basis of the fact that yielding starts in the ductile zone while the first fracture nucleates in the hard zone from the interface plane, comparisons with monolithic materials will be made. As a result of the geometry the primary fracture of a monolithic hard phase bar will occur after roughly half the displacement as for the compound bar. In a mild steel bar primary yielding is expected at the same place as in a compound.
bar, and the yielding then continues across the bar until the whole cross section is plasticized. The monolithic mild steel cannot sustain the applied load as effectively as the compound however, as the hard phase of the compound, also when partially fractured, still carries all the compressive force. The hard phase also carries a small portion of the tensile force, and the load carrying ability of a monolithic mild steel bar is therefore always lower than that of the compound.

Bending force as a function of the displacement was recorded from the test, and in the same diagram also the expected graphs of monolithic hard and ductile materials were superimposed, figure 15.

![Figure 15. Bend test graph of a two layer material, a), expected graph from bend test of monolithic hard, b), and ductile, c), bars.](image)

Shown by the figure, the work necessary to fracture a monolithic bar is always smaller than the corresponding work for complete fracture of a compound bar.
In comparison with a monolithic hard material the maximum fracture force of the compound is somewhat diminished, but the initial crack nucleates at a much larger degree of deformation. When the initial crack nucleates, the accumulated work is raised by about 120% and the structure will still not totally fail as the ductile phase is bearing a substantial load. The total work done in reaching the final fracture is about 6 times as large as the corresponding work for a monolithic highly hardened body of the same dimensions.

In comparison with a monolithic, mild steel, specimen, the compound has the advantage that it contains the highly hardened material on the surface that is exposed to the applied force. The resistance against surface deformation and wear is therefore substantially improved in comparison with that of a pure mild steel. At the same time the ductile material zone provides the toughness (ductility) under this type of force application.

Also in this case an FEM-study was undertaken in order to reveal the resulting stress and strain pattern.

Conclusions

Powder forging has been studied as a hot working method to produce fully dense high precision parts. In the first section of the work the topic is studied from a basic point of view. The process of deformation and the problem of attaining full density and final shape without serious faults arising from the hot working operations was then investigated, and here the risk of folding and stress cracking was found to determine the limits of the forging geometries available.

Presintering and forging temperature was also studied. Presintering of the preform before forging was then found to give no beneficial effects to the material investigated. On the contrary, the forging pressure is raised and the risk of oxidation increases when the pore channels are getting smoother and rounded off during presintering.
A main conclusion to be made is that densification is assisted by shear under compressive forces, and that increased forging temperature reduces the pressure required to reach a given strain and therefore will be in favour when as high as possible in the α-phase region. Presintering is of no value provided that a plain iron is to be forged. When presintering is undertaken in order to dissolve carbon or for any other type of diffusion alloying, preheating of coarse is necessary to reach the wanted level of material strength.

The optimization of preform design is found to be mainly a question of prohibiting forging faults to arise.

In the second part of the work a specific application is studied in which two different materials are combined into compound bodies. In this study the influence of the interface between the two materials was investigated. The main conclusion to be achieved is that in tensile straining the strength of the interface is quite sufficient when distorted as well as when sharp, but that the distorted interface might be less advantageous in impact.

For compound applications with the interface oriented parallel the force the interface boundary can act as a valuable crack arrester thus resulting in improved toughness. No failure of the interface boundary was found. A certain advantage of compound materials is demonstrated in bending with load applied at the hard surface of the specimen whilst the ductile zone was confined to the elongating region. In such a geometry the compound material was a substantial improvement in comparison with both monolithic hard and soft materials.

Summarising the whole work it was found that valuable products can readily be made by powder forging when using the guide lines for preform and product design outlined.
Acknowledgements

This work has been in progress for a very long time, sometimes intensively and at other times only periodically. It was started at Chalmers University of Technology, and followed me under my carrier at Linköping University and is now concluded at the University of Luleå. During this time a number of people have lent support in the form of different types of assistance; metallographic work, calculations, typing etc, and a list of all of them to whom I would like to express my gratitude is too long to present.

There are however three people without whose enthusiastic engagement the work would not have been performed. Professor Kenneth Easterling initially gave me the interest for scientific studies, Professor Hellmut Fischmeister has been my counsellor for the first part of the work and Professor Krister Källström finally has given me the motivation and spirit to terminate the thesis.

To all from whom I have received support I wish to express my greatful thanks.
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DEFORMATION AND DENSIFICATION OF POROUS
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H.F. Fischmeister, B. Arên and K.E. Easterling

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DEFORMATION AND DENSIFICATION OF POROUS PREFORMS IN HOT FORGING*

By H. F. Fischermeister,† B. Arén,‡ and K. E. Easterling†

ABSTRACT

Observations are reported on plane-strain upsetting, between unlubricated flat punches, of powder preforms of varying density at 1160°C (1435 K). Macroscopic deformation behaviour is characterized by the pressure/strain relationship and by the ratio of lateral to vertical flow. Increased preform porosity affects lateral flow similarly to increased friction in forging of dense material.

Densification is studied as a function of strain and pressure. Lower preform density requires not only larger strains but also higher pressures for a given final density.

The density distribution in typical forgings is charted by means of hardness measurements. Zones of incomplete densification are revealed where local pressure was reduced by lack of constraint, or where strain was impeded by friction effects.

I. INTRODUCTION

The hot forging of powder preforms is a rapidly developing technology, but very little information about the process has been published. A useful review of certain technical and economical aspects has recently been given by Hirschhorn and Bargainnier and a few reports of industrial experience with the forging of sintered preforms have appeared. However, no research on the basic processes of deformation and densification of hot porous preforms has been reported so far, though the cold deformation of porous iron has recently been studied by Kuhn et al. and Antes.

At present the design of tools and preforms for sinter forging is entirely a matter of trial and error. A rational approach to the problem would require a quantitative description of the deformation and densification behaviour of porous metals at high temperatures and strain rates. The present work is intended as a first step in this direction.

Beside the basic question of densification, it appears especially useful

† Chalmers University of Technology, Gothenburg, Sweden.
‡ POWDER METALLURGY, 1971, Vol. 14, No. 27
to study the points in which the deformation of a porous metal deviates from that of dense material. The theory of conventional forging is concerned exclusively with incompressible materials. To predict the flow of a porous material, the simultaneous decrease of volume must be incorporated. Our study was directed toward a type of deformation which is well understood in conventional forging theory, i.e. upsetting between flat dies in a plane-strain configuration. The strains investigated do not produce complete densification. Production forging of powder preforms is normally carried out in closed dies and with the aim of achieving nominally full density. However, upsetting between flat dies is applicable as a model for the initial stage of closed-die forging until the lateral flow of the preform forces the material against the die walls. The mode of initial material flow is an important consideration in the choice of preform and die geometry. Dead zones created during initial deformation in which densification lags behind other regions may be difficult to compact later on, when they are enclosed by a shell of denser and stronger material. This directs special interest to the study of local density distributions resulting from different modes of deformation.

The present paper is a preliminary report which leaves many important areas unexplored. Both upsetting and closed-die forging are being investigated in more detail in the continuation of this work.

II. EXPERIMENTAL PROCEDURE

Two types of deformation were studied, namely (a) pure compression by upsetting between flat dies and (b) shear deformation. The two techniques are illustrated in Fig. 1.

The dimensions of both the upsetting and shear specimens were 10 × 10 × 100 mm. The bars were made by compacting carefully weighed (± 1%) amounts of Höganäs HC 100.25 iron powder without lubricant admixture in a floating-die tool. Only bars compacted to within ± 1% of the nominal height were used. Four series of bars were made with densities between 4.885 and 6.250 g/cm³. Hereafter, densities are expressed as relative densities \( \rho^* = \rho/\rho_M \) or porosities \( \Pi = 1 - \rho^* \), where \( \rho_M = 7.80 \text{ g/cm}^3 \) (\( \rho^* \) and \( \Pi \) in per cent.).

The bars were heated to 1200°C (1475 K) by slowly introducing them into the hot zone of a tubular furnace with a hydrogen atmosphere (15 min to reach temperature), and holding them there for 6 min. No prior sintering treatment was employed. Transfer of the specimens from the furnace to the die took between 2 and 3 s, during which the temperature dropped by ~40 degC, as recorded by a thermocouple
embedded in a typical test-bar. Only in a few exceptional cases were signs of internal oxidation seen in the densified zones of the forgings.

Forging was carried out on a 250 tonf toggle-press fitted with plane-ground dies. The lower die was equipped with a force transducer and the moving upper die with a displacement transducer. Signals from these were recorded photographically from the screen of an oscilloscope. The oscillograph record of a typical force cycle is shown in Fig. 2.

In upsetting long bars between flat dies, as shown in Fig. 1(a), a plane-strain condition is realized in the centre cross-section. All deformation measurements as well as metallographic studies were performed on
Densification of Porous Preforms in Hot Forging

specimens taken from that section. The plane-strain condition results from the restraint of longitudinal material flow exerted by the friction between tool and workpiece surfaces. The upsetting of elongated bars is a well-established technique for plane-strain experiments in conventional forging research. In the case of our porous preforms, the validity of the plane-strain condition in the centre section is verified by the observation that the increase in length of the bars during upsetting was always much less (of the order of a few per cent. only) than the increase in width, up to the highest strains investigated.

No lubricant was used between the dies and the preforms.

III. Macroscopic Deformation Behaviour

1. Pressure/Strain Relationship

Fig. 3 shows the forging pressure necessary to produce a given strain. The maximum force recorded in the forging cycle was divided by the measured area of the deformed specimen to derive the true mean pressure. The measurements on the preform series with intermediate porosities are less certain than the others.

The deformation behaviour of preforms of varying density becomes more similar at high strains. However, all porous preforms are clearly more easily deformable than fully dense material. (The fully dense specimen was produced by isostatic cold compaction, sintering for 14 h 11—p.m.)
in hydrogen at 1450°C (1725 K), and hammer-forging in a mild-steel can, followed by machining to the starting dimensions used for the porous bars.) The initial portions of the curves in Fig. 3 can be considered as (compressional) stress/strain diagrams of the porous preforms; later the friction at the tool surfaces produces deviations from the true stress/strain curve. The yield pressures of preforms of varying porosity can be determined by extrapolating the curves of Fig. 3 to zero strain. An independent determination is possible by extrapolation of the density/pressure measurements presented in Fig. 8. In analogy to Eudier’s model for the tensile strength of porous solids, the yield pressure should obey the relation

\[ P_y = 1.15 \left( \sigma_y - k_1 f_p^{2/3} \right) \]  

where \( \sigma_y \) is the yield stress of the pore-free metal, \( k_1 \) a constant, and \( f_p \) the volume fraction of pores. The factor 1.15 derives from von Mises’ yield criterion in plane strain. The yield stresses extrapolated from Figs. 3 and 8 agree well and their porosity-dependence fits equation (1) (see Table I).
Densification of Porous Preforms in Hot Forging

Table I. Yield Pressures of Porous Preforms

<table>
<thead>
<tr>
<th>Preform Porosity, %</th>
<th>Yield Pressure, tonne/cm²</th>
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<tbody>
<tr>
<td></td>
<td>From Fig. 3</td>
</tr>
<tr>
<td>37.4</td>
<td>0.58</td>
</tr>
<tr>
<td>32.7</td>
<td>0.63</td>
</tr>
<tr>
<td>25.7</td>
<td>0.84</td>
</tr>
<tr>
<td>19.8</td>
<td>0.88</td>
</tr>
</tbody>
</table>

Kuhn et al.³ and Antes⁵ have shown that in cold upsetting of porous preforms a strain-hardening equation of the type

\[ P = k_2 \varepsilon^n \]

is obeyed. In our hot-forging experiments, a relation of the type

\[ P - P_Y = k_3 \varepsilon^n \]

is obeyed initially, i.e. to strains <0.5. Above this level, the apparent hardening index \( n \) increases with increasing strain (i.e. increasing densification and friction effects). However, the strain-hardening law of equation (3) is characteristic of the regime where normal slip of dislocations is the predominant deformation mechanism. It can hardly be expected to hold for hot-forging, where thermally activated dislocation movements (mainly cross-slip) must play the dominant role. Hence, the values of the strain exponent \( n \) (0.60–1.21) are quite different from those encountered in strain-hardening at room temperature.

Fig. 3 shows that the apparent hardening coefficient \( (dP/d\varepsilon) \) (increase of deformation-resistance due to the combined effects of densification and friction) increases with preform porosity, as it did in the studies of Kuhn et al. and Antes. The explanation of this behaviour lies in the fact that the flow stress of a porous body increases with densification. As will be shown later (Fig. 6), preforms of high porosity densify faster when strained than those that have a smaller fraction of collapsible pores left. In doing so, they gradually approach the flow stress of originally less porous preforms subjected to the same strain, though they never quite reach the same level.

2. Lateral Flow

For a completely dense material deformed in plane strain, a decrease in height results in an equal increase in width, to satisfy the condition of constant volume which is imposed by the incompressibility of the material. In a porous material subjected to the same type of strain,
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the volume decreases during deformation and the width increment
becomes less than the decrease in height.

The ratio of width-to-height deformation can be characterized by an
apparent\(^\dagger\) plastic Poisson's ratio

\[
v_p = \frac{1}{2} \frac{d \ln w}{d \ln h} = \frac{1}{2} \frac{\varepsilon_w}{\varepsilon_h}
\]

For an incompressible body, \(v_p = 0.5\); with increasing porosity, we
expect \(v_p\) to fall below 0.5. Fig. 4 shows a plot of \(\varepsilon_w\) vs. \(\varepsilon_h\) for preforms
of varying starting density. The apparent plastic Poisson's ratios
determined from Fig. 4 for various levels of densification are given in
Table II.

<table>
<thead>
<tr>
<th>Table II. Apparent(^\dagger) Poisson's Ratios</th>
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<tbody>
<tr>
<td>Preform Porosity</td>
</tr>
<tr>
<td>------------------</td>
</tr>
<tr>
<td>Initial ((\varepsilon = 0))</td>
</tr>
<tr>
<td>Linear stage (v_p)</td>
</tr>
<tr>
<td>(Values in parentheses give extension of linear stage in terms of (\varepsilon) and (\rho^*))</td>
</tr>
<tr>
<td>(\varepsilon &gt; 0.50)</td>
</tr>
</tbody>
</table>

The strongest deviation from incompressible behaviour is found in the
initial stage of upsetting, with values of \(v_p\) as low as 0.13, indicating
that material flows at first predominantly in the direction of punch
movement, with little widening of the compact. As the density in­
creases, lateral flow is enhanced. This lateral flow, which is desirable for
good filling against the die walls, will be obtained most easily with pre­
forms of low porosity. It is worth noting that the effect of high preform
porosity will persist even at high degrees of compaction.

Fig. 4 shows a linear stage in the \(\varepsilon_w/\varepsilon_h\) plots at large strains with
higher values of Poisson's ratio (see Table II), indicating that lateral
flow now becomes nearly as great as flow in the direction of the applied
pressure. Table II states the strain and density levels at which this
stage commences. It is worth noting that preform porosity is reduced
by roughly one-half before lateral flow reaches its full extent. This
tendency of porous preforms to deform mainly in the direction of pressure
is an important advantage from the point of view of die design for flashless forging.

It must be remembered that our experiments were made without
\(^\dagger\) Friction effects are not eliminated in the experiment.
Densification of Porous Preforms in Hot Forging

3. Friction Effects

For fully dense materials, the shape of the pressure/deformation curve in upsetting can be calculated from the balance between applied stress and the friction between tool and workpiece. Under plane-strain conditions, the pressure distribution over the workpiece surface is given by

\[ P = P_y \exp \left( \frac{\mu w - 2x}{h} \right) \]  

where \( x \) is the coordinate in the direction of lateral flow, with the origin in the centre of the workpiece. Equation (5) predicts a pressure peak in the centre, followed by a decrease to the level of the flow pressure at the edge \( (x = w/2) \). The mean pressure is obtained by integration of equation (5) between \( x = -w/2 \) and \( x = +w/2 \), giving

\[ \bar{P} = \frac{P_y h}{\mu w} \exp(\mu w/h) - 1 \]  

Fig. 5 shows the theoretical variation of forging pressure with \( w/h \) according to equation (6) for a few values of \( \mu \) (broken lines) superimposed on our measurements on porous preforms. Considering rough trends only, increased porosity has the same effect as increased friction. This may appear surprising until one remembers that energy consump-
tion for densification must have an effect on the lateral flow similar to that of energy consumption by friction and that high porosity gives greater scope for this. However, the decisive process takes place in the bulk of the material rather than at the surface. We are at present working on a modification of the simple theory of upsetting to allow for the compaction of each volume element as it is transmitting stress.

IV. DENSIFICATION

To establish how much strain is necessary to obtain a desired density, the mean density of forged specimens was calculated from the known starting density and from the change in cross-sectional area at the lengthwise centre of the bar. This method of density determination rests on the assumption that material flow in the lengthwise direction is negligible near the centre. This condition is inherent in the assumption of plane strain at the centre section of the bar, the justification of which we have discussed earlier.

Densification is very nonuniform throughout the cross-section. The measurements reported in the following figures refer to the overall mean
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density in the region of plane-strain deformation. The local variations
of density in this region are reported later (cf. Figs. 10–13).

1. Densification and Strain

Fig. 6 shows the densification produced by strain in preforms of
various porosity levels. Although highly porous preforms densify
more than less porous ones, they do not reach the same level within the
strain range studied (production forging in closed dies will normally
employ strains in the lower part of this range).

The results in Fig. 6 can be summarized by the expression

\[ \bar{\rho} = k_p \cdot \varepsilon^{n_p} \]

with values for the constant as stated in Table III.

The densification exponent \( n_p \) shows a very nearly linear dependence
on preform density (\( f_p^0 = \) volume fraction of pores in the preform):

\[ n_p = 0.41 f_p^0 - 0.01 \]
Equations (7) and (8) are presented merely as a convenient way of summarizing our measurements. There is at present no physical model to justify these special functional relationships. In fact, the densification results can be represented almost equally well by a relation of the type

\[ I = I_0 \cdot \exp(-ks) \]  \hspace{1cm} (9)

An important question in die and preform design is the relation between preform and final density at various strain levels imposed by the shape relation between the preform and the finished forging. This information is summarized in Fig. 7.
2. Effect of Pressure on Densification

Fig. 8 shows the densification of the preforms with increasing pressure. The curves differ from the compressibility curves obtained in normal cold compacting by the existence of a yield pressure, below which no flow of material and hence no densification occurs.

The effect of preform density becomes more noticeable as the pressure increases. Denser preforms require a higher pressure to achieve a given strain than more porous ones (Fig. 3), but the density produced by this strain will be much higher (Fig. 6) for the denser preforms than for the more porous ones.

The influence of preform density becomes very great at large strains: the pressures required to produce 95% density from the lowest and the highest starting level are in the ratio of 1.75:1, and this ratio increases rapidly as more complete densification is attempted. This effect, however, is peculiar to plane-strain forging where lateral flow is restricted.
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only by friction and by the deformability of the porous material; in
closed dies with a sharp restriction to lateral flow the influence of pre­
form density would be expected to be much less, and this is borne out
by experience with production-type forgings.

In normal cold compacting, the pressure/porosity relationship is often
of the type

\[ \Pi = \Pi_0 \cdot \exp(-\kappa P) \] (10)

where the compressibility index \( \kappa \) is related to the plastic properties of
the powder. It is well known that though this relation is only a rough
approximation it holds quite well over the limited range of densities
that are of interest in compaction under production conditions. Our
hot-forging data do not comply at all with this relation. A fairly good
fit is obtained with the expression

\[ P = k_p \cdot \Pi^{-n_p} \] (11)

where \( k_p \) and \( n_p \) are constants for any given preform density. Their
values are given in Table IV.

Again, it must be emphasized that equation (11) is not indicated by
any theoretical model, but is offered as a convenient though approximate
numerical description of the hot-compactibility of our preforms.

<table>
<thead>
<tr>
<th>Preform Porosity, %</th>
<th>( k_p ), tonne/cm(^2)</th>
<th>( n_p )</th>
</tr>
</thead>
<tbody>
<tr>
<td>19.8</td>
<td>0.68</td>
<td>0.472</td>
</tr>
<tr>
<td>25.7</td>
<td>4.62</td>
<td>0.549</td>
</tr>
<tr>
<td>32.7</td>
<td>7.25</td>
<td>0.703</td>
</tr>
<tr>
<td>37.4</td>
<td>14.0</td>
<td>0.909</td>
</tr>
</tbody>
</table>

V. Density Distribution

The local-densification process was charted by means of a dense grid
of Brinell-hardness impressions made on the centre cross-sections of
forged bars. The relation between porosity and hardness (Fig. 9) was
established on specimens cut out from the centre regions of forgings
which previous microscopic investigation had shown to be of homogene­
ous density.

Fig. 10 shows the hardness distribution of a forging upset to a strain
Densification of Porous Preforms in Hot Forging

of \( \varepsilon = 0.51 \) together with micrographs of the porosity in regions of particular interest. Fig. 11 shows the porosity of corresponding regions at higher strain. The density distribution is characteristically different from that obtained in cold compacting loose powder in a die, in that

The centre is a region of high densification. The slip-line-field analysis of plane-strain forging shows that the centre is a region of intense deformation, whereas in powder compaction it is a dead zone. Friction at the punch surfaces produces dead zones at the centre of the top and bottom surfaces both in forgings and in powder compacts. Zones of minimum densification appear at the centres of the lateral surfaces owing to the absence of restraint in forging and the diminution of transmitted pressure by friction in cold powder compacting.

Fig. 12 shows the densification of the homogeneous centre region as a function of strain. Densities were determined by weighing plane parallel rectangular slabs machined from the centre portions of bars. The strain values are those that were applied to the bar as a whole. Practically full density is reached at \( \varepsilon \approx 1 \), while the mean density of preforms of similar starting density at this strain is only 97%.

In the development of powder forging as a production technique, a trend towards preforms of simple shape is expected. This will often call for large shear deformation during forging. There is also a feeling
Fig. 10. Isohardness contours (reflecting density distribution) and microstructures in centre cross-section of bar upset to $\varepsilon = 0.51$. Micrographs $\times 80$. 
Densification of Porous Preforms in Hot Forging

Fig. 11. Microstructures in centre cross-section of bar upset to $\varepsilon=1.20$. Micrographs $\times 80$. 
among practical workers in the field that high shear deformation is
beneficial or even necessary for attaining good particle bonding and
high strength, especially impact strength. An attempt was made to
study this mode of deformation with the tooling described in Fig. 1(b).
The amount of shear was controlled by the sideways separation between
the upper and lower punch, and by the length of stroke. Bars with
inlaid nickel wire nets were used to indicate the local shear produced,
and the hardness distribution of similarly deformed bars was measured
to chart the densification achieved. Some preliminary results are
shown in Fig. 13.

In contrast to upsetting between flat dies, the centre of the shear
region is now a zone of minimum densification. The situation bears
some resemblance to that encountered in shearing metal sheet with large
tool clearances. For this case, the work of Chang and Swift indicates
that transverse cracks will form in the centre zone. The existence of
this zone of incomplete densification must be borne in mind in the
design of tooling and preforms for sinter forging.
Fig. 13. Hardness contours (reflecting density distribution) in bars deformed to shear strain of $\gamma \approx 0.33$ and $\gamma \approx 0.60$ in the central fibre. The approximate shear values were determined from wire markers in bars forged under identical conditions.
A significant difference in deformation behaviour between porous and dense materials is found in the reduced lateral flow caused by the coupling between densification and deformation. Densification is most rapid in the initial stage of deformation, where, in consequence, lateral flow is almost absent. This is a welcome effect from the point of view of forming technology. In closed-die forging, it will delay the build-up of die-wall friction, which forms the main obstacle to the attainment of high densities in conventional powder compaction. If it were desired to forge powder preforms in open (flash-producing) dies, the initial uniaxial deformation would diminish the problem posed by the volume difference between the preform and the finished part.

With respect to the pressure necessary to achieve a certain lateral flow, e.g. for filling a die cavity, the coupling between deformation and densification produces a reduction of lateral flow similar to that caused by an increased coefficient of friction in forging dense material.

The pressure/strain relationship for porous preforms does not follow a normal strain-hardening law, in contrast to the observations of other authors\(^4\)\(^5\) in cold upsetting. The resistance to deformation increases continuously with strain, owing to the increase in flow stress that follows on reduction of porosity.

Original preform density has a strong and persistent effect on the deformation characteristics. Thus, the deformation behaviour is not determined solely by the momentary (overall) density of the compact, but also by the way in which this has been reached. The same is true for the compactibility of hot preforms, as indicated by the pressure/density or strain/density relationships.

The process of densification in itself is basically different from that encountered in the cold compaction of loose powders in a die with regard to both pressure requirements and density distribution. In contrast to compacts made from loose powder, forged specimens have maximum density in the centre, owing to the high pressure and strain to which this region is subjected. Zones of incomplete densification arise where the effective pressure is reduced by the lack of geometric constraint, or where deformation is kept at low levels by the effects of friction and geometry. Forging in closed dies will eliminate the dissipation of pressure in the lateral regions, but not the dead zones that friction creates at the centres of the top and bottom surfaces.

REFERENCES
5. H. W. Antes, ibid.

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THE INFLUENCE OF PRESINTERING AND FORGING TEMPERATURE IN POWDER FORGING

B. Arén, L. Olsson and H.F. Fischmeister

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The Influence of Presintering and Forging Temperature in Powder Forging

B. G. A. ÅREN*, L. OLSSON* and H. F. FISCHMEISTER*

Abstract:
The influence of temperature and presintering on the deformation and strength of iron powder preforms is studied in open die forging between flat plates. Approximately plane strain conditions were ensured by experimental geometry. Increased preform temperature reduces the resistance to deformation and decreases the strength of forgings by way of grain size changes. Contrary to other authors' reports, no special effects were found in the vicinity of the γ-α transformation despite careful study of this region. Strong presintering (beyond the range of technical sintering times) increases the deformation resistance of powder preforms.

The density-strain relationship is not affected by either temperature or presintering, nor does the lateral spreading of the preforms during hot compaction vary much with either factor. Oxygen penetration into the preform depends strongly on preform porosity, but becomes negligible (under experimental conditions) above 80% of theoretical density. Prolonged presintering increases the risk of internal oxidation by producing smoother, rounded pore channels allowing faster diffusion of the protective atmosphere.

Introduction

This paper reports a study on powder forging with flat, open dies under conditions approximating plane strain. Although most industrial powder forging is done in closed dies, the study of plane strain deformation has some justification as a basis for the understanding of more complicated deformation modes. In the early stages of deformation in a closed die — before lateral flow has brought the compact in touch with the die walls — the deformation has great similarity to the plane strain mode. These early deformation stages are of considerable importance for the design of preforms and dies. The influence of presintering and of forging temperature, which is the main object of the present investigation, can be studied very well in open die forging. Finally, it is of direct interest for continuous hot compaction processes. Powder rolling, for instance, implies essentially a plane strain deformation.

The early literature on powder forging has been reviewed by Hirschhorn and Bargainier1, Several further papers were presented at the 1970 International Powder Metallurgy Conference in New York2, and at the symposium on powder forging3 arranged by The British Powder Metallurgy Joint Group in the autumn of 1970. Most of this literature is concerned with the feasibility and technology of producing components. For only a few papers have dealt with the basic deformation aspects of powder forging.

The deformation of cold porous preforms has been studied by Kuhn, Hagerty, Gaigher and Lawley4, Kuhn and Downey5, and Antes9.

Studies of some basic aspects of actual hot forging have been presented by Davies and Dixon7 and by Fischmeister, Arén and Easterling9. The present work is a continuation of the last mentioned report.

Experimental

Our experimental procedure has been described in detail in an earlier report6, so that only a summary is needed here. Test bars 10 x 10 x 100 mm were cold compacted from unlubricated Hoeganaes HC 100.25 iron powder to various density levels between 4.855 and 6.250 g/cm³. In the following, densities are expressed as relative densities \( \rho = \rho/qM \cdot 100\% \), related to an assumed theoretical density \( qM = 7.80 \text{ g/cm}^3 \). They were heated to forging temperature in a tube furnace. The time to reach temperature was 15 min. This was followed by a 5 min soak. Transfer from the furnace to the die took about 3 sec, during which time period the specimens were protected from oxidation by the hydrogen effusing from its pores. The temperature drop in transfer (as recorded by a thermometer embedded in a typical test bar) was of the order of 40°C. Part of the material was presintered batchwise in a hydrogen tube furnace, where the bars were held for either 4 or 24 hrs at 1100°C, followed by cooling in hydrogen. As expected, the effects of presintering were small so that sintering times were necessary to demonstrate them.

Fig. 1 Upsetting geometry.
Upsetting (Fig. 1) was carried out in one stroke in a transducer instrumented toggle press. After the stroke, the specimens were allowed to cool on the die platen. The tool surfaces were not preheated and no die lubricant was used. The geometry of the specimen end of its deformation provides for practically plane-strain conditions in the center section (Fig. 1). All density determinations, metallographic and fractographic observations reported relate to this center section.

Fracture experiments were made in a charpy tester with a 10 kg pendulum with bars that had been given a V-notch after forging.

**Macroscopic Deformation and Influence of Temperature**

**Densification**

In our earlier study it has been shown that plane upsetting between open dies gives highly localized densification, associated with regions of high local strain (Fig. 2). The center is the region of highest density. Following the flow line pattern, an X-shaped band of high density emanates from it, leaving low density zones at the top and the bottom and especially at the sides. In the present work we have studied the progress of densification in the center zone as compared to the overall density values reported earlier, using slabs machined from the center regions. The results are shown in Fig. 3.

It is borne out that the center region densifies much more quickly than the rest. However, densification at the center is still slower than what would be expected in closed die forging. The difference is due to weaker lateral restraint in an open die, where friction is the only restraining factor. In general, the difference will be expected to be a function of the size and geometry of the compact.

Forging temperature has a negligible influence on the densification produced by a given strain (Fig. 4), although the pressure to produce this strain of course does depend on temperature, as will be shown in the following section.

**Forging pressure**

Special interest attaches to the temperature dependence of forging pressure. Davies and Dixon have reported an abrupt change in the room temperature tensile strength of iron powder bars forged above and below the α-γ transformation. Since this might indicate a strong effect of the phase transformation on forgeability, we made an especially careful study of the temperature region 820-1080°C. Preforms of 74.3% density were used. The actual forging temperature was determined before each test by means of a dummy specimen with embedded thermocouple, heated and transferred to the die platen in the same fashion as the preforms. Results for forging to constant strain (ε = 0.85) are shown in Fig. 5. A smooth decrease of forging pressure with temperature is found, with no particular variation in the vicinity of the transformation temperature.

We also extended our study of the relation between forging pressure and strain to lower strain values. Figure 6 shows our earlier results (obtained at 1160°C) together with the present measurements at a slightly lower temperature. We now find that at small deformations, the pressure-strain relationship is concave towards the strain axis. It is interesting to compare this result with the stress-strain curves obtained by Kuhn and Downey for cold repressing presintered compacts in a closed cylindrical die, which show similar initial curvature. As will be shown in the following section, there is very little lateral spreading at strains below 0.2 in open die forging. Thus the similarity to closed die forging is understandable.

**Lateral flow**

It was found in our earlier work that porous preforms show less lateral spreading on upsetting than compact specimens. The reason is, of course, that part of the work of compression is used to reduce the pores in the material. As a matter of fact, the reduced lateral

![Fig. 2 Density variation in the cross-section of a specimen strained to ε = 0.3 as reflected by Brinell hardness (from ref. 1).](image)

![Fig. 3 Center density in open die forging, compared to overall mean density and expected density in closed die compaction, vs. true strain. Forging temperature 1160°C.](image)

![Fig. 4 Densification vs. strain at various forging temperatures.](image)

![Fig. 5 Temperature dependence of the forging pressure required to produce a given strain (ε = 0.85).](image)
Mean forging pressure $P$ vs. strain according to earlier work, depending on original preform porosity. In the graph, the ultimate tensile strength and lower yield stress values obtained. A monotonous decrease with forging temperature is noted, resembling the change in forging pressure for constant strain (Fig. 5). The abrupt change reported by Davies and Dixon is not reproduced.

### Strength as a function of forging temperature

As mentioned earlier, Davies and Dixon reported an influence of forging temperature on the ultimate tensile strength of forged specimens at room temperature in the vicinity of the α-γ phase transformation. Since our own results show that temperature produces no abrupt change in forgeability, it remained to test the influence of temperature on strength directly. Test bars with gage section 2 mm by 6 mm were machined from the center portions of the forged bars, where density was reasonably uniform. The density was determined individually for the center portion of each tensile specimen. It ranged from 98.2 to 99.7% (as compared to 96% in Davies and Dixon's experiments). Figure 8 shows the ultimate tensile strength and lower yield stress values obtained. A monotonous decrease with forging temperature is noted, resembling the change in forging pressure for constant strain (Fig. 5). The abrupt change reported by Davies and Dixon is not reproduced.

### Effects of Presintering

Presintering can be expected to influence the deformation properties of a powder compact by producing stronger particle bonds and a coherent skeleton. Figure 10 shows the effect of presintering on the pressure required to produce a given strain. As expected, presintering...
Ultimate tensile strength $\sigma_{uE}$

N/mm$^2$

<table>
<thead>
<tr>
<th>Density (%)</th>
<th>Strength (N/mm$^2$)</th>
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<tbody>
<tr>
<td>98.0</td>
<td>340</td>
</tr>
<tr>
<td>98.5</td>
<td>330</td>
</tr>
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<td>99.0</td>
<td>320</td>
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<tr>
<td>310</td>
<td>99.5</td>
</tr>
<tr>
<td>320</td>
<td>100</td>
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</tbody>
</table>

Fig. 9a Strength vs. density of specimens forged to constant strain ($\varepsilon = 0.05$). Filled points in fig. 9a are for those specimens for which grain size has been measured (see fig. 9b).

Yield strength $\sigma_y$

N/mm$^2$

<table>
<thead>
<tr>
<th>Grain size ($\mu m$)</th>
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<tr>
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<td>260</td>
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<td>0.30</td>
<td>230</td>
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<tr>
<td>0.34</td>
<td>220</td>
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Fig. 9b Strength vs. grain size of specimens forged to nearly constant density at various temperatures.

Mean pressure $P$ torr/cm$^2$

<table>
<thead>
<tr>
<th>Presintering</th>
<th>Preform porosity (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>hrs at 1100°C</td>
<td>19.9% 32.7%</td>
</tr>
<tr>
<td>0</td>
<td></td>
</tr>
<tr>
<td>4</td>
<td></td>
</tr>
<tr>
<td>24</td>
<td></td>
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</table>

Fig. 10 Influence of presintering on deformability of porous preforms.

Mean relative density $\rho^*$

<table>
<thead>
<tr>
<th>Relative Density (%)</th>
<th>Mean relative density $\rho^*$</th>
</tr>
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<tbody>
<tr>
<td>60</td>
<td>0</td>
</tr>
<tr>
<td>80</td>
<td>0.8</td>
</tr>
<tr>
<td>85</td>
<td>0.6</td>
</tr>
</tbody>
</table>

Fig. 11 Influence of presintering on densification of porous preforms (circles: preform density 19.9%, triangles: 32.7%).

Fig. 12 Influence of presintering on lateral spreading.

Impact strength kgm/cm$^2$

<table>
<thead>
<tr>
<th>Presintering</th>
<th>Preform porosity (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>hrs at 1100°C</td>
<td>19.9% 32.7%</td>
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<tr>
<td>4</td>
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</tr>
<tr>
<td>24</td>
<td></td>
</tr>
</tbody>
</table>

Fig. 13 Influence of presintering on toughness of specimens forged to various densities.

Sintered or heated to forging temperature but not forged.

Forged
No presintering except for heating to 1100°C, immediately followed by cooling.

Scanning electron micro-fractographs of unforged specimens raises the resistance to deformation. The effect at very long presintering is appreciable.

The relationship between strain and overall density (Fig. 11) is not affected by presintering, despite the fact that the pressure to reach a given strain is increased. The insensitivity of the density-strain relation to both temperature and presintering would be trivial in closed die forging, but in open die conditions it indicates that the lateral spreading of the compact is unaffected by both parameters. With regard to temperature this has already been shown (Fig. 7b).

Direct measurements of the effect of presintering on lateral spreading are shown in Fig. 12. The influence is seen to be very limited.

An interesting effect is found in the impact strength of forged specimens (Fig. 13). The impact strength values are only qualitative, since the main intention in fracturing the specimens was to obtain surfaces for micro-fractography. The specimens — as forged — were given a 1 mm V-notch and then struck in a 10 kpm Charpy pendulum.

The specimen thickness under the notch root varied with the degree of upsetting, and only specimens that had not been forged at all had the regulation thickness of 10 mm.

In order to allow at least qualitative comparisons, the fracture energies were divided by the area under the notch, despite the fact that Charpy V-notch values are not truly proportional to the fracture area. With all these reservations in mind, the results nevertheless indicate a strong effect of presintering. The scatter band presents data for all specimens that had not been presintered — both forged and unforged ones. (The latter were merely heated to forging temperature).

At low densities, the impact strength curve is almost horizontal. Obviously, it is immaterial with regard to impact strength whether the final density has been reached by heavy cold compaction plus light sintering or by light cold compaction plus some hot deformation, as long as the density remains low. Quite a different trend is shown by the strongly presintered specimens. The impact strength of unforged bars is about doubled by long presintering. This is easily explained by the increased particle bonding and the rounding of the necks between the particles demonstrated by the SEM fractographs in Fig. 14.

Forging temp. 820°C 1015°C 1210°C

Cleavage fracture

Site of notch

Location of cleavage zones in fractured bars.

Cleavage region (with ductile portions). Predominantly ductile regions.

Fracture surface of a specimen forged to 85% density at 1160°C (SEM).
A great deal of the impact strength gained by presintering is destroyed by a slight hot deformation as shown by the initial drop in impact strength seen in Fig. 13. It must be concluded that part of the particle bonds formed during presintering are disrupted on initial deformation, and that appreciable plastic flow is required to repair this loss in strength.

Some observations on the general fracture characteristics of forged powder materials may be worth recording. Although the low values of impact strength and the macroscopic appearance of the fracture surfaces would lead one to classify all fractures as brittle, there is considerable ductile rupture on a micro scale. The relative amounts of local ductile and cleavage character vary throughout the cross section. With a low power microscope, one can identify zones of predominant cleavage character (crystalline appearance) especially in samples forged from high density preforms with a high degree of upsetting and at high temperatures. The cleavage zones occur in a region near the center of the bars (Fig. 15), indicating a correlation between fracture character and density.

High magnification SEM pictures reveal that the fracture character inside these "crystalline" zones is partly cleavage and partly intergranular (Fig. 16a), while outside them the fracture is microductile (Fig. 16b). At the boundary, there is a mixture of microductile and cleavage fracture.

Pores occupy a greater share of the fracture surface than what is accounted for by their volume fraction, indicating that crack propagation is associated with the pores.

**Oxidation**

In specimens that had not been forged to high density, the fracture surfaces revealed an oxidized zone, except for the face that had remained in contact with the platen after the forging stroke. When a specimen was turned over to another face immediately after forging, the oxidation pattern was shifted accordingly (Fig. 17). We therefore conclude that oxidation occurred only after the forging stroke. This is supported by the complete absence of oxide inclusions at the particle boundaries in high power micrographs of forged specimens.

Oxide shells of considerable thickness can be built up on the internal surfaces, as demonstrated by Fig. 18.

**Conclusions**

At the time when this work was begun, there was a great deal of discussion about possible beneficial effects of presintering. We believe that this study has shown that there is no such beneficial effect with regard to the factors investigated here. On the contrary, the forging pressure is raised and the risk of oxidation increases when the pore channels are getting smoother and rounded off according to Fig. 14a to c. There is no improvement in strength after forging to high strains and densities. Presintering seems to have only negligible influence on the lateral spreading behaviour of porous preforms.
When diffusion alloying is attempted the heat treatment should be given after forging unless excessive grain growth is a danger. Increased forging temperature reduces the forging pressure required to produce a given strain. No special effects were found in the vicinity of the α-γ phase transformation of iron. However, one important aspect of presintering has not been covered in this study. Work in progress at this department indicates that presintering can play an important role with respect to the ability of the preforms to sustain inhomogeneous deformation during forging. This will be the subject of a later report.

References:
MATERIAL FLOW IN THE POWDER FORGING OF A GEAR PROFILE
AS A FUNCTION OF PREFORM SHAPE

B. Arên

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Czechoslovak Socialist republic 1974,
Vysoke Tatry 1 - 3 Oct III.
MATERIAL FLOW IN THE POWDER FORGING OF A GEAR - PROFILE AS A FUNCTION OF PREFORM SHAPE

B. Arén
Linköping University
Department of Mechanical Engineering
Linköping, Sweden

Abstract
Powder preforms of unalloyed iron with differentiated shapes were forged to gear-profile form. The preforms, of 82% density, were preheated to 900°C and forged under plain strain conditions in a closed preheated (200°C) die.

The influence of tooth height vs total height of the preforms on the localisation of remaining porosity in the forgings, the risk for stress cracking and the tendency for folding were investigated. Densification during different stages of forging up to 94% relative density was studied by Brinell-hardness testing and by metallography. The geometry relation between the shape of preform and the shape of the final part, which yields the most uniform distribution of remaining porosity is developed.

The results are discussed in relation to the fundamental flow behaviour of porous materials. They indicate concentrations of compressive stresses as well as shear flow under hydrostatic pressure, resulting in local densification.
Introduction

The interest of powder forging branches into two main philosophies. One of them aims at complete density in order to reach the full strength of the material. The other philosophy aims at densities around 97% giving strength levels very near those of compact material, though with reduced ductility and impact strength. As compensation, lower forging pressure and reduced tool wear are expected.

The purpose of this work is to study the nonuniform densification that occurs during the early stages of forging for systematically varied preform geometries. The results will not be relevant for full-density forgings, where the last stage of the stroke will level out most of the nonuniformity studied here. Rather, the present results are relevant to powder forging primarily aiming at close tolerances, where a small amount of residual porosity is accepted.

The location of residual porosity to uncritical regions of the sample is of great importance and might be the best motive for studying the interplay between the preform and the final shape.

In that sense, however, as nonuniform flow also sets the conditions for the later occurrence of defects like folding and cracks, the results should be relevant to all types of powder forging.

Basic deformation behaviour of porous material

To lay a foundation for the study, some fundamental behaviour of porous materials will be discussed.

Due to the compressibility of pores, the plastic deformation is accompanied by volume change so that tensile stresses expand, and compressive stresses compress the material.

An interesting factor in the process is the proportions of shear and unaxial flow. A very rough idea of the densification produced by shear flow was developed by forging specimens with inlaid nickel nets to the geometry shown in fig. 1(1). In fact no pure shear flow will be found
in this experiment, especially not at the surface zones. At the centre, however, the degree of shear will be recorded with fair accuracy by the distortion of the nickel net. The densification can be related to the shear by the deformation of the grid elements.

Fig. 1 Densification measured by the change of grid element area \( A \) relatively \( A_0 \) versus shear \( \tan \theta \) (a) for a shear type forging (b).

It is a weakness of the experiment that it is not possible to separate the contribution to densification of shear and compressive flow and also that the net might not exactly follow the porous material. Even if the importance of shear for the densification process is not fully clarified, it can be shown that lack of shear may have important consequences in some particular deformation geometries. Notice the material flow against a wall that gives "dead" zones of high porosity.

Another case seems to occur in the forging of shapes containing planes of symmetry where two material flows impinge as against an imaginary wall. Such an impingement will occur as one component of the flow geometry shown in fig. 2.

Here the horizontal material streams join into one vertical stream, which has given a zone of low density at the centre. At large defor-
Impingement flow in extrusion.

The parts were cracked in the "fold zone" besides the tensile stress cracking at the top.

Tensile stresses tend to extend the material (1), leading to rapid pore growth and finally to formation of cracks. The cracks close again towards the end of the forging stroke (fig. 3). This is in agreement with results by Bockstiegel and Björk (2).

For the geometry in fig. 2, however, the stress cracks were not
reasonably closed even for very high forces due to less favourable geometry.

Surface cracking is slightly reduced by presintering (3) but at the price of higher flow stress (4). Another cause of cracks is folding that may happen in the forging of undercut sections (2) as well as in extrusion (fig. 2), and the risks for strain cracking and folding therefore will set the rules for the geometry relation between preform and die.

Lastly it is worth noting that the surface zones of powder forged samples often get lower density than the centre material. There seem to be two reasons for this effect, absence of shear flow discussed above, and cooling that will rise the flow stress of the surface material.

**Experimental procedure**

Even after accurate die charging, e.g. by weighing, the problem of finding the best preform shape remains. For a model study allowing the relevant parameters to be easily varied, the forging of gear rods was chosen.

Preshaped preforms were made by machining presintered bars (16x10x100 mm) of Hoeganaes AHC 100.29 iron powder of 82.4% mean relative density. The specimens were heated to 900°C in hydrogen atmosphere for 6 min. before transfer to the forging die (200°C), which took between 1 and 2 s and caused a centre temperature drop of about 15°C (50°C at the surface was calculated). Forging was carried out in a screw press fitted with a closed die tool that was lubricated by a suspension of graphite in mineral oil.

The resulting structures were studied by Brinell hardness testing of sections taken 1 mm from the lateral surface of the specimens. Mirror symmetry was utilized in the derivation of hardness maps from the indentation grids (typically 150 indentations a sample).

Brinell Hardness vs. relative density is given earlier (1). Brinell
hardness was chosen to represent the material rather than quantitative metallography. It is work saving and has a depth-integrating effect. As a matter of fact, even quantitative metallography will not give a relevant representation of the structure as there is an effect not only of the relative porosity but even of pore size, shape etc. To cancel out residual effects of work hardening, the specimens were annealed at 550°C for 1/2 h before testing.

Due to a large height to width ratio (1.6:1) of the green compacts from which the preforms were machined there was a slight variation in density corresponding to a range of 39 to 46.3 Brinell hardness units (80.9-84.6% density). The results are corrected for this variation.

Parameters investigated

Experiments were made with five groups of preshaped preforms designed to give fully dense forgings of equal form. The functional dimensions defined in fig. 4 are tabulated below.

Fig. 4 Functional dimensions of preform (a) and fully dense part (b). $L_B = l_B$ and $L_T = l_T$.

The deformation is mainly plane, but the clearance between die and work to allow the introduction of the hot preform caused a thickness increase of 5±1/2% ($T > t$). Dimensioning of the preform by scaling-up the final shape in proportion to the densities, considering width flow, would give:
\[
\frac{h_B}{H_B} = \frac{h_T}{H_T} = \frac{h_o}{H_o} = \frac{\rho_M \cdot T}{\rho_p \cdot t}
\]

the middle equality can be rewritten:

\[
\frac{h_T}{h_o} = 1
\]

<table>
<thead>
<tr>
<th>Group</th>
<th>Preform dimensions</th>
<th>Fully dense dimensions</th>
<th>Parameter</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>( h_o )</td>
<td>( h_T )</td>
<td>( h_B )</td>
</tr>
<tr>
<td>A</td>
<td>10.00</td>
<td>12.00</td>
<td>4.00</td>
</tr>
<tr>
<td>B</td>
<td>10.00</td>
<td>10.00</td>
<td>5.00</td>
</tr>
<tr>
<td>C</td>
<td>10.00</td>
<td>8.00</td>
<td>6.00</td>
</tr>
<tr>
<td>D</td>
<td>10.00</td>
<td>6.00</td>
<td>7.00</td>
</tr>
<tr>
<td>E</td>
<td>10.00</td>
<td>5.00</td>
<td>7.50</td>
</tr>
</tbody>
</table>

Dimensions in mm:s of preforms

The scaled-up preform will first be met at the top of the tooth by a rigid punch and material will flow to the base region before die contact is reached (fig. 5a).

\[ a: \frac{h_T}{h_B} = \frac{H_T}{H_B} \]
\[ b: h_T < H_T \]

Fig. 5 Initial contact situations at different preform height ratios (fully dense part shadowed).
With a preform tooth lower than that of the final part material is raised into the tooth from the base (fig. 5b). Confirmed by the study, the height ratio giving the most homogenous densification will be found within these limits:

\[ \frac{h_t}{h_B} < \frac{H_t}{H_B} < \frac{H_t}{H_B} \]  \[3\]

Specimens from five preform groups were forged to different stages of deformation to reveal the development of zones of higher and lower density.

To compensate the density variation of preforms, densification is represented by local hardness increment relative to corresponding sites of the sintered preform. A summary view over the forgings with rough values of the mean density is given by fig. 6.

High preform tooth

Analysing the results from group A with the largest \( \frac{h_t}{h_B} \) ratio, it is evident that the teeth will densify prior to the base and that local folding will appear at the base of teeth. This fold will later move to the centre of the base area and there meet another fold from the adjacent gear. Prior to full density the structure therefore will get a local weakness in consequence of impingement flow.

Figs. 7b-e show contours of equal relative hardness increment viz. 0-25, 25-50, 50-75 and 75-100% of the span between minimum and maximum hardness increment of the specific sample. (For a minimum and maximum hardness increase of 10 and 20 Brinell units respectively, the divisions would be 10-12.5, 12.5-15, 15-17.5 and 17.5 to 20 units). Fig. 7 and corresponding figures are aimed to show the localization of greater and smaller density change. The total density levels and variations are better studied in figure 6.

Figs. 7b and c show as expected the initial densification at the tooth top as well as a local densification zone at the tooth bottom corner due to the geometrically induced stress concentration. Later on
Brinell hardness increment over preform, units:  

- 7-17  
- 17-27  
- 27-37  
- 37-47  

- Preform  
- Fully dense forging  

Fig. 6 The progress of densification expressed by iso-Brinell hardness increment contours and rough values of actual percentage mean density.
Fig. 7 Forgings of the "A"-group. Folds marked by arrows.

This local zone expands until the increased flow stress of the densified material will either sustain or balance the applied stress. Finally these zones will spread and dissolve the density variation. The locus of main densification will move downwards from the tooth top during the stroke due to earlier densification and increasing area of die contact.

In B-group of preforms the densification pattern is similar. So will, for instance, the locus of main densification move from the top of the tooth to its core during the stroke. The advantage of the decreased height ratio is absence of extrusion flow along the bottom of the profile and less serious folding. Compare for instance the fig. 8e with the fig. 7e.
Fig. 8 Forgings from the "B"-group.

Fig. 9. Microstructures (75x) from a sample (8d) of the "B"-group.
**Low preform tooth**

For the preforms of lowest tooth-to-base height ratio, the "E"-group, a different flow pattern emerges. The initial indentation at the base results in a clear densification pattern along the directions of maximum shear flow (fig. 10b).

![Fig. 10 Forgings from the "E"-group.](image)

Continued deformation will extend the densified zone in front of the indenters and also expand it along the tooth flank due to an expanded area of contact and by forcing of material from the base into the tooth volume (fig. 10c).

Further deformation gives high density in the base and lower density within and below the tooth. Though the geometry in principle is an extrusion no surface cracks were found.

**Moderate preform tooth**

The groups C and D representing the upper (roughly) and lower limits of the relation $|3|$ gave less density variations than groups B and E respectively, although they confirm the main principles earlier discussed (figs 6 and 11) and the most uniform densification would therefore be found for a design between types C and D as predicted. The folding in geometry C seems to be harmless.
Discussions

The results are well according to known behaviour of porous metals. Of certain interest is the importance of slip and porosity variation. For "E"-preforms (fig. 16) the initial contact results in a higher stress in the \( \alpha \)- than in the \( \beta \)-direction, locating the primary densification to the \( \alpha \)-directed region (fig. 12a). Cf. fig. 10b.

![Fig. 11 Forgings of roughly the same density from the C and D groups (figs. 7n and s).](image)

![Fig. 12 Direction of densification and material transport in the forging of "E"-preforms.](image)

Later on, the contact will expand along the gear flank and result in a surface densification. Material forced into the tooth will now be pre-densified and cause a situation (fig. 12b) corresponding to fig. 10c. The densification results in a raised flow stress and a reduction of the remaining deformation to be done in the \( \alpha \)-direction. In coincidence with a raising force due to extended die contact, this increases the densification in the \( \beta \)-direction (fig. 12c). The resulting structure
will be dense in the \( \beta \)-direction and have "dead zones" at the teeth top and base and in front of the indenter (fig. 10e).

Also for eccentric die entering an effect of shear was found (fig. 13).

![eccentricity](image)

Fig. 13 Hardness pattern for a slightly forged B-sample in eccentric (4% of teeth interval) die entering. (Maximum hardness increase 7.4 units).

Conclusions
The risks for cracking and folding and the localization of remaining porosity delineate rules for preform design.

Despite softly rounded die engraving, forging of wave profiles from prismatic preforms led to stress cracking. Due to favourable die geometry the cracks were closed. Though apparently faultless, assurance must be found that they have got an acceptable strength (e.g. in fatigue) before closed cracks should be accepted. For another case, crack closure was far more difficult.

Experiments with preshaped preforms indicated damageous folding for large tooth/base height ratio (geometry A), a tendency that would diminish (B and C) and finally disappear when the actual ratio decreased.

The structure development was strongly affected by the preform height ratio with principally different patterns for initial die-work contact at the top or bottom of the preform teeth. The latter (E) will be less favourable due to a great portion of porosity along the centre-plane of the tooth that has to be eliminated during the last stage of
forging when the predensified base has become rigid.

Geometrically induced concentrations of compressive stress as well as shear flow under hydrostatic pressure were found to be useful instruments for density localization. Wall effects due to friction or mirror symmetry, however, would reduce the densification rate.

The most uniform densification was found for a moderate preform height ratio |3|, but an even more uniform distribution of residual pores might be expected for a design where also the horizontal geometry is taken into account.

Moderate preform shapes (C and D) might also cause the least die wear while other geometries might be tool destroying by the strong flow of material past the corners of the tool profile.

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F E M - SIMULATION OF THE DENSIFICATION PROCESS
IN THE POWDER FORGING OF A GEAR PROFILE

B. Arén

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FEM-SIMULATION OF THE DENSIFICATION PROCESS IN THE POWDER FORGING OF A GEAR PROFILE

Björn G A Arén

REPORT

LiH-IKP-R-40

1974-10-23
FEM-simulation of the densification process in the powder forging of a gear profile

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Preface

This work is a first approach to simulate the densification process in powder forging by use of the method of finite element analysis. This is without a suitable computer program available for plastic flow analysis, executed by an iterative process using the following schedule:

elastically induced stress \rightarrow local flow \rightarrow local densification and associated rise of local stiffness \rightarrow input of new data of material stiffness and changed geometry \rightarrow new elastic calculation etc.

Nevertheless in spite of this somewhat special method the results generally seem to be in agreement with the results of comparative forging tests. However, the discrepancies can readily be explained and compensated for.

For the calculations a computer-program "FEMFAB II" has been used. This program, designed for elastic analysis of two-dimensional problems was developed by tech. lic. Kenneth Axelsson and tech. lic. Mats Fröjer at the Dept of applied statics at CTH. Their permission to use the program is gratefully acknowledged.

In order to simplify the use of the program in this application, the subroutine for stress calculations has been extended at certain points (Appendix 1). This work has been done by tech. lic. Gunnar Andersson, LiH, who has also contributed valuable criticism during the work.

MANY THANKS!

Västra Harg 21 sept 1974

Björn Arén
FEM-SIMULATION OF THE DENSIFICATION PROCESS IN THE POWDER FORGING OF A GEAR PROFILE

Introduction

Powder forging has been a topic of great interest in recent years for the production of structural parts. The procedure is briefly as follows: compacted preforms, so called green bodies, of around 80% relative density are heated to forging temperature and are then formed and densified by plastic working. This is preferably done by forging in one stroke, with the aim of close tolerances. At the same time the material forged will reach full density in order to obtain maximum strength.

However, the porosity of green bodies leads to some problems, firstly risk of oxidation, which in practice requires that heating of the preform must be undertaken within a protective atmosphere, and secondly that the porous body requires certain deformation conditions.

Another important factor is poor knowledge of the rules controlling the densification process during plastic deformation of porous materials. One difficulty in the calculations is that plastic deformation takes place with a volume change, which makes the number of equations disproportionately large for analytical studies of the deformation.

However a number of empirical deformation laws for porous materials have been presented, and the aim of this study was to find out the extent to which these laws can be used in a FEM-analytical simulation of a powder forging process.

Mechanical properties of porous materials

Although a number of reports concerning different empirical and semi-empirical deformation laws for porous materials
have recently been presented, complete investigations of specific materials are still not available. Therefore the mechanical behaviour during different deformation processes must be extracted from different sources, which causes discrepancies between different input data used.

Although the mechanical behaviour of porous materials is strongly influenced by pore size, shape and distribution, as well as the state of contact that is established between different powder particles during compaction and presintering, it can normally simply be related to the porosity or relative density. The influence of other factors is normally accounted for by the introduction of different correction coefficients.

Deformation parameters to be used in this work are: stiffness (Young's modulus, \(E\)), rate of contraction (Poisson's ratio, \(\nu\)), yield strength (\(\sigma_s\)), and a criterion for plastic flow.

To quantify material stiffness, the resistance to elastic deformation expressed by Young's modulus has been used. The theoretical relationship due to Mc Adam (1) has previously shown good agreement with test results for both green compacts and sintered bodies of different densities. The Mc Adam formula in a rewritten form is as follows:

\[
E = E_M \cdot \left(\frac{\rho}{\rho_M}\right)^{3.4} \tag{1}
\]

where:

- \(E\) = Young's modulus of the porous material
- \(E_M\) = Young's modulus of the matrix
- \(\rho\) = density of the porous material
- \(\rho_M\) = density of the matrix

Since the pores are compressible, plastic deformation is associated with a volume change, causing the rate of contraction during plastic flow (\(\nu_p\)) to differ from the
value (0.5) for deformation without a volume change. With increasing porosity, this difference will increase ($v_p < 0.5$), as illustrated by figure 1.

Actual $v_p$-values are calculated for different degrees of porosity by interpolation of the initial $v_p$-rates (at $\varepsilon = 0$) for different density curves.

![Figure 1. Horizontal true strain as function of vertical true strain in pure upsetting of prismatic preforms of iron with different green densities during hot forging at 1160°C (2).](image)

Kuhn and Downey (3) have presented a formula \[ v_p = 0.5 \cdot \left( \frac{\rho}{\rho_M} \right)^2 \] for Poisson's ratio during plastic flow as a function of the relative density of the material:

This formula differs from our own in that the relative density is raised to the second power, while our own study (2) indicates that it should be raised to the third power. Although Kuhn and Downey have not studied powder forging of precompacted bodies of iron, but confined their study to deformation of sintered aluminium bodies, their experimental conditions would be expected to give reliable results. In the following analysis, $v_p$ will therefore be calculated using their formula 2].
In an earlier paper Kuhn and Downey presented a transformation of the Levy-Mises criterion for plastic flow in order to make it applicable to porous materials also (4). The comparative stress, \( f \), would in the case of porous material be:

\[
f = \left[ 3 J_2' - (1 - 2v_p) J_2 \right]^{1/2}
\]

where:

\[
J_2' = \frac{1}{6} \left[ (\sigma_1 - \sigma_2)^2 + (\sigma_2 - \sigma_3)^2 + (\sigma_3 - \sigma_1)^2 \right]
\]

\[
J_2 = - (\sigma_1 \sigma_2 + \sigma_2 \sigma_3 + \sigma_3 \sigma_1)
\]

\( \sigma_1, \sigma_2 \) and \( \sigma_3 \) are principal stresses

\( v_p \) = Poisson's ratio for plastic flow of the porous material.

To enable this flow criterion to be applied to a porous body with locally varying density, a knowledge of the local variations in yield stress is required. The calculation of yield stress is accomplished by using the following formula, due to Ishimaru et al (5), which takes porosity into account. For uniaxial tension, this equation has the form:

\[
\sigma_S = \sigma_{SM} (1 - K \pi^{2/3})
\]

where:

\( \sigma_S \) = Yield strength of the porous material

\( \sigma_{SM} \) = Yield strength of the matrix

\( K \) = a constant depending on the powder geometry

\( \pi \) = relative porosity \( (\rho/\rho_M) = (1 - \pi) \).
There is good agreement between experimental results from different powder materials and the relationship given by Ishimaru. The value of the constant $K$ lies in the range $1.70 < K < 2.26$ for powders of varying grain size and geometries, with one exception, which is invalid in this investigation. For this work, a value of 2 has been assumed for $K$, which appears reasonable when the materials used are compared with those studied by Ishimaru.

Forging experiments

In a separate paper (6) experiments on powder forging of gear rods from porous preforms of different shapes have been presented.

These preforms were made by machining presintered green bodies formed from Hoeganaes AHC 100.29 iron powder, with a relative density of 82.4%. Prior to forging the preforms were heated to $900^\circ$C and were then struck in a closed die-tool preheated to $200^\circ$C, to different stages of deformation. The preform geometries used, all designed to give the same final fully dense shape, are shown in figure 2.

![Figure 2. Porous preforms of different shapes (A-E), all designed to give the same final shape (F). The latter shape is shown by shading the figures (A-E) to show the shape relationships.](image-url)
The densification of the forged samples was mapped by taking Brinell hardness measurements, and checked metallographically. The results of these traverses can be seen in figure 3.

Figure 3. The development of densification during forging of preforms of different shapes shown by Brinell hardness maps of local hardness levels. The total relative densities are also given.
For one of the preform geometries (series D) the preformed teeth had the same dimensions as the shape of the tool and therefore the shape of the final forging. This type of forging can therefore be principally described as material being forced from the base upwards into the cavities of the tool without shape change of the tooth geometry during densification (figure 4).

![Diagram](image)

Figure 4. Forging of porous preform (a) to finished fully dense gear rod (b) with the same tooth geometry.

Disregarding the fact that there is a slight lateral plastic strain ($\varepsilon < 5\%$) caused by the clearance necessary to introduce the hot preform into the closed die tool, the deformation can be thought of as principally two-dimensionnal.

The dimensions of the preform (small letters) and those of the fully dense forged gear rod (capital letters) resulting from the difference in densities are as follows:

\[
h_o = H_o \cdot \frac{\rho_M}{\rho} \cdot \frac{T}{t}
\]

where:

- $h_o = h_B + h_T/2$ = mean height of the preform
- $H_o = H_B + H_T/2$ = mean height of the compact part
\[
\rho_M / \rho = \text{the reciprocal value of the relative density of the preform}
\]
\[
T = \text{width of the compacted part}
\]
\[
t = \text{width of the preform}
\]

Actual dimensions are given in figure 5 and the associated text.

![Figure 5. Dimensions (mm:s) of porous preform ("Serial D") and compact part, a and b respectively:](image)

- \(h_T = H_T = 6.0, h_B = 7.0,\)
- \(H_B = 5.25, l_T = L_T = 4.0,\)
- \(l_B = L_B = 8.36, T = 10.45,\)
- \(t = 9.95\)

The densification process associated with forging of preforms of this shape (figure 6) is heavily influenced by the actual conditions of deformation. When the tooth shape is intact and the basal line moves parallel to the gear surface, the relative strain underneath the base of the teeth is essentially greater than the relative strain beneath the tops of the teeth. Because of the variation in degree of deformation, there is also a corresponding variation in densification of the parts forged.
As seen in figure 6, densification starts beneath the teeth bases and then slowly grows horizontally, which means that the tops of the teeth and the material at the base vertically below, will be densified in the latter stages of the process. The slight lateral strain during deformation mentioned earlier may have influenced the results but only to a very small extent, and therefore has been neglected.

**FEM-analysis**

Due to the complex behaviour of the material investigated, an analytical treatment of the plasticizing process is extremely complicated, and may even be impossible. Although it is possible to get a reasonably correct idea about densification during the forging stroke via empirical statements, it is interesting to complete them with some type of analysis. The finite element method was considered suitable for the calculations, since this can conveniently be executed by computer.

Since the geometry given above is constant during deformation it can easily be studied, the element grid in this case being simplified. In this examination the element grid was kept intact within the main part of the body under investigation, and only the two basal element rows were adjusted as shown in figure 7.
Figure 7. Modification of the FEM-grid during the process of compaction (a + b + c) by decreasing the height of the basal elements. The system used contains only one of the two symmetrical segments from which the gear rod is formed.

The continually changing material behaviour during compaction is taken into consideration by simulating the behaviour of the different elements. This has been achieved using an iterative method:

I A prescribed displacement is introduced, building up a stress field which locally exceeds the yield strength of the material.

II For those elements where the calculated stress is larger than the yield strength, plastic flow with associated densification is also assumed to occur.

III The hardened elements are then promoted to a higher level of material stiffness. The procedure I + II + III is repeated until full density of the body is achieved.
As suitable FEM-programs to solve plastic problems associated with a change in volume were not available (and may not exist), the method described has been solved practically using the following scheme:

1. A computer program, "FEM-FAB II" (7), designed for elastic analysis of the stress field in a prestrained body has been used for the calculation of the stress levels in the different elements. During these calculations the elements are considered to be fully elastic.

2. The material of the investigated body has been divided into one of 8 different levels of stiffness, each one representing a given density level, and each of which is associated with certain material properties. These properties are found from equations 11 - 14.

3. Once the flow stress predicted by the calculated values of elastic stress, has exceeded the yield strength of an element, the element is introduced into a higher level of stiffness in accordance with the actual calculated stress level.

4. This procedure is then repeated, and the geometry of the body is adjusted to compensate for the volume change caused by densification.

The calculated density distributions are given in figures 9 and 10.

The computer-program "FEMFAB II" is a program where elastic stresses and displacements of nodes and structural elements are calculated. The calculations are carried out from data concerning the geometrical constitution of the structure, including its different elastic properties and applied stresses introduced by prescribed displacements and forces. The program and its facilities are fully described by the authors (7).
In this paper two comparative configurations of elements are studied, both structures having 51 nodes and 74 elements (figure 8), but with different structure design.

Because of the mirror symmetry of the gear rod, the study can be limited to the section shown in the figure. The complete section can be determined by considering identical changes equidistant from the line of symmetry.

Symmetry also dictates that horizontal displacements of nodes on vertical symmetry lines must be equated to zero. Thus these structural nodes are only permitted to move vertically. The upper bounding line is also fixed because of its constant geometry.

Figure 8. The different structures studied. Node numbers are indicated by standing figures and element numbers by angled figures.
Deformation is assumed to occur between two fully rigid die-tool surfaces which move towards each other in the vertical direction. The applied load is represented by a uniform vertical displacement of nodes 45 - 51.

The basic FEMFEM-II-program calculates the induced stress under plane strain conditions in terms of horizontal and vertical mean stresses in different elements ($\sigma_x$ and $\sigma_y$), and the principal stresses ($\sigma_1$ and $\sigma_2$) in the plane and their principal directions ($\alpha$) in the chosen coordinate-system.

In order to use Kuhn and Downey's flow criterion $|3|$, the third principal stress ($\sigma_3$), must be known.

In the case of biaxial strain this is given by:

$$\sigma_3 = \nu (\sigma_1 + \sigma_2) \quad |5|$$

Since Poisson's ratio ($\nu$) is used individually for each element, the actual value of plastic strain ($\nu_p$) is used. This approach is discussed later.

The normal subroutine in the computer program for stress calculations is extended with subroutines to calculate the third principal stress $|5|$ and the relevant comparative stress $|3|$ of the different structural elements. Furthermore, another subroutine has been introduced which executes stress comparisons, and the results from these are also used to rank the material of different structural elements into the correct stiffness level. This subroutine also prints the results.

Also contained in the subroutine for stress calculations is a section which calculates the porous volume of material lost due to the densification process. This decrease in macroscopic volume is then expressed as a pure vertical movement of the base line.

The subroutine modified in this manner is presented in Appendix 1.
In the FEM-study the preform geometry has been simplified in some cases when compared with the geometry of the forged samples.

I The width of each structural element is chosen as 1.00 mm, which changes the dimensions $l_T$ and $l_B$ (figure 6) from 4.00 and 8.36 to 4.00 and 8.00 mm respectively.

II The broadening of the preform during the deformation is neglected, which changes the relation $|5|$. The heights $H_o$ and $H_B$ for the fully dense part are adjusted from 7.85 and 4.85 to 8.25 and 5.25 respectively.

III The relative density is raised in steps of 0.025 from 0.825 to full density (1.000). The different stiffness levels calculated by the formulae $|11| - |14|$ give the values given in the following table.

<table>
<thead>
<tr>
<th>Level of material stiffness $\rho/\rho_M$</th>
<th>Relative plastic Poisson's ratio $\nu_p$</th>
<th>Yield stress $\sigma_s$ N/mm$^2$</th>
<th>Young's modulus $E$ N/mm$^2$</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>0.825</td>
<td>0.341</td>
<td>40.57</td>
</tr>
<tr>
<td>2</td>
<td>0.850</td>
<td>0.361</td>
<td>46.38</td>
</tr>
<tr>
<td>3</td>
<td>0.875</td>
<td>0.383</td>
<td>52.50</td>
</tr>
<tr>
<td>4</td>
<td>0.900</td>
<td>0.405</td>
<td>59.06</td>
</tr>
<tr>
<td>5</td>
<td>0.925</td>
<td>0.427</td>
<td>66.18</td>
</tr>
<tr>
<td>6</td>
<td>0.950</td>
<td>0.451</td>
<td>74.20</td>
</tr>
<tr>
<td>7</td>
<td>0.975</td>
<td>0.475</td>
<td>83.77</td>
</tr>
<tr>
<td>8</td>
<td>1.000</td>
<td>0.490</td>
<td>100.00</td>
</tr>
</tbody>
</table>

Table: The 8 levels of stiffness used in the FEM simulation of the densification process. The usual value 0.5 for the plastic Poisson's ratio of solid material is replaces by 0.490 in order to avoid instability phenomena which would affect the results of the computer calculations.
The yield strength for compact materials is assumed to be 100 N/mm² (Hot deformation). It might be argued however that only the variation of flow stress with density would affect the densification process, thus the true value of the yield strength is relatively unimportant.

The computer calculations have been initiated by a slight displacement (0.0013 mm) of the lower limit, increasing in two steps to 0.0019 mm in order to raise the stress level and thus expand the densified zone due to plastic flow. Because of the raised hardness however, the stress increase will be self generating, thus the prescribed displacement must be reduced in order to keep the step-wise densification at a reasonable level. During the final stage however, the displacement must be increased again in order to induce the stress needed for total plastification.
Figure 9. FEM-simulated process of densification for structure type a. Structural elements with different stiffnesses are shown with a different level of shading. (Level 1 pure white, = level 8 very dark).
Results from repeated calculations are given in figures 9 and 10.

In the calculations, elements are not promoted to a higher density level until the stress level has reached the yield strength of the higher level. Thus an element will belong to level 1 until its actual stress has passed the value 46.38 N/mm$^2$ (see table).

Figure 10. The densification process for structure type b.
In contrast, no relegation of an element to a lower stiffness level has been made as such an occurrence is unlikely in the practical forging process. Tensile stresses should apparently be able to reduce the density, but in this case primarily by cracking and not in a homogeneous manner. Although no routine to check for tensile stresses of a critical value has been introduced into the program, such an approach would seem to be reasonable.

Discussion

The two structure types seem to give approximately the same calculated densification development, starting at the bottom corner of the gear teeth and ending at their tops. (Also in geometry b the structural element 2 will be the last to be densified, which is not clearly shown in figure 10). A comparison of figures 9d - 10c, 9e - 10d and 9g - 10e shows good agreement, indicating insensitivity to element structure design.

At later stages however, this insensitivity appears to diminish (9j - 10g and 9k - 10h), which maybe because of the limited number of elements and also their orientation. This could be explained by local effects of self-perpetuation caused by stiffness, and ought to be limited with an increasing number of elements.

Of greater importance are the disturbances that appear to be created by the interaction of adjacent structural elements of different stiffness level. E.g. figure 9g and other figures from the intermediate stage show a zone where elements of the highest stiffness levels alternate with elements of relatively low density.

To a certain extent it may be assumed reasonable that porous areas surrounded by stiffer material would be difficult to compact, but this is not a satisfactory explanation of the effect. One possible explanation could be that the model used for densification totally neglects the material flow between the elements that must take place in reality.
Because of this, the hardened element would sustain an escalating load leading to further densification.

The postulated model has not taken macroscopic material flow during compaction into consideration. The only consideration of plasticity between structural elements has been the use of a plastic Poisson's ratio $\nu_p$ instead of the corresponding elastic value $\nu$ ($\nu < \nu_p$) for the stress calculations.

It seems clear that the interaction across the element boundaries must be taken into consideration more effectively within a refined model in order to reduce the instability occurring in the simulated densification process between adjacent elements.

A comparison of compaction as a function of true strain in closed-die upsetting with open upsetting between flat dies (figures 11 and 12) clearly shows the reason why the material flow across the element boundaries should be taken into account.

![Graph](image)

**Figure 11.** Mean density in closed and open die upsetting together with "centre density" in open die upsetting as functions of true vertical strain. Expressions are illustrated in figure 12.
Figure 12. Closed die upsetting a, and open die upsetting b. "Centre density" represents the density of the section shadowed, c, of the open die upset body (=half the height and half the width in the centre).

It seems reasonable to assume that interference between two adjacent porous elements should cause densification due to macroscopic material flow. This is less extensive than that found in closed die upsetting (dashed line in figure 11) but of a higher degree than in the case of open die upsetting (dash-dotted line). In a refined analysis this effect should be taken into account.

Finally, the effect fixing all the nodes at the upper limit is discussed.

In the present version of the program only nodes 2 and 18 are allowed to move, while the more interesting nodes 1, 4, 8, 12, and 19 must be kept fixed. Because of this it appears wrong to give larger freedom to only two of the node points and so these were kept fixed also.

In a true forging process the coefficient of friction in hot forging with poor lubrication can be estimated to be 0.3 - 0.5 (9), and a check of the computer results shows that the forces of sticking friction on several occasions are exceeded. Whether fixing the node points has affected the results or not is difficult to determine, but in principle this is of course a weakness.
In conclusion, some of the results of this work seem to give reasonable results, although improvements with regard to material flow over element boundaries and also the mobility of element nodes along the upper limit would be valuable.
References

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4. B. Arén  

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6. Y. Ishimaru, Y. Saito och Y. Nishino  
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7. K. Axelsson och M. Fröier. "FEMFAB II"  
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SUBROUTINE MT32SE(A, V, SM, TEMPV, MEP, NOD, NRELM, NH, NHTE, 
1NLKV, IP, PNYV12, XOR, YNOD)

C
C
C
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C

DIMENSION A(1), V(1), NOD(1), TEMPV(1), MEP(1), NHTE(1), EM(24, 1), 
1 SM(12, 1), SIGI(3), PNYV12(1), XOR(1), YNOD(1)
DIMENSION GRX(3), GRY(3)
DIMENSION SIGS(8), X(3), Y(3), N(100), M(100)
SIGS(1) = 40.57
SIGS(2) = 46.38
SIGS(3) = 52.50
SIGS(4) = 59.06
SIGS(5) = 66.18
SIGS(6) = 74.20
SIGS(7) = 83.77
SIGS(8) = 100.0
R1 = 0.
R2 = 0.
MN = 3
NF = 2
MRS = 3
MF = MN*NF
WRITE (IP, 800)
WRITE (IP, 810)
WRITE (IP, 830)
DO 90 NO = 1, NRELM
NUNIV5 = MEP(NO)
TEMP = TEMPV(NUNIV5)
WRITE (IP, 815)
READ (11) ((SM(M, N), M = 1, MRS), N = 1, MF)
IF (TEMP.GT.0.001), READ (11) (SIGI(JA), JA = 1, 3)
DO 35 L = 1, NH
DO 35 J = 1, MN
DO 35 N = NO*(NO-1)+J
DO 9 K = 1, NF
LEN = NEKV*(L-1) + NF*(NN1-1) + K
KOL = NF*(J-1) + K
EM(KOL,L) = V(LEN)

C * EM IS USED FOR STORAGE OF ELEMENT NODAL DISPLACEMENTS.
C ONE COLUMN FOR EACH LOAD CASE *
CONTINUE

INR = 2*MRS*(L-1)
DO 10 JA =1,MRS
10 IJL = INR + JA
A(IJL) = 0.

C * A IS USED FOR STORAGE OF STRESS VECTOR *
DO 10 KN = 1, MF
A(IJL) = A(IJL) + SM(JA,KN)*EM(KN,L)
10 CONTINUE

IF ((NHT(L).EQ.0).OR.(TEMP.LE.0.001)) GO TO 20

A(INR+JI) = A(INR+JI) + SIGI(JI)

SIGX = A(INR+1)
SIGY = A(INR+2)
TAUXY = A(INR+3)

CALL PRIN (SIGX,SIGY,TAUXY,SIG1,SIG2,PANG,IP)

A(INR+4) = SIG1
A(INR+5) = SIG2
A(INR+6) = PANG
PNY=PLYV12(NUNIV5)
A(9)=PNY*(A(4)+A(5))
P22=A(4)*A(5)+A(5)*A(6)+A(6)*A(4)
A(1)=SQRT(1.*P12+(1.-PNY)*P22)
A(2)=SIGS(NUNIV5)
A(3)=SQRT(2.*PNY)
NUNIV6=NUNIV5

IF (A(1)-A(2)) 27,21,21
21 IF (NUNIV6.EQ.8) GO TO 27
IF (A(1)-SIGS(8)) 23,22,22
22 NUNIV6=8
GO TO 27

i=7
IF (A(1)-SIGS(I)) 26,25,25
25 NUNIV6=1
GO TO 27
26 I=I-1
GO TO 24

27 CONTINUE
N(N0)=NUNIV6
DO 30 I=1,MN
K=NO-1)*MN+1
J=NOU(K)
GRX(I)=XNO(J)
Y(J)=YNO(J)
XU=(GRX(1)+GRX(2)+GRX(3))/3.
YU=(GRY(1)+GRY(2)+GRY(3))/3.
X(1)=GRX(1)-XO
X(2)=GRX(2)-XO
X(3)=GRX(3)-XO
Y(1)=GRY(1)-YO
Y(2)=GRY(2)-YO
Y(3)=GRY(3)-YO
I=(X(2)*Y(3)-X(3)*Y(2))*3.
AREA=ABS(I)*0.5
REDA=AREA*(1.-SQRT(PNYV12(1)/PNY))
KL=RI+REDA
PNY=PNYV12(NUNIV6)
REDA=AREA*(1.-SQRT(PNYV12(1)/PNY))
R2=R2+REDA
COMP1=RI/6.
COMP2=R2/6.
A(6)=SQR(T(2.*PNY)
A(7)=SIGS(NUNIV6)
35 CONTINUE
50 CONTINUE

GO TO 24
DO 130 NEA=2,8
  NRELA=0
100  DO 120 NO=1,NRELM
     I=0
     IF (N(NO)-NEA) I=120,110,120
110   NRELA=NRELA+1
     M(NRELA)=I
120   CONTINUE
     WRITE (IP,802) NEA,NRELA
     WRITE (IP,863) (M(I),I=1,NRELA)
130  CONTINUE
     WRITE (IP,800)
     WRITE (IP,800)
800  FORMAT (*//,'//26(*)
810  FORMAT (*//28X,'POWDER FORGING')
815  FORMAT ('')
830  FORMAT ('ELEMENT',2X,'SIGMA-1',4X,'SIGMA-2',8X,'F',8X,
     1 'PROP.NR',2X,'FLOW ST.',2X,'DENS. R.',4X,
     2 'NEW P.NR',2X,'NEW F. ST.',2X,'NEW D. R.')
840  FORMAT (1X,I5,3G11.5,I10,G12.5,G10.4,I10,G13.5,G11.4)
850  FORMAT ('BOUNDARY DISPLACEMENT AT START=',F8.6,' MM')
851  FORMAT ('NEW BOUNDARY DISPLACEMENT=',F8.6,' MM')
860  FORMAT (*//28X,'INPUT DATA FOR NEXT RUN')
861  FORMAT ('CARD GROUP 2')
862  FORMAT ('0',2I5)
863  FORMAT (1X,10I5)
147   RETURN
148   END
**INPUT DATA**

**GENERAL INFORMATION**

NUMBER OF NODES (NNDOD) = 51
NUMBER OF ELEMENTS (NRELM) = 74
NUMBER OF NODES PER ELEMENT (MN) = 3
NUMBER OF FREEDOMS PER NODE (NF) = 2

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NUMBER OF NODES WITH FIRST FREEDOM PRESCRIBED TO ZERO = 12

NUMBER OF NODES WITH SECOND FREEDOM PRESCRIBED TO ZERO = 0

NUMBER OF NODES WITH THIRD FREEDOM PRESCRIBED TO ZERO = 0

NUMBER OF FREEDOMS PRESCRIBED IN OTHER WAYS = 7

NODES WITH ALL FREEDOMS PRESCRIBED TO ZERO

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NODES WITH FIRST FREEDOM PRESCRIBED TO ZERO

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45 51

FREEDOMS NOT PRESCRIBED TO ZERO

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**NODAL LOADS**

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NO VOLUME FORCES

NO TEMPERATURE LOAD

**NECESSARY NUMBER OF ELEMENTS IN A = 1445**
RESULTS

LOAD CASE NO = 1

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OPTIMIZING THE PREFORM SHAPE IN POWDER FORGING
OF A LINEAR GEAR PROFILE

B. Arèn

Published in
Powder Metallurgy Int. Vol 7, No 1, 1975
Optimizing the Preform Shape in Powder Forging of a Linear Gear Profile

B. G. A. ÅREN

Abstract
Powder preforms of systematically varied shapes were forged to a gear-profile form.
Thence, the accuracy of tooth height versus total height of the preforms on the localization of remaining porosity in the forgings, the risk of cracking and the tendency for folding were investigated. Densification during different stages of forging up to 94% relative density was studied by Brinell-hardness testing. A geometry relation is suggested for the optimum shape of preform versus shape of the final part, with respect to uniform distribution of residual porosity.

Optimierung der Vormformgeometrie beim Pulverschmieden eines Zahnnastengprofils

Pulverpreformen mit systematisch geänderten Geometrien wurden zu einem Zahnprofil geschmiedet. Der Einfluss der Zahnhöhe in bezug auf Gesamthöhe der Vorform auf die Lokalisation der Restporosität in den Schmiedestücken, die Möglichkeit der Ribbildung und die Tendenz zur Überlappung wurden untersucht. Die Verdichtung während verschiedener Stadien des Schmiedens bis zu 94% Dichte wurde durch Brinellhärtemessung untersucht. Eine Geometrieevaluation der préformes pour le forgeage par poudres des profils d'engrenages linéaires

Une étude systématique a porté sur des préformes de poudre diverses utilisées en vue du forgeage des profils d'engrenages. On a étudié l'influence du rapport hauteur de dent/hauterelatif de l'épaisseur de la poudre grâce à des essais de pile de garniture. Le degré de densification pour différentes étapes du forgeage jusqu'à 94% relative density was studied by Brinell-hardness testing. A geometry relation is suggested for the optimum shape of preform versus shape of the final part, with respect to uniform distribution of residual porosity.

Introduction
Densification in powder forging is a process of great complexity. Earlier work on this topic as well as the closed-die forging indicates non-uniform material flow during the deformation. This is a consequence of the nonuniformity of the stress-strain field within the workpiece during the stroke, due to deformation geometry as well as to friction between die and work. Other causes of nonuniform flow could be porosity and temperature fluctuations in the preform.

Two main philosophies of powder forging are emerging. One of these is that of porosity-free material in order to reach the full strength of the base metal; in principle this can always be achieved when the forging pressure is high enough. The other philosophy aims at densities around 97% which permit strength levels very near those of compact material, though with reduced ductility and impact strength. In compensation, lower forging pressure and reduced tool wear are expected. The main merit of this approach, however, lies in the possibility to achieve closer tolerances, i.e. because the residual porosity acts as a "buffer" to variations in preform weight.

The purpose of this work is to study in detail the non-uniformity of densification during the early stages of forging for different series of systematically varied geometries. The main part of the investigation is restricted to the early stages (up to 94% mean density), owing to limitations of the equipment used.

Because of this restriction, the present results will be relevant mainly to that variant of powder forging which aims at close tolerances and allows a certain amount of residual porosity. The possibility of localizing this residual porosity, either to uncritical regions of the sample or into some desired distribution, by advantageous interplay between the preform shape and that of the final part, would be of great importance for this type of powder forging.

Insofar, however, as nonuniform initial flow also sets the conditions for the latter occurrence of defects like folding and cracks, the results should be relevant to the optimizing of preform design for all types of powder forging.

Some Aspects of the Deformation Behaviour of Porous Materials
As a result of the compressibility of the pores, the plastic deformation of a porous material will be accompanied by a change in volume in such a manner that tensile stresses tend to expand, and compressive stresses decrease the volume. The exact densification path during an actual forging operation will therefore be difficult to predict, although basic flow criteria for homogeneously porous materials have been presented during the last few years.

In an earlier work on free upsetting of porous iron compacts under plane strain conditions it was shown that there is a difference between the vertical and the horizontal true strains, owing to the densification which accompanies deformation. This can be expressed by the decrease of Poisson's ratio for plastic flow, \( \nu_p \), with the porosity. For hot-compaction of non-presintered iron powder compacts (where frictional effects were not eliminated), the Poisson's ratio was found to decrease from 0.35 to

0.13 when the relative densities \( (\rho_v = \rho_q) \) varied from 0.802 to 0.626 (For compact material \( \gamma_p = 0.5 \)).

Another important aspect of the densification process concerns the propagation of stresses and uniaxial flow. Though only the hydrostatic component of strain will reduce the volume of pores, deformation will create a pore geometry that facilitates later pore closure. The lack of a shear component in some particular deformation geometries may have important consequences, for instance when the material flow comes to a stop against a wall, giving "dead" zones of high porosity. As has been shown earlier, the extension of the porous material by tensile stresses can lead to pore coalescence and finally to the formation of cracks. With proper forging geometry, such cracks will be repaired before the end of the stroke. This is illustrated in Fig. 1, which deals with the forging of a wave profile.

The tendency towards surface cracking can be diminished by presintering but at the price of higher initial deformation resistance.

Another reason for the occurrence of surface cracks is the folding of surface material. Such folding may happen, e.g. in the forging of undercut sections.

In the design of powder forgings the risks for strain cracking and folding will outline a corridor for the geometry relations between preform and die that are suitable for forging. For simple preform shapes, this has been treated in detail by Kuhn and Downey.

Lastly it is worth noting that the surface zones of powder forged samples often show lower density than the bulk. There seem to be two main reasons for this: absence of shear flow, discussed above, and surface cooling. Cooling will rise the flow stress of the surface material so that a higher porosity remains despite the fact that the material transmits the flow stress required to densify the interior zones.

Prismatic Preform
From the preceding discussion on flow behaviour, it is clear that even with accurate preform weight the problem of finding the best shape of the preform remains.

Initially the forging of a wave-profile from a plain preform (Fig. 3) of 80% relative density was studied. These preforms, of dimensions \( 10 \times 10 \times 100 \) mm, were made by compacting 62.5 grams of Hogeganas HC 100.24 iron powder without lubricant admixture with an accuracy of ± 0.1% in weight and ± 0.5% in height. (The length and width were given by the tool.) The bars were heated to 950°C (1225°F) by introducing them into the hot zone of a tube furnace with hydrogen atmosphere, and holding them there for about 6 min. The transfer of the specimen from the furnace to the die took between 2 and 3 s, during which the centremperature dropped by ca. 30°C as recorded by a thermocouple embedded in a typical test bar. Calculations reported separately would imply a temperature drop of ca. 70°C at the surface.

Forging was carried out in a toggle press fitted with a closed-die tool, and instrumented for recording force vs. deformation during the stroke. A suspension of graphite in mineral oil was used for lubrication, as in conventional die forging. The die was not preheated during these preliminary experiments.

As expected, the flow pattern during deformation was rather complicated, due to changes in the area of contact with the die as well as to previous local deformation.

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Brinell hardness was preferred for characterization of the material over quantitative metallography because it is less work consuming and has a depth-integrating effect. As a matter of fact, even quantitative metallographic measurements of relative porosity will not be directly representative of the final properties, because of the effects of pore shape, size and distribution.

In order to cancel out residual effects of work hardening, the specimens were annealed at 550°C for 1/2 h before hardness testing.

The hardness distribution for four consecutive stages in the forging of a workpiece to almost full final density (98 to 99 % relative mean density) is shown by Fig. 3.

As expected, the flow pattern during the deformation was complicated owing to the continuously changing area of contact with the die as well as the hardening by previous deformation. As shown in the figure, the densification starts at the points of indentation (a), whereafter it widens (b) until the zones impinge (c) and finally merge together (d).

As mentioned earlier (Fig. 1) tensile stresses will open cracks near the surface. This process is shown by the white zones in Fig. 3c, indicating decreased hardness, immediately prior to cracking. In the present geometry the cracks close again at a later stage. Traces of these cracks can be seen as a pattern on the surface of the forged samples, although no remaining internal porosity was found by metallography, cf. Fig. 4.

To summarize the results with plain bar preforms: Densification progresses quickly in the vertical direction, but only slowly in the lateral direction, producing a strongly non-uniform densification pattern, which results in crack formation in zones exposed to tensile stresses. The cracks close again towards the end of the forging stroke where die filling is completed. This is in agreement with results presented by Bockstiegel and Björk and by Sjöberg.

**Preshaped Preforms**

Even if the forgings from plain prismatic preforms were apparently free from remaining flaws after crack closure, it was evident that preshaped preforms would be required for the forging of more complicated shapes.

A gear rod with trapezoid tooth profile (Fig. 5) was selected for a model study as this geometry would allow the relevant parameters to be varied in a simple manner. In this case the study was restricted to the initial stages of densification.

This time the preforms were machined from presintered (1 hour at 900°C in an atmosphere of cracked ammonia) bars (16 × 10 × 100 mm) of Hoeganaes AHC 100.29 iron powder with a mean relative density of 0.973. The bars were preheated to 600°C (1173°K) in a tube furnace with hydrogen atmosphere with a holding time of 6 min. The temperature difference between the furnace and the centre of the sample was then about 5°C. Transfer of the specimens from the furnace to the forging die, preheated to 200°C (473°K), took between 1 and 2 s causing a centre temperature drop of ca. 15°C (corresponding to a calculated drop of 50°C at the surface). Forging was carried out as in the preceding section.

In comparison to the previous forging of wave-profiles there were two restrictions. The first one is that the force was limited by the construction of the tool, restricting the final mean density to a level around 95 %. The second restriction was a slight variation in the preform density after cold compaction, owing to the large height-to-width ratio (1:6:1) of the green compacts from which the preforms were machined. This variation of the density was the same in all specimens, corresponding to a range of 39 to 46.3 Brinell hardness units (80.9 to 84.6 % density). The results presented in Fig. 9 and following figures are corrected for this systematic variation.

1. Parameters Investigated

Five groups of experiments were made with different preshaped preforms (Fig. 6), designed to subject the teeth and the base to varying amounts of compression.

The dimensions of preforms and fully dense parts are tabulated below. The notations are defined in Fig. 7.

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**Symbols**

- < 50 HB
- 50 < 60
- 60 < 70
- 70 < 80
- 80 < 90
- 90 < 100
- 100 < 110
- 110 < 120
- 120 <

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**Fig. 1** Successive stages during forging of a wave profile from a plain prismatic powder preform showing the formation and re-sealing of cracks in tensile stressed surface zones.

**Fig. 2** Brinell hardness vs. density in uniform portions of forged specimen. Impressions made with 2.5 mm dia indenter at loads of 31.25 or 15.625 kp.

**Fig. 3** Successive stages during forming of a wave profile from a plain bar-shaped preform. The porosity distribution is indicated by Brinell hardness impressions on fully annealed material, cf. Fig. 2.
Crack traces not oxidized

Fig. 4 Crack traces remaining after closure (a) and sound microstructure of the previously cracked zone (b) (50x). (cf. Fig. 1)

Fig. 5 Gear rod made by powder forging. Total length ca. 100 mm.

All preforms were designed to give equal dimensions at full density and therefore they all contained the same amount of material in correspondence to a constant mean height, $h_o$ ($h_o = h_T + h_B/2$).

Table 1 Dimensions of different groups of preforms versus dimensions of fully dense shapes.

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The deformation is principally in plain strain, but a small width clearance between the work and the die, which was necessary to allow the introduction of the hot preform, caused a thickness increase of $5 \pm 1\%$ from preform (t) to forged detail (T).

For a start one might try to determine the preform dimensions from those of the final shape by a simple scaling-up in proportion to the densities, allowing for width flow. This would give:

$$\frac{h_T}{h_o} = \frac{H_T}{H_o} = \frac{h_B}{h_T}$$

(1)

The middle equality can be rewritten

$$\frac{h_T}{h_o} = \frac{H_T}{H_o} = 1$$

(2)

However, this will give nonuniform deformation unless one uses a tool with multiple punches. With a rigid punch, the scaled-up preform will first be struck at the top of the tooth, and material will flow to the base region before it meets the forming surface of the punch there (Fig. 8a).

The opposite type of material flow will be enforced when the preform

Fig. 6 Porous preforms (white) of different shapes (A to E), all designed to give fully dense details of equal dimensions (F). Shaded areas indicate fully dense profile at end of forging stroke.

Fig. 7 Essential dimensions of preform (a) and fully dense shape (b).

Fig. 8 Two types of initial contact between work and die (a and b) at different height ratios teeth/base.
tooth is lower than the finished tooth (Fig. 8b). Now material has to be raised into the tooth from the base. The optimum ratio of tooth-to-base heights will be found somewhere between the cases shown in Fig. 8a and 8b.

\[
\frac{h_T}{h_b} < \frac{h_r}{h_b} < \frac{H_T}{H_b} \tag{3}
\]

These limits are represented by the specimen groups D and C respectively. The present study has confirmed that the most homogenous densification is found within these limits.

A number of specimens from the preform groups listed in the table were forged to different stages of deformation in order to reveal the development of densification zones. Such local densification was mapped by Brinell hardness impressions on sections of the forged and annealed material as in the case of the wave profile.

To compensate for the initial density variation in the preforms which was mentioned above, the densification is represented by the local hardness increment relative to the corresponding site of the sintered preform. A summary view over the results is presented in Fig. 9, together with approximate values of the mean relative density \( \frac{\delta}{\delta_0} \) of the different forgings. These estimates were derived by planimetric profile measurements of longitudinal sections in combination with the known weights of the specimens and their actual thicknesses.

2. High Preform Tooth

Analyzing the results from the group with the largest \( h_T/h_b \) ratio (group A) it is evident that the teeth density prior to the base and that the flow from the teeth also causes a local folding at the concave corner at the base of teeth. This folded zone moves by macroscopic material flow towards the centre of the base area (Fig. 9d) until it meets the folded zone from the adjacent gear. The structure prior to full densification will therefore show a local region of weakness in consequence of the impingement of two material flows. Figures 10b–e show the local density variation by contours of hardness increment in quartiles of the total span between minimum and maximum hardness increment of the specific sample. (For a specimen with a minimum and maximum hardness increase of 10 and 20 Brinell units respectively, the quartiles would be 10–12.5, 12.5–15, 15–17.5 and 17.5 to 20 units.) It should be kept in mind that Fig. 10, and corresponding later figures, show neither the total hardness increase nor the absolute range of variation. They are merely intended to focus the attention to the localization of greater and smaller changes, i.e., regions of preferential densification. The absolute levels of density as well as the variation are better studied in Fig. 9.

Figure 10b and c clearly show the primary densification localized to the tooth top as expected.

There is also a local densification zone just below the concave corner at the tooth bottom (see arrow). This is due to a local plastic flow caused by the geometrically induced stress concentration. Later on, this densified zone expands until the increased flow stress of the densified material will locally sustain the applied stress. In this way the zones of high densification spread and dissolve the density concentration. The zone of most intense densification, initially at the tooth top, will move downwards along the stroke (Fig. 10b–d) since the flow stress of material that has already been densified is increased. At the same time the area of contact at the tooth flanks increases, changing the stress distribution in the tooth.

Another tendency to be observed here is a concentration of densification to regions of large shear flow in connection with a hydrostatic pressure. The densification pattern in the base below the tooth in Fig. 9d and 10c coincides with the known “X”-pattern of maximum shear flow. Another, and even cleaner example of this phenomenon will be found in Fig. 12.

The next group of preforms, group B, also has a high tooth-to-base height ratio (\( h_T/h_b \)) but less than group A. Its densification pattern is partly similar to that of group A. For instance, the locus of strongest densification again is found to move from the top of the tooth towards the core. The effect of stress concentration at the tooth base is clearly brought out by the hardness increment chart in Fig. 11c.

The main advantage of the decreased tooth height ratio, however, seems to be absence of extrusion flow along the bottom of the profile. Although folding must be expected at a later stage, its effect will be less serious. Compare, e.g., Fig. 11e and 10c.

3. Low Preform Tooth

Turning now to the preforms of lowest tooth-to-base height ratio (\( h_T/h_b \)), the “E”-group, a different densification pattern emerges. This is shown by Fig. 12.

In this case the initial contact between preform and die occurs at the base area. Indentation of material between the teeth results in a clear "X"-pattern of preferred densification along the directions of maximum shear flow (Fig. 12b).

Continued deformation not only extends the densified zone in front of the primary indentations but also expands the densified region along the tooth flank by forcing of material from the base into the tooth volume (Fig. 12c).

Further deformation yields a region of well compacted material at the base, in combination with a less compacted region within and below the tooth. Again, densification is seen to spread much more slowly in the lateral than in the vertical direction, and the highest amount of porosity is found at the tooth top where the last contact with the die will occur. It is interesting to note that no surface cracks were formed although the geometry chosen, because of lack of initial restraint at the tooth surface, is principally a type of extrusion.

4. Moderate Preform Tooth

The specimen groups C and D representing the upper (roughly) and lower limits of the relation (3) showed on the whole less inhomogeneous densification (Fig. 9) than groups B and E respectively. However, the density pattern confirms the principles which emerged from the previous observations (Fig. 13). The most uniform densification would be expected for a preform design between types C and D, as predicted by relation (3).
The key to the observations presented in this paper is given by two basic phenomena: the behaviour of a porous material under local indentation, and the influence of shear flow on densification.

For the angular gear profile, the optimum of preform shape with respect to homogeneous deformation will be found in the interval defined by eq. (3), whose limits are represented by preform types C and D in Fig. 6. In such a preform both the tooth and the base are subjected to compression, but mild upsetting of the tooth precedes the compression of the base. This preform shape does not allow high tensile stresses to develop anywhere near the surface and thus avoids cracking. Neither does it produce objectionable folding at the corner between tooth flank and base (which is a preferred location for folding when a high tooth, as in preform type A, is excessively upset before it has been enclosed by the punch cavity). Preforms within the preferred interval also will produce little sliding of material under pressure against the die surface, and should therefore give low tool wear.

Acknowledgements
Helpful discussions with Prof. H. Fischmeister, Chalmers University Techn., are gratefully acknowledged.

References

Though the specimen in Fig. 13a exhibits a slight folding tendency, this does not seem to be serious. Both geometries C and D would probably be acceptable for production, although the final densification at the top is more sluggish in geometry D (cf. Fig. 9s).

Discussion
The dependence of flow stress on porosity in both cases follows the $\sigma = \sigma_0 \times (p/p_0)^n$ relation predicted by Eudier\(^{10}\).
POWDER FORGING OF COMPOUND STEELS

PART I

B. Arén

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Abstract

Compound bodies consisting of ductile and hardened material have been studied in order to determine the effect of tensile stress perpendicular to the interfacial plane. Samples with both sharp and diffuse interfaces have been investigated. It is shown that compound structures with considerable interfacial strength may be readily produced by powder forging. High compound strength is shown to be favoured by a sharp interface between regions of dissimilar material.
2. Introduction

Powder forging is a process by which metal powder in a cold die-tool is compacted to a porous preform. This preform is then heated to forging temperature and hot forged, bringing the material to full density and simultaneously forming the body into its final shape.

The main advantages of the process are: freedom to produce specific alloys; close dimensional and weight tolerances; excellent surface finish; insignificant material waste and little or no need for secondary machining operations. The development of powder forging methods have made it possible to produce material of full density and strength in an economic manner (1). The technique has so far mainly been used to produce bodies of monolithic material.

By combining different types of materials within the same body a great variety of composite structures can be made, and several applications, e.g. tungsten reinforced turbine blades have been reported (2). However, the possibility of combining different materials by powder metallurgy appears to have been limited to true composites, whereas compound materials with a macroscopic distribution of components have not been reported, figure 1.

![Figure 1. Composite material, left, and compound material, right.](image)

Several patent applications show the commercial interest for the method, one example being the production of bushings
in a concentric distribution (3). Other applications are gripping, cutting and forming tools, transmission parts and similar load-bearing components where a hard surface is required in combination with a ductile base material. For successful performance of such compounds it is of great importance to ensure that the interface should not be the cause of failure.

The object of this and a following work (4) is to examine whether compound materials can successfully be made by powder forging, and to investigate the mechanical behaviour of such compounds with hard/ductile properties.

From experiments on hard-ductile laminates (5, 6) it is found that interface boundaries can act as crack arrestors and raise the impact strength to very high levels. This is also confirmed in the proceeding work.

In an attempt to achieve the lamellar effect in a transverse application an irregular interface was chosen. Both sharp and diffuse interfaces have been studied in order to examine the importance of controlling the interfacial structure.

In this report the effect of tensile stress perpendicular to the interface is studied, and the next report will present the effect of tensile stress parallel to the interface.

3. Experimental procedure

In this study five different powders were used: a water atomized pure iron powder, Hoeganaes AHC 100.29; two pre-alloyed powders, Hoeganaes ATST-A and A o Smith 46 F2; and two diffusion alloyed powders, Hoeganaes Distaloy-AB and -AE. Graphite was added to give a carbon content of 0.45% in the final products.
The main chemical compositions are given in table I.

<table>
<thead>
<tr>
<th></th>
<th>Mn</th>
<th>Ni</th>
<th>Mo</th>
<th>Cu</th>
<th>Fe</th>
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<td>-</td>
<td>Balance</td>
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<td>1.50</td>
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</tr>
</tbody>
</table>

Table I: Chemical compositions of the powders. Minor amounts of other elements are neglected.

Three types of test bars with dimensions 100x10x10 mm:s were prepared. Type A is a monolithic reference body and types B and C are compound bars with sharp and diffuse interfaces, figure 2.

![Figure 2](image_url)

Figure 2. Forged bodies tested. Type A (top figure) is monolithic, types B and C (bottom figure) are compounds with sharp and diffuse interfaces, respectively.

The precompaction was carried out using a floating die tool in a hydraulic press to a constant pressure of 785 MPa. Lubrication of the die tool was achieved by "painting" the tool surface with a solution of zinc stearate in ethanol. No lubricant was added to the powder. The density after cold compaction was roughly 90%.
The test bodies of types B and C were made with three material zones using the same type of material in the ends of the body in order to achieve a force balance both in preform compaction and in the forging operation. In order to localize the powder portions to given sites, a funnel with 3 divisions was put into the die cavity. After the powder portions had been placed into the correct sites and precompacted slightly, the funnel was carefully removed so that the powder was not disordered (type B). After this operation the preforms were compacted in the usual manner.

In order to obtain a diffuse interface in the type C preforms the funnel was equipped with mixers that combined both alloyed and unalloyed powders in the interface zone when the funnel was removed from the die.

Prior to forging the preforms were heated in a tube furnace for 10 minutes at 1050°C in a hydrogen atmosphere. Transport to the forging die, which was heated to 275°C, took 2 to 3 seconds, causing a temperature drop in the centre of the preform of about 30°C as recorded by a thermocouple.

The forging strokes were made by a friction screw press, and the actual forces were controlled by a load cell and registered by a light beam recorder. Because of elastic deformation of the die tool it was impossible to use a higher forging pressure than 900 MPa.

Although the strokes were initiated from the same starting position, the resulting pressure varied, causing a density variation between different forgings. Another reason for the variation was the difference in flow stress between different materials being forged. Forged pieces with overall densities of 98 - 99% were used for tensile tests, and samples from the interval 96.5 - 98.5% were used for impact tests.

Densities given are overall values which means that the
centre density of the samples is higher, a well known result from powder forging in closed dies (7, 8). Accordingly the centre density of the pieces is assumed to be 99.5% or even higher.

Specimens were prepared for tensile testing and impact testing (Charpy key-hole) as shown in figure 3.

![Figure 3. Compound bars for tensile (left) and impact (middle) testing. Cross-section at the key-hole notch, right.]

The tensile test bars were made by turning. The interface was confined to the middle of the bar in order to achieve symmetry and avoid influence from anomalies in the stress field introduced by clamping the ends. An advantage of turned specimens is that the material in the test section of the bar has the maximum density.

The test bars for impact tests were designed to Swedish Standard SIS 112350. As shown in figure 3, the key-hole shaped notch was positioned at the interface. In order to locate the notch hole correctly, it was drilled and broached prior to the heat treatment. The saw cut to complete the key-hole notch, however, was inserted after the hardening operation to avoid distortion of the test bar.

All specimens were heat treated as follows:

1. Austenitizing for 30 min. at 850°C in hydrogen.
2. Quenching in oil preheated to 70°C.
3. Tempering for 60 min. at 275°C.
The procedure was chosen in order to give maximum difference in hardness between alloyed and unalloyed material. The hydrogen atmosphere afforded good protection against oxidation, and only a very thin surface oxide layer was formed which was therefore neglected. The oxidation was probably caused by the transport from the furnace to the oil quench bath.

Tensile testing was carried out in a test machine especially designed for small test bars. As a consequence of the test bar size, the strain gauge was attached to the force mechanism of the machine which unfortunately resulted in a somewhat lower precision, cf figure 5. Impact testing was undertaken with a conventional Charpy hammer.

After tensile testing some bars were cut lengthwise and the hardness along the bar axis was determined from Vicker's hardness impressions (300p) in order to examine the interface.

4. Results

Results of tensile testing are presented in table II.

Stress-strain curves for both monolithic and compound materials were drawn, and some examples are shown in figures 4 - 6.
<table>
<thead>
<tr>
<th>Material</th>
<th>Type</th>
<th>$\sigma$ [MPa]</th>
<th>$\varepsilon$ [%]</th>
</tr>
</thead>
<tbody>
<tr>
<td>AHC 100.29</td>
<td>A</td>
<td>490</td>
<td>11</td>
</tr>
<tr>
<td>AoS 46 F2</td>
<td>A</td>
<td>1560</td>
<td>0.5</td>
</tr>
<tr>
<td>ATST-A</td>
<td>A</td>
<td>980</td>
<td>1</td>
</tr>
<tr>
<td>DISTALOY-AB</td>
<td>A</td>
<td>500</td>
<td>13</td>
</tr>
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<td>DISTALOY-AE</td>
<td>A</td>
<td>980</td>
<td>4</td>
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<td>2</td>
</tr>
<tr>
<td>AHC 100.29-AoS 46 F2</td>
<td>C</td>
<td>580</td>
<td>5</td>
</tr>
<tr>
<td>AHC 100.29-ATST-A</td>
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<td>610</td>
<td>9</td>
</tr>
<tr>
<td>AHC 100.29-ATST-A</td>
<td>C</td>
<td>600</td>
<td>6</td>
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<td>AHC 100.29-DISTALOY-AB</td>
<td>B</td>
<td>470</td>
<td>6</td>
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<tr>
<td>AHC 100.29-DISTALOY-AB</td>
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<td>7</td>
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<td>AHC 100.29-DISTALOY-AE</td>
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<td>AHC 100.29-DISTALOY-AE</td>
<td>C</td>
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</table>

Table II. Ultimate tensile strength and total elongation from 4 - 5 specimens of each type. Individual test results differ less than 7% and 20% from the mean values noted in the table.

![Stress-strain graph](image)

Figure 4. Stress-strain graph for monolithic AHC 100.29 material expressed as applied force divided by original square section versus elongation divided by original length of the test bar.
A fully hardened ATST-A-material was chosen as one of the partners to the ductile AHC material. The heat-treatment undertaken gave this material a very hard and brittle structure. Fractures of monolithic ATST-A-bars therefore always occurred near the clamping heads indicating a true sensitivity to stress concentrations. This can readily be explained by underoptimal annealing, figure 5.

Figure 5. Stress-strain graph for monolithic ATST-A-material. Deviation from Young's modulus is a result of a clamping fault.

As expected, the curves produced by the compounds could have been predicted by a simple half and half superposition of the monolithic curves from the relevant materials. This is illustrated in figures 4 - 6 and 17.
Figure 6. Stress-strain curve for a AHC 100.29-ATST-A compound with a sharp limit between the two materials.

In compounds with both sharp and diffuse limits between materials involved fractures were found to occur in the more easily strained AHC-material, fig 7.

Figure 7. Typical compound test bar of AHC-ATST-A type with a sharp, (upper figure) and diffuse (lower figure) limit between materials involved. Fractures were always located in the AHC-part.

The localization of radial shrinkage and succeeding fracture is a natural result of the mechanical resistance of the coarse head and the hard structure, respectively, at the two ends of the mild steel zone.
On studying the test bars in figure 7, a clear difference in colour is also seen between the two parts of the bars which illustrates well the two types of connections. This is a result of the fact that the oxide layer formed during the heat treatment easily cracks and falls off when the surface is extended during the test.

Plots of hardness measurements taken from the axis of the test bars show that the required sharpness of the borderline between the two materials is reached, figure 8.

![Graph showing hardness measurements](image)

**Figure 8.** Vicker's hardness (load 300p) along the fully dense centre part of the axis of a AHC 100.29-ATST-A test bar with sharp borderline.

The sharpness of the interface is also verified by metallography, figure 9.
Figure 9. Etched structure (1 min in 3% nital) of a sharp interface between fully dense structures of AHC 100.29 and ATST-A steel in normalized (light) and martensitic (dark) condition. Magnifications 43.5 X, 168 X and 1260 X respectively.
The diffuse type of connection is more complex as it is produced by local mixing of the powders. The hardness impressions are large in comparison with the size of the powder particles, thus the diamond pyramid indents both soft and hard particles in the same impression in the diffuse zone. Although the impressions were distorted, the measurements made in the diffuse zone clearly show the effect of mixing. From a AHC-AoS 46 F2 combination presented in figure 10 it can be seen that the diffuse zone is more than 1 mm wide.

![Graph](image)

Figure 10. Vicker's hardness (300p) along the fully dense part of the central axis of a AHC 100.29-AoS 46 F2-compound with diffuse borderline.

Charpy impact testing was performed using a conventional impact tester with a potential energy of 147 J. The results, figure 11, are not directly comparable with other impact measurements because the test bodies were not exactly of standard type.
Figure 11. Charpy impact diagrams.
5. **FEM-analysis**

In order to examine the stress-strain situation at the interface an FEM-analysis was undertaken. A compound bar stressed in tension of AHC 100.29 and ATST-A material with a sharp interface was studied using the computer program ADINA (9) which is designed for the analysis of plastic deformation processes. The program, originally developed by Prof. K.J. Bathe, at MIT, Boston, USA, was completed with a post processor ADINAPLOT (10) for presentation of the results. Practical versions of the programs have been refined by M Sc G. Berglund and M Sc B. Ekmark, both at the University of Luleå. Mr Berglund also performed the calculations.

Calculations were performed under the assumptions that the behaviour of the materials investigated is linear elastic and linear elastic-linear plastic with isotropic hardening respectively. The risk of fracture nucleation was taken into consideration by comparing the calculated stress with the expected fracture stress, and no calculations were done exceeding these stress levels.

For both materials Young's modulus was set as $0.21 \times 10^6$ MPa and Poisson's ratio as 0.3. The hard phase material is assumed to be fully elastic within the whole stress range. For the ductile material a yield point of $0.43 \times 10^3$ MPa was set followed by isotropic hardening with a modulus of $0.14 \times 10^4$ MPa, which means that some plastic strain takes place before fracture in this zone. This choice of material parameters was based on test results previously presented, figures 4 and 5.
The cylindrical transversely divided bar was analysed using cylindrical coordinates with a FEM-grid pattern shown in figure 12.

In order to limit the calculations, the end heads of the bar are not analysed. The resistance to radial tension caused by the heads is compensated for by an adjustment of the boundary conditions.

Figure 12. Test bar studied, left, and element structure used, centre. Local section further developed in figure 13, right.
Artificial tensile load was applied in a number of steps, the first 10 of which were large and used to reach a stress level just below the point of first yielding, the succeeding ones were one decade smaller in order to keep the effects of local yielding under control. After applying 37 load steps the linear tension in the mild phase reached 11.9%, a value to be compared with the linear elongation at fracture, 11%, as measured on a monolithic bar, figure 4.

Computer plots of total equivalent stress, total shear stress and incremental equivalent strain at this stage were derived for the chosen section of the bar. The plots are shown in figure 13, where the dotted area beneath the horizontal line is associated with the hard phase.

![Figure 13. Computer plots of chosen section of compound bar showing equivalent stress (MPa), left, shear stress (MPa), centre, and incremental strain (%), right.](image)
The maximum effective stress found in the mild steel phase at this stage was 639 MPa located at the centre axis 2.6 mm (2.4 mm before straining) away from the hard/mild interface. This distance δ compared with the bar diameter φ corresponds with a δ/φ-ratio of 0.60 before straining.

The plot of effective stress, figure 13 left, does not correctly present the situation at the interface. In reality a stress step is developed, caused by the fact that the local strain at the interface is of the same order of magnitude in both materials, and parallel to the interface the strain is equal. In the unplasticised hard phase the stress is forced to a high level whilst the competitive stress in the mild phase remains low, associated with plastic flow. The plastic strain however is very small and, as a result of the restraining effect, the hard phase depresses the stress of the mild phase close to the interface. Figure 14, combining the effective stress with the effective strain illustrates this statement.

From figure 14 it is evident that stress values given for interface nodes are integrated mean values derived from the adjacent integration points. A better idea of the true stress situation at the interface is therefore given by the extrapolation towards the interface of the calculated values from the mild and hard material zones respectively. A conclusion of this is that the iso-stress lines in the ductile phase close to the interface are overestimated.

By studying the shear stress-plot, figure 13 centre, one can find a maximum value of 188 MPa at the interface, but in comparison with the effective stress this shear stress value is low.

The map of the incremental effective strain, figure 13 right, shows that the maximum straining is found along the central axis 2.3 mm (2.1 before straining) away from the interface, corresponding to a δ/φ value of 0.53 before straining.
Figure 14. a) Calculated effective stress.  
b) Associated effective strain (in principle).  
c) Stress-strain curves for hard and mild phase steels.

Due to the chosen element size figures 13 and 14 do not correctly describe the conditions close to the interface but nevertheless it expresses the stress situation rather well. By using smaller elements a more accurate result will be obtained however.
6. Discussion

For tensile straining of a compound of serial type the elongation is given by the equation

\[ \varepsilon = f_1 \cdot \varepsilon_1 + f_2 \cdot \varepsilon_2 \]

where \( f_1 \) and \( f_2 \) are fractions of the two materials. The theoretical stress-strain graph for such a compound can be derived in accordance with the method given in figure 15.

![Stress-strain curve for a serial-compound strained in tension.](image)

Figure 15. Stress-strain curve for a serial-compound strained in tension.

Applying this theory to the test bars investigated, one can derive stress-strain graphs like the one shown in figure 16.

![Predicted stress-strain graph for an AHC-ATST-A compound type B. To be compared with figure 6.](image)

Figure 16. Predicted stress-strain graph for an AHC-ATST-A compound type B. To be compared with figure 6.
A comparison between measured and calculated graphs shows good correspondence. One should be careful however, not to take the conclusions too far. Different factors may be affected by discrepancies e.g. the amount of material of each phase is double in monolithic bars compared with the compound ones.

It is a well known fact that the larger the stress affected volume is, the lower the tensile and the yield strengths will be. A stress-strain graph derived from testing a compound would therefore be expected to be higher than a corresponding curve derived by calculations on the basis of test results from monolithic bars. In the latter case the double volume of test material is used to give the stress-strain master curves.

This volume effect is even more obvious with the impact test bars with sharp interface where all the compound bars exhibit larger impact values than monolithic bars of the material having the lower impact strength involved. This effect might be expected however as impact testing gives a more brittle material behaviour than tensile testing, and that the size dependent effect is more pronounced for brittle materials.

A mild/hard test body with sharp interface can in theory only provide half as large a volume for the localization of fracture as a monolithic body. In reality the actual volume is even smaller because of the restraining effect of the hard zone. This is also confirmed by experiments showing that fracture occurs in the mild zone away from the interface.

A diffuse interface seems not to be favourable, the most striking example of this is the AHC-ATST-A combination where a very low impact strength is recorded.

Considering the impact strength of the two basic structures,
one reason for the poor result of the diffuse combination may be found: In the mild steel the ductility is important and in the hard steel the hardness is valuable.

In a soft ductile material coarse islands of hard phase will diminish the ductility and mild islands located in a hard material will correspondingly diminish the hard volume available. For both phases therefore coarse islands of the opposite phase will be disadvantageous, and for compounds with a diffuse interface this might be the reason to the poor impact strength found. This assumption is also justified by the well known effect of reinforcement of plastics with fibre glass (11). In applications where ductility of the plastic matrix is important for the total strength an increased addition of glass reduces the strength, which is associated with the low fracture strain of glass. Correspondingly an increased addition of plastic phase will reduce the strength of a glass structure in situations where the fracture stress is most important.

7. Conclusions

The work presented in this report shows that compound structures with considerable strength of the structural interface can easily be produced by powder forging. A new method for the production of valuable compounds by hot working in addition to earlier methods is therefore made available.

In comparison with the weak material involved both the tensile and the impact strengths typically increase due to compounding with a hard phase. The strength of the compound is favoured by a sharp connection between the two phases.
8. Acknowledgements

A basis for this work is a masters thesis by M Sc Håkan Bodsten, which was performed at Linköping University under my supervision. In his work Mr Bodsten studied the tensile strength of interfaces between hard and mild steels loaded in tension transversely the interface. This was done in tensile testing as well as in impact testing of Charpy key-hole bars. For the permission to use and develop his results he is gratefully acknowledged.

The computer calculations was executed by M Sc Gunnar Berglund and typing and graphic work was done by Miss Carola Löfgren and M Sc Esa Vuorinen. Language corrections have been made by B Sc John C Ion and Ph D Harald Herbertsson has contributed with valuable criticism and helpful discussions. They are also gratefully acknowledged.
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POWDER FORGING OF COMPOUND STEELS

PART II

B. Arén

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Powder Forging of Compound Steels Part II

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Abstract

Compound bodies made by powder forging of ductile mild steel and highly hardened alloyed steel have been studied to determine the effect of tensile stress parallel the interfacial plane. The material produced is examined in tensile testing and in 3-point bend testing. It is shown that fractures nucleated in the hard structure zone are arrested by the ductile phase on reaching the interface. In comparison with monolithic bodies of the two materials the compound structure is advantageous for the geometries investigated. The improvement of the product is a result of the combination of hardness and ductility from the two base materials.
2. **Introduction**

In an earlier report (1) the powder forging of low alloy and mild steel powders in transversely combined compound steel specimens was studied. The test material was produced by heating precompacted green bodies in a hydrogen atmosphere prior to forging in air. The main conclusions from the mechanical tests undertaken were that a weak material could attain higher strength by combination with a harder compound, and that the interface between the two materials should preferably be sharp.

In this report compounds with an interfacial plane parallel to the stressed direction are studied. Another difference is the choice of material; those used in this study are mild steel AHC 100.29 with 0.45% carbon with a partner material of highly hardened M2 high speed steel having a carbon content of 0.85%. In order to protect the high speed steel against oxidation the test bars were encapsulated during the hot working procedure.

3. **Experimental procedure**

One reason governing the choice of material was the attempt to produce as large a difference in hardness as possible between the resulting structures, although the AHC-material was also chosen to be equivalent to the material used in previous work.

The M2-powder was made by Davy-Loewy Ltd, UK, and was specified with an oxygen-content of about 300ppm. This powder was delivered in annealed form and could, with some difficulty, be compacted into laminated green bodies together with the AHC-powder. Because of the lower compressibility of high speed steel powder however, this component of the green bodies attained a lower green strength than the mild steel part.

The compacted laminated preforms were machined to 7.5 x 13 x 96 mm:s and then encapsulated in iron sheet capsules with external dimensions 10 x 15 x 100 mm:s as designed to fit the forging die tool. Prior to the introduction of the preforms
the capsules were internally painted with collodium and powdered with titanium powder. The preforms were put into the capsules which were then sealed by welding a cover to the open end.

To prevent further oxidation of the high speed steel some precautions were taken: The capsules were fitted to steel tubes by which they were evacuated during the preheating operation so that the gases produced e.g. by vaporization of the zinc stearate addition could be removed. At the same time the internal layer of titanium acted as an oxygen getter, a method earlier used by Olsson et al (2).

The specimens were preheated to $1240^\circ$C for 30 minutes in a tube furnace and then pulled by the evacuation tube into the die tool where the material was compacted in one stroke to full density. The evacuation of the capsules was performed using a vacuum-pump and the degree of vacuum in this process was continously checked in order to ensure that there was no leakage that could allow the atmosphere to enter the capsules. These precautions were probably more rigorous than necessary, but the resulting material behaviour was fully acceptable.

After forging at about 900 MPa the compact material was kept at $875^\circ$C for 2 hrs for stress relief and was then heat treated to give maximum strength to the high-speed steel. The procedure was as follows:

1. Austenitizing at $1200^\circ$C.
2. Furnace cooling to about $150^\circ$C, taking about 1 h.
3. Annealing at $650^\circ$C for 1 h.
4. Cooling to room temperature, taking about 1 h, 15 min.

The material hardness after heat treatment was found to be 650 HV units in the high speed steel phase and 180 HV units in the mild steel, using a 300g test load.
The mechanical behaviour of the material was examined in tensile testing and in bend testing. The tensile test bar was designed with two parallel zones of high-speed steel at the edges surrounding a centre zone of mild steel.

After forging and stripping the capsule the sample had a rough surface and was therefore machined by grinding to produce the test bar geometry shown in figure 1.

![Tensile test bar of compound high speed steel/mild steel with parallel geometry. Mild steel zone in the centre. High speed steel shaded.](image)

Figure 1. Tensile test bar of compound high speed steel/mild steel with parallel geometry. Mild steel zone in the centre. High speed steel shaded.

Two layer bend test bars were also produced in the same way with square section 6 x 6 mm:s, each material layer 3 mm:s thick. The types of material, production method and heat treatment were the same.

In order to eliminate effect of deformation hardening in the mild steel phase due to the grinding operation, the test bar was annealed at 200°C for 10 minutes and subsequently cooled in air.

The tensile test bar was strained to fracture and the fractured bar was studied in the SEM in order to reveal the deformation at the border zone.

Bend testing was performed as a 3 point test with the hard zone exposed towards the applied load, a stress situation reflecting realistic applications. In this test it was also
possible to locate the position of the initial stress fracture and to study the fracture development from this original crack. Simultaneously the bending load as a function of actual displacement was recorded in order to measure the energy used during deformation. In order to be able to study the fracture development, the oxide layer was removed by polishing before the bend test was started.

In addition to mechanical tests comparisons with results from FEM analytical calculations were made.

4. Mechanical tests and FEM-analysis

Tensile test

The tensile test experiments were undertaken in order to examine whether the interface parallel to the strain direction would fail or not. Another aim was to study the crack arresting effect of interfaces as reported for laminated bodies by different authors (eg 3, 4).

As expected, the fracture of the high speed steel components was brittle and occurred prior to the ductile fracture of the mild steel component. In spite of the fact that the fracture characteristics were quite different in the two materials, the interface did not delaminate, figure 2. The degree of border zone deformation close to the final crack surface was very extensive, but no interface failure appeared.

Because of the different sensivity to oxidation of the two steels a clear oxide pattern was formed on the surface of the test bar clearly showing the sharp boundaries between the two types of steels. Figure 2, left. Mild steel dark.

Interesting questions about compound materials are to what extent local stress is raised by the hard-ductile boundary and whether the material will accommodate this stress. As
revealed using light microscopy and SEM, the interface appears to be very tough in the sense that it did not separate in spite of heavy local deformation.

Figure 2. Tensile crack in laminated test bar. Left: side view, 8x. Centre and right: front views of the same fracture surface by the SEM, 60x, and by light microscopy, 16x, respectively. Parallel lines seen in left and centre photos are traces from the surface grinding.

In order to establish the strain and the stress pattern in the test bar when the brittle zones had cracked, a FEM-study was carried out. A symmetrical approach was used to form the FEM-network, which means that only the process after cracking of the hard zones is studied, figure 3.

The element structure as shown in figure 3 contains very small elements in the crack tip zone. Surrounding element are gradually made larger until the maximum element size is reached.
The computer programs ADINA and ADINAPLOT were used again (5, 6). A linear elastic - linear plastic model with isotropic hardening was chosen to characterise the material behaviour. The nucleation of fracture is taken into consideration by comparing the calculated stress with the expected fracture stress.

![Figure 3. Model of test bar after cracking of hard material section with the FEM-studied zone line screened, left. Element structure for the FEM-analysis, right. Dark centre section of the right hand graph is further developed in figure 4.](image)

For both materials, Young's modulus was defined as $0.21 \cdot 10^6$ MPa and Poisson's ratio as 0.3. The hard phase material is assumed to be fully elastic up to a yield/fracture stress level of $0.305 \cdot 10^4$ MPa, thus fracture was defined by the attainment of this value. For the ductile material a yield point of $0.43 \cdot 10^3$ MPa was fixed followed by an isotropic strain hardening of $0.14 \cdot 10^4$ MPa, which means that plastic strain takes place before fracture in this zone. The choice of material parameters was based on earlier measurements (1), and results from Olsson et al. (7).
The FEM-structure was stressed in a stepwise manner by applying artificial load in 14 steps before reaching the stress level of fracture. The first ten of these were used to attain the flow stress in the ductile phase, but the following four loadsteps resulting in continued plastic strain were considerably smaller.

By comparing the calculated area contraction close to the borderline with the actual contraction as measured on the broken bar, the point of fracture can be estimated to be reached in between loadsteps 14 and 15. This also corresponds to the true tensile strength of the ductile material, expected to be around 1000 MPa. The maximum stress in the ductile zone close to the hard material was calculated to be 968 MPa and 1702 MPa in loadsteps 14 and 15 respectively.

By calculating the value of step 14, the maximum shear stress was found to occur in the hard zone at about 0.1 mm:s away from the interphase boundary line and about 0.2 mm:s from the fractured surface. The maximum shear stress level at this point was 570 MPa and the maximum value at the interface was about 250 MPa, estimated from the computer list.

Magnified local plots of element grid, incremental strain, equivalent stress and shear stress are presented in figure 4.

When analysing figure 4 however, it must be noticed that the computer plot program (4) is not totally relevant for the execution of the calculated results adjacent to the borderline. In the reality the stress increase/decrease is infinitely sharp across the interface. This means that the stress in the hard phase in fact is higher and in the mild phase is lower at the interface than what is shown e.g. by figure 4c. The reason is discussed in detail in the previous report (1).
Considering that the strain of the two phases must be consistent at the interface, also expressed in figure 4b, one should remember that the stress in the hard steel phase is essentially larger than the corresponding stress in the adjacent mild steel phase. The stress gradient perpendicular to the interface e.g. shown in the equivalent stress plot, figure 4c, is therefore not true. In fact a stress step increase would better illustrate the actual stress situation. With these reservations in mind however, the plotted figures seem to illustrate the stress-strain situation very well at the crack tip, and the use of smaller elements will make the results even better.

In order to obtain references for the compound bar studied, monolithic bars of the hard and the mild steel types of the same geometry were analysed by FEM. Plots of equivalent strain
from these calculations are given in figure 5, which presents stress patterns where the difference in plastic deformation between the two materials results in a different degree of stress concentration, as would be expected.

![Graphs showing stress patterns](image)

**a: high speed steel**  
**b: mild steel**

Figure 5. Local plots of equivalent stress for a monolithic material of; a: the hard and; b: the mild steel type. To be compared with the compound material shown in figure 4c.

**Bend test**

Bend testing took the form of a 3-point bend test with the ductile material in the elongated zone and the hard material in the compressed zone of the test bar. During the test the ductile zone is elongated in plastic deformation. The hard zone is strained elastically, and reaches the critical level of tension after a period of time. Fracture nucleates in the hard phase, initiated at the boundary zone, and the development of fracture in the hard phase is then associated with a local plastic strain of the ductile phase, figures 6 and 7.

From figure 6, it is evident that the ductile zone plasticises at a displacement of 0.20 mm. In the interval 0.20 – 0.96 the slope of the graph changes during the development of a fully plasticised ductile material zone. At 0.96 mm a rapid fracture develops in the hard phase after nucleation close to the interface.
Figure 6. Load-displacement graph from bend testing of a 2-layer material. The elongated zone consists of ductile material.

Figure 7. Bend test bar of the 2-layer material. The ductile part is stretched in the strained zone while the hard part is almost completely broken. 3X.

From figure 7 one can see that the interfacial zone is not separated even with a large local tension in the ductile zone. The shape of the hard zone crack will be discussed later.
An FEM-analysis of the bend test bar was performed. The node structure used was in principle the same as in the tensile test study, figure 8. This study was performed under the condition prior to reaching the fracture stress in the hard material zone.

Figure 8a shows location of applied load, the element grid system and a magnification of the deformation.

Figure 8b shows a graph plotted of the incremental effective strain. As can be seen in the figure there is no effect at the interface between the two materials although the hard material is elastically strained while the ductile material is plastically strained. This is to be expected however as the strain in this case is strongly effected by the geometry.

Figure 8c shows a plot of the equivalent stress and illustrates a stress concentration in the hard phase close to the interface well focused to the point of the forthcoming fracture nucleation. The maximum value of equivalent stress was 3049 MPa, in accordance with the measured tensile strength of the hard material, 3050 MPa (7).

To balance the tension in the extended zone also a corresponding compression is developed at the top surface. As a result of the compressive character this stress does not lead to fracture. In the centre section of the hard phase a local stress minimum therefore occurs below the point of load application.

As a result of the concentrated load the stress field is centred to the region around the symmetry plane. In a bend situation where a concentrated load is not applied, the results would be similar, however.
Figure 8. Computer plots from bend test; a: element grid, b: incremental effective strain, c: effective stress (MPa), and d: shear stress (MPa).
Although the strain in the majority of the ductile zone is very small, the calculated stress level indicates that the material is fully plastic. In a true material however, the yield strength is expected to differ from site to site and uniformity of plastic strain is therefore not guaranteed in practice.

Figure 8d shows the shear stress pattern in the deformed bar with the maximum calculated value of 409 MPa focused to a point in the hard phase. No concentration of shear stress can be found at the interface.

5. Discussion

When analysing the laminated bar strained in tension an interesting effect is found in the localization of the equivalent stress. In the compound bar the maximum stress in concentrated to the hard phase adjacent to the crack tip. In front of the crack tip the stress is much lower as a result of the plastic deformation of the mild steel phase, and a crack arresting effect is therefore to be expected in this type of interface.

Comparing the compound bar with a monolithic mild steel bar, figures 4c and 5b one can find that the high stress region in front of the crack tip apparently is less extended in the compound specimen. The reason to this might be that the hard phase will restrain the material deformation perpendicularly to the crack direction. If this also will result in a better crack arresting effect than will be offered by the monolithic mild material may be the topic for a futural investigation. An important improvement however is reached by the fact that the total deformation of the compound bar is much less than the corresponding deformation of a monolithic mild steel bar at the same level of applied load.

In comparison with a hard monolithic specimen, figures 4c and 5a, the improvement offered by the compound is even more obvious as a result of the substantially decreased stress peak at
the crack tip. The total elongation to be offered before fracture of the compound bar therefore is of a larger magnitude.

On studying the load-displacement graph from the bend test some intervals of interest can be found. Models of the tensile stress distribution for certain parts of the graph are given in figure 9.

![Figure 9](image)

Figure 9. Stress situations in principle in the bend testing of the compound bar. $\delta$-values in accordance with figure 6.

Picture a) shows the situation before plastic yielding, picture b) the situation of partial yielding and picture c) the condition after full yielding of the mild steel phase. Pictures d) and e) show the stress distribution prior to and after fracture of the hard phase.

From figure 9, it can be seen that the hard phase is principally stressed in compression but that tensile stresses occur locally after plasticising of the mild steel phase. Tensile stresses in the hard phase are then set up at the interface and the stressed area increases during continued deformation. The maximum value is reached just before the crack initiation at $\delta=0.96$ mm, corresponding to figure 8c and 9d.
When once started the crack propagates very fast perpendicularly to the interface as a result of the stress concentration of the crack tip until that force balance is achieved for the actual geometry. Later on the crack opening is slowly extended during the continued bend test and then the crack direction is turned towards the point of load application, figure 7.

After crack occurrence the ability of the hard phase to bear tensile force is diminished. One reason is that a substantial area is fractured. Another reason is that the sharp crack concentrates the stress so that the nominal tensile stress to be born is reduced by a large stress concentration factor. During continued bending the portion of hard phase loaded under tension is continually diminished towards zero. Up to the point of final fracture which is a result of strain embrittlement of the ductile phase, only the hard phase carries compressive forces.

On the basis of fact that yielding starts in the ductile zone while the first fracture nucleates in the hard zone from the interface plane, comparisons with monolithic materials will be made.

As a result of the geometry the primary fracture of a monolithic hard phase bar is to be expected after half the displacement as for the compound bar when assuming that the influence of the ductile phase may be neglected.

In the compound however, the presence of the mild steel layer brings the position of neutral stress down the centre line of the hard phase zone. The mild steel layer will then suppress the primary fracture. At the same time plastic deformation of the ductile phase is associated with the movement of dislocations. Dislocations pile up at the interface and form crack nucleating pores. These diminish the normal fracture stress of the hard phase and the mild phase layer therefore can also promote fracture nucleation. The total effect of these two mechanisms is complex, and is not studied in detail. In the following discussion their influence is neglected.
In a mild steel bar primary yielding is expected at the same place as in a compound bar, and the yielding then continues across the bar until the whole cross section is plasticized. A graph from a bend test of such a material therefore differs from the graph in figure 6 in the following ways: Softening starts at the same stress and strain levels but as a monolithic mild steel cannot sustain the applied load as effectively as the hard phase of the compound the increasing slope of the graph after primary yielding is substantially lower. Some increase in load resistance is expected during the plastic flow however.

Also when the hard phase of the compound is broken it is still able to carry compressive force. In addition to this, it also carries a small portion of the tensile force. The load carrying ability of the compound bar is therefore always larger than that of a monolithic mild steel bar.

As a function of the crack opening, illustrated in figure 9e the straining of the ductile material will be more uniform. This is an improved way of using a mild steel section in a deformed bar.

As the ductile part of the compound carries only tensile stress, a comparison with a monolithic mild steel bar predicts that even after fracture of the hard phase the compound bar carries at least the same load as a monolithic mild bar.

The expected graphs of monolithic hard and ductile materials are superimposed onto the load-displacement diagram from the compound bar bend test, figure 10.
Figure 10. Bend test graph of a two layer material, a) expected graph from bend test of monolithic hard, b) and ductile, c) bars.

Shown by figure 10, the work necessary to fracture a monolithic bar is always smaller than the corresponding work for complete fracture of a compound bar. In fact a monolithic hard material needs even less work to final fracture than needed for primary crack of a compound bar. The maximum force is less for the compound, but there is a substantial improvement of impact strength.

Comparing the pure mild steel with the compound bar it can be seen that no strength is lost by replacing the fully ductile material with the compound used.

6. Conclusions

From the results it can be seen that a longitudinally designed compound bar loaded in tension will primarily fracture in the hard phase. This crack however does not readily propagate into the mild steel zone. When considering the risk of fracture nucleation a compound bar loaded in pure tension exhibits superior behaviour to that of both monolithic high speed steel and monolithic mild steel.
In bending the improvement is even more obvious provided that the mild steel is confined to the strained zone.

For a compound with the hard phase in the strained section an elevation of the primary yield point can be expected, but the risk of cracking of the hard zone requires certain precautions, as primary fracture propagates very fast once it has initiated. It is arrested by the ductile zone however, but this only occurs when a substantial part of the square section is broken, and at that moment the remaining stiffness is less than $1/8$ of the original.

Summarizing the results however the study has shown that valuable compound products can easily be made by hot working of powders. The full potential of products may be realised if the direction of applied force is taken into consideration. For a situation in which contact forces are applied to the hard phase material the compound is very advantageous. In situations where tensile stresses of significance are induced in the hard part, the compound structure can however be unreliable.

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