PRODUCTION OF TWO-PART COMPONENTS

BY POWDER FORGING.

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by

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DEDICATION

To: My Mother; My Father;

My Sister; Nimet

And

My Brothers; Mehmed Firinci,

Dr. H. Ibrahim And Mahir Basdogan

SYNOPSIS

The work presented in this thesis is an investigation of the hot forging of powder preforms in which different powders are used in various parts of the forging. A consideration of the possible advantages and applications of this type of forging is given with reference to the relevant published literature.

In the powder compaction part of the process the powders were pressed in two layers. This was followed by the sintering operation, after which the preform was hot forged in a closed die to give simple consolidation without appreciable flow. The two powders used in these tests were ATST-A low alloy atomised steel powder and AHC 100.29 atomised iron powder.

Tests were carried out over a range of processing conditions, and mechanical and metallurgical tests were performed in both the as-sintered and as-forged conditions. Particular attention was paid to the properties of the bond between the two materials.

Recommendations are made concerning the processing conditions under which the optimum properties are obtained for the particular powders being investigated. Finally proposals are made for future work on the forging of two part powder preforms.

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CHAPTER ONE

INTRODUCTION

The development of powder metal forging (also referred to as powder forging or sinter forging) has been proceeding for some 14 - 16 years. The hot forging of powder preforms particularly has emerged and developed during the last decade as a promising technique to overcome the deficincies of both classical powder metallurgy and conventional forging from solid billets.

The underlying principle of the process is universal, but there are variations in detail. The objective is to produce components of finished tolerance or requiring only a final grinding operation.

The powder forging technique is based upon the application of both pressure and heat on to the powder preforms. The preform is designed in such a way as to fill the forging dies in the most beneficial manner when it is forged.

The preparation of powder preforms is a cold compaction process. Pressure is applied to cold powder in a compaction die by using a suitable compaction press. Cold compaction is followed by the sintering operation in which the preforms are kept at certain temperature for a predetermined time. In order to prevent oxidation in the porous structure, the sintering operation is carried out in a controlled atmosphere furnace. The objective of the sintering operation is to provide sufficient bonding between the particles to enable the preforms to be hot worked.

After sintering, the preforms are either immediately hot forged or left to cool down before being reheated to forging temperature. The transfer from the furnace to the forging press is not normally carried out in a protected atmosphere, but the time interval is designed to be very short so that oxygen penetration is kept at an acceptable minimum level.

The overall objective of the process of powder forging is to impart to the components properties that are comparable to those of the conventional forgings.

Powder forging can compete with conventional forging mainly for two reasons:

(1) It reduces the amount of machining compared with conventional forging.

(2) It enables very good surface finish and dimensional accuracy to be obtained when compared with conventional forging.

The conventional methods for manufacturing preforms from solid metal have at least two disadvantages; the expense, 1. in labour, 2. in wastage of material would be considerable and would increase rapidly with the complexity of preform shape.

However, the use of powder preforms has considerable potential; compacts can be formed to precise dimensions, good surface finish, close weight control and preselected distribution of density in complex shapes. With these features, the powder forging technique provides a flashless forging of precision quality and very high density.

Other advantages are possible by use of the powder forging route, namely material savings, improved detail and surface, high preduction rates, material rationalisation (to gain certain specific properties), partial elimination of machining, and structural homogeneity.

Another advantageous aspect of powder forging is the production of unforgeable and heat resistant alloys which are not possible to forge by conventional methods. (1, 2, 3).

There are some quite recent reports, pointing out the fact that powder forging is now becoming a viable technique even in terms of energy and investment costs in comparison to conventional forging. (4, 5, 6).

For comparison purposes some numerical data comparing powder and conventional forging are given below.

Overall cost saving of about 30% obtained by use of powder forging. (7, 8). The material utilisation during the production of parts by conventional forging or casting is often very poor because of the flash formation on the former, and runners and risers on the latter. In addition, machining losses may result in the weight of the finished part being only 30%-50% of that of the raw material used. If powder forging is used to produce the semifinished article, then material utilisation is usually in excess of 90% (7, 8).

As a result of extensive laboratory, and field trials (9, 10, 11, 12) it has been demonstrated conclusively that powder forged components have equivalent performance levels to conventionally made parts in nearly all cases, and in some cases better results can be obtained. An example shows that powder forged connecting rods for Porsche cars have superior fatigue performance when compared wity the equivalent drop forged parts (7, 8).

Nowadays, auto transmission parts and roller bearing cups and races are being made by powder forging. Since 1972 Volvo has made over 200.000 forged P/M steel hexagenal nuts for its trucks (13).

Powder preforms and hence the finished forgings have very good weight control. This is an important factor in some applications, particularly in the automobile industry. It is reported that weight control is possible within the acceptable limits of ± 0.75 ÷ 0.50 % (7, 8).

At present stage of development, powder forging offers considerable energy savings over conventional forgings and casting. (14). The energy used to produce a tonne of finished products by the powder forging route is 70% ÷ 90% of the energy for the conventional route, depending on the operations used to make bar or billet. Within the foreseeable future, such advances as larger powder plants and the advent of induction sintering could reduce the energy consumption to 30 - 40% of the current conventional forging figures.

In one of the comparative cost estimates of high quality metal powders, it has been argued that todays market prices could be reduced by 20% to 30% in comparison to the conventionally produced forged parts (6). This would necessitate improvement or change in scrap recycling, enlargement of the plant capacity and better utilization of existing equipment.

Furthermore there are reports that a few but important, technical developments would be unthinkable without the application of powder metallurgy. Examples are in gas turbine engines, space vehicles, and aviation technology. (13, 15). Typical parts are the high pressure compressor discs, compressor off shaft, forward inner seal, high pressure turbine forward shaft, forward outer seal, and the high pressure turbine disc.

Since the carbon content of any one powder can be altered within quite wide limits, the properties of powder forgings can be varied considerably. This ability to vary the carbon content, also enables a wide range of heat treated properties to be obtained from one alloy composition. In fact, this is a unique feature of powder forging.

The versatility of the process also enables two or more types of metal powders to be used in a preform in such a way that the properties of each metal powder are put to best use. This has been shown by Haller (16) to be a feasible technique.

It has been recently reported (17) that duplex alloys produced by powder metallurgy were stronger and more ductile than those obtained by vacuum melting and casting. In that work duplex Fe-20 vol % Ni-C alloys were prepared by powder metallurgical techniques. They have better mechanical properties than those of the Fe-20 vol % Ni-C homogeneous alloys.

In general two layer sintered compacts are used for reasons of material properties, ease of production, design and economy. For example the first applies when a composite sintered component is required with two different materials possessing different properties as in the case of electrical contacts. Secondly, with respect to process engineering, a strongly joined interface between sintered composite materials may be decisive for the selection of a production route. Finally, as has been mentioned above, possible savings in material, energy and labour would be dominant factors in production.

The two-layer powder compaction technique offers the possibility of producing sintered compacts of high dimensional and mass accuracy. The different properties of both powders must be taken into account when selecting the pressing and sintering conditions.

There are various combinations for producing two layer powder preforms; among these, a combination of the two layer compacting technique with the sinter infiltration technique represents an economic method of producing single or two-layer preforms.

Another advantage of the use of two-layer powder preform is when using a layer of material which cannot be produced by melting and casting, for example consisting of metals, metal-metal oxide or metal-metalloid which are mutually insoluble in the liquid state. With these materials, customary cladding techniques are difficult and costly, and therefore two layer powder compaction would be necessary. It offers the possibility of producing two layer materials as sintered products of high dimensional and mass accuracy, with great strength at the interface.

Usually in producing two-layer powder components, two powder layers of different composition are filled one on top of the other and are compacted into a unique stable compact. The depth of intermesting between the two layers is dependent on the particle shape and particle size distribution of the two powders. (18).

The object of the present work is to show the feasibility and necessity of two part component compaction and forging with regard to material properties, pressing and sintering conditions.

CHAPTER TWO

HISTORY OF POWDER FORGING & LITERATURE SURVEY

2.1 Early Application of Powder Forging

The powder forging technique as a means of producing components, has had its applications at about 5000 B.C. during the time of ancient Egyptians. They heated iron oxide in charcoal fires, thus reducing the oxide to a spongy metallic iron. A coherent metallic mass was then obtained by hammering the porous metal in its hot condition. Finally usable shapes were obtained by simple forging operations. (19).

Modern developments of powder forging began about sixty years ago when Coolidge prepared tungsten by powder compaction, sintering and forging (swaging) before it was drawn into a wire (20).

Thirty five years ago researchers in the American Electro Metal Corporation forged tungsten alloy preforms on an experimental basis and this became the accepted method for production of rocket motor nozzles. (21).

Forging of encased or canned metal powders was succesfully applied several years ago to Beryllium and other specialty metals that are otherwise difficult or impossible to form (22).

Beryllium can only be obtained in powder or hot pressed billet form because a cast ingot of beryllium is almost unworkable. Because of the toxic nature of this element, and in order to reduce oxidation and provide workability canning the powder before forging is necessary (22, 23).

2.2 Hot Pressing Technique

Savarvold, in 1929 suggested the idea that both pressure and heat could be applied simultaneously to iron powders in the forming die. Although this hot pressing technique had some advantages, the presence of some practical difficulties required to be overcome as mentioned by Jones (24). Hot pressing eliminates furnace equipment as well as producing a compact to the desired final dimensions. The process also produces better physical properties when compared with cold pressed compacts. The time needed for heating up the powder in the die is the major disadvantage, as well as the oxidation problem of grain surfaces prior to compaction. Densities, obtained by hot pressing were nearly 100% of theoretical for iron, brass, copper and bronze.

However, a successful solution has been reported by Goetzel (19) which involves cold pressing, of powder into a preform before it is hot pressed, thus facilitating handling and also relieving oxidation problems.

Hot pressing of iron and steel powder preforms was also investigated by Koehring in 1942 (25). Using bar shaped preforms of iron powder + 0.13% C and heating to 1200^{o}C in an hydrogen atmosphere for one hour before forging He found mechanical properties to be similar to those of a steel of similar composition.

Wulf (26) obtained a tensile strength of 660 MN/m^2 and 16% elongation with iron powder plus 0.48% C, and 860 MN/m^2 tensile strength and 16% elongation with a simulated S.A.E. 4340 powder (an alloy of carbon, manganase, silicon, nickel, chromium and molybdenum).

2.3 Hot Forging of Iron Powder

Ishimaru (27) et al. investigated seven different types of iron powder in order to select the optimum ones for the powder forging process. In that work a compaction pressure of 400 MN/m^2, a sintering temperature of 1120^{o}C and sintering time of 30 min in a cracked ammonia atmosphere were kept unchanged for each powder type, and forging was carried out in a closed die at 900^{o}C, followed by annealing at the sintering temperature. The density of each compact was nearly 7.8 g/cm^3 after the forging operation. The only significant difference between the forged parts was in the elongation results. Hoeganeas MH.100.24 reduced iron powder gave an ultimate tensile strength of 290 MN/m^2 with an elongation of 27%, whereas Hoeganeas AHC100.29

atomized powder gave an ultimate tensile strength of 290 MN/m^2 with an elongation of 38% The lower elongation performance of the reduced powder was attributed to the residual oxide content.

Zapf (28) also reported that reduced powder gave poorer elongation results, the best values being obtained with an electrolytically produced powder.

2.4 Hot Forging of Prealloyed and Blended Powders

Huseby (29) reported that EMP 4600, a Ni-Mo prealloyed atomized powder proved to be ideal for powder forging preforms. A comparison between a blended and a prealloyed S.A.E. 4600 powder was made by Cundill (30). The major difference was the lack of hardenability in the blended alloys. This was attributed to segregation caused by insufficient diffusion of the alloying elements into the iron matrix.

2.5 Effect of Preform Density in Hot P/M Forging

Huseby (26) et. al used two different preform densities of 6.0 g/cm^3 and 7.0 g/cm^3 of EMP 4640 steel powder, sintered at 1120°C for 15 min. in cracked ammonia atmosphere and forged at 1120°C to full density. The preform density of 6.0 g/cm^3 needed a forging pressure of 1080 MN/m^2 to achieve full density whereas the preform of 7.0 g/cm^3 density needed a pressure of 770 MN/m^2 only.

Antes (31) also pointed out that the load necessary to achieve a required density increases with decreasing preform density.

Marx (32) et al investigated six different initial preform densities ranging from 5.37 g/cm^3 to 7.38 g/cm^3 of the same mass of MH.100 iron powder. These were sintered at 1100°C for 20 minutes in a cracked ammonia atmosphere and forged between flat dies (upset) at 1100°C to the same final height. It was found that the forged density increased slightly with increasing initial density.

2.6 Effect of Die Preheating in Hot P/M Forging

According to Zapf (28) the hot forging of iron-nickel powder preform alloys showed that if the dies were heated to 310°C prior to forging, the load required to attain a given density was reduced. In that work, forging temperatures, ranged from 500°C to 900°C. The billets of 6.0 g/cm^3 initial density were presintered in one series of tests at 885°C for two hours, and cooled down. They were preheated to the required forging temperature. In the other series of tests the billets heated straight up to the forging temperature without sintering forging was carried out in heated dies of 310°C at a maximum forging pressure of 400 MN/m^2 for both series of billets. It was found that the unsintered billets had a higher forged density when compared to the presintered ones. The process was a combination of compaction, heating to forging temperature, forging and final sintering.

The following table includes the results of sintering temperature effects after forging at 1000°C on the mechanical properties which were found by Zapf. That work also concluded that the optimum sintering temperature after forging was 1050°C.

Material	Sintering Temp. °C for 2 hr.	Tensile Strength MN/m^2	Elongation %	Hardness B.H.N.
Iron + 2% Ni	1000	427	26.9	109
Compacted	1050	372	43.0	106
at 800 MN/m^2	1100	389	39.8	117
Forged	1150	349	42.2	104
at 1000°C	1200	362	43.5	103
	1250	362	44.3	106
	1300	353	44.9	99

2.7 Effect of Sintering and Forging Temperature on the Forged Properties

Davies (33) et al used spongy iron powder MH.100.24 to investigate the effect of sintering and forging temperature on the forged properties. The initial density of all the billets was 6.79 g/cm^2. Sintering and forging were carried out at the same temperature, varying from 650°C to 1200°C sintering time was kept constant as 20 minutes. Upset forging on a Petro-Forge machine produced forged billets of same final density. All the upset forged billets forged below the 900°C range showed peripheral longitudinal cracks. These tests resulted in an optimum sintering and forging temperature of 1100°C, which was a compromise of overall properties. The work was also correlated with the results of other investigation on the forging of solid iron (34, 35).

2.8 Rapid Heating of the Preforms

In a report by Mocarski (36) et al. it was mentioned that rapid heating of the preform does not provide sufficient time for proper burn-off of the compaction lubricant, thus leading to internal rupture of the preform. Also reduction of the oxides inside the powder cannot be achieved effectively during induction sintering. They suggested the following sequence of operations to overcome these problems; furnace sinter the preform, then cool to room temperature, and finally reheat by induction furnace to forging temperature.

2.9 Induction Sintering in Hot P/M Forging

Vernia (37) also studied the effect of induction sintering on forged billets using seven types of commercial iron powder. A 5% hydrogen and nitrogen gas mixture was used as protective atmosphere. It was found that the carbon losses of the process were of the same order as in furnace sintering. Also the strength and density of the forged billets were essentially equal with or without a pre-sinter.

The transverse rupture strength of the atomized iron powder + 0.5% carbon forged slugs (forged at 915°C) were 1380 MN/m^2 and 1400 MN/m^2 in the case of induction sintering (1150°C - 60 sec.) and induction reheated (presintered at 1120°C - 40 min.) respectively.

Knapp (38) studied the effect of induction sintering without a protective atmosphere on the properties of the forged prealloyed powder of 4600 + 0.2 ÷ 0.3 % C. A comparison was also made between furnace heating and induction heating. It was concluded that the hardness, tensile strength and yield strength were comparable to wrought products for either furnace or induction heating. The highest ductility was obtained for wrought material, followed by the furnace and induction heating method respectively. The most significant difference was found in the Charpy impact test, where the furnace heated preform billets had only half the impact values of wrought 4620 bars, and the induction heated preforms half that of the furnace heated preform billets. It was believed that the results were due to residual oxide content.

2.10 Oxidation of Preforms in Hot P/M Forging

Oxidation of preforms prior to forging operation is an important problem. Anon (39) also mentioned this problem and suggested the idea that the problem could be overcome by rapid heating of the preforms using induction heating techniques, after cooling from the sintering stage.

The significance of total oxygen content on the fracture toughness of powder forged Ni-Mo steels was demonstrated by Pilliar et al. (40). Sufficiently high sintering temperatures to achieve low total oxygen either by oxide reduction or complex oxide vapour reactions have been shown to be important in producing powder forgings processing high fracture toughness values.

Bastian et. al. (41) also have examined the fracture characteristics of three different powder forged steels in terms of oxygen levels. Higher oxygen levels decreased the spacing of the oxide inclusions on the prior particle boundaries causing a decrease in the resistance to fracture. They stated that, the fracture resistance of a powder forged steel, measured by crack opening displacement testing, can be modified by heat treatment. Treatments decreasing the strength and hardness and increasing the work hardening capacity, increased the crack opening displacement values.

Crowson et.al (42) in their work on the 4600 modified prealloyed powder, concluded that both forged densities and final oxide contents were dominant factors in achieving acceptable mechanical and physical properties. They found that the oxide content of manganese is quite difficult to reduce below 2200^{O}F and can adversely effect the P/M steel forging properties. Sintering at higher temperatures (2300^{O}F and 2400^{O}F) over longer periods of time, can substantially reduce the amount of these oxides. They also found that the porosity prevalent in the sintered preform has to be collapsed in such a manner to avoid defects and cracks which could adversely affect the resultant forged properties.

It was mentioned by Neiderman (43) that to fully gain the beneficial effect of the chromium and manganese in water atomized steel powder, it is necessary to reduce the chromium and manganese oxides that form during the powder production. It was also suggested that using high temperature sintering (1260^{O}C or 1315^{O}C) it is possible to use a shorter sintering time of approximately 5 to 10 minutes in either a dissociated ammonia or nitrogen based atmosphere.

In an another report by Cook (44) concerning the hot forming of AISI 4027 steel powder, it was mentioned that the use of manganese as an alloying element in atomized steel powders has been limited because if it oxidizes, hardenability and notch toughness will be low - the former due to the depletion from solid solution, the latter to the number of crack-initiating brittle oxide inclusion. In that report also high temperature sintering (1205^{O}C and 1260^{O}C) was found to lower the oxygen content and increase the impact energy. The resultant improvements were due to the reduction of the number of brittle crack forming oxide inclusions and the increase of the amount of alloying elements in solution.

2.11 Effect of Processing Variables in Hot P/M Forging

Cull (45) has given some comparative values of the mechanical and metallurgical properties of powder forged steels with regard to the effect of processing variables. He mentioned the effect of graphite diffusion in the iron + 0.5% graphite preforms at 1100°C. The resulting mechanical propertie: which are given below, were after quenching from 850°C and tempering for 1 hr. at 500°C.

Time at Heat (Min)	UTS tonf/in^2	UTS h bar	Elongation %	R.A %	% Graphite Diffusion
8	25.0	38.6	20	25	10
10	29.2	42.0	23	35	25
19	39.5	45.5	25	40	95
39	39.2	45.1	25	35	95

Percentage graphite diffusion was estimated metallographically, the final carbon content being <0.5 % owing to some oxide reduction. It was noted that percentage oxide reduction approximated to the percent carbon diffusion. He also studied the effect of carbonyl nickel addition to the iron powder. The tests on 1% nickel added to iron powder samples of both normalized and quenched from 900^{O}C, showed increasingly improved mechanical properties as the diffusion time at 1125^{O}C increased from 5 min to 90 min From a series of powder forging tests he concluded that a very good approximation to wrought material can be acheived, particularly when powder forged Fe-C alloys contain little or no manganese and residual elements. After normalising mild and carbon steel powders + 0.5 ÷ 1 % added graphite forgings were found to have mechanical properties equal to or better than the equivalent wrought material. It was one of the early reports which was included in the same work of Cull; powder forged connecting rods and automotive gears being subjected to very severe engine-testing program, gave similar results to their conventional wrought counterparts.

Fishchmeister (46) et. al. mentioned the diffusion characteristics of alloy powders for powder forging which were made with plain iron powder plus admixed alloying elements. They pointed out the fact that if the alloying additions needed for hardenability are made in the form of a low-melting alloy powder, diffusion times can be very much reduced. A condition is that the molten prealloy wets the iron particles, reducing the diffusion distance to the order of one particle radius. They also found that it would be desirable for the alloy, to be penetrated quickly along the grain boundaries of the iron, further reducing the diffusion distance. Low melting alloys of magnanese with copper were examples which satisfied the conditions. In that work it was reported that since only small amounts of low-melting alloy powders are needed, solid-liquid alloyed plain iron powders appear to offer great flexibility in alloy composition at a cost substantially below that of conventional prealloyed powders. A particular advantage of combining solid-liquid alloying with forging is that the voids left in the place of the low-melting alloy particles upon melting will be filled during plastic deformation of the preform.

2.12 Effect of Iron Powder Contamination in Hot P/M Forging

Steed (47) investigated the effects of iron-powder contamination on the properites of powder forged low alloy steel. He found that the presence of small amounts of iron powder contamination (< 5 %) has little effect upon the tensile properties of the alloys examined. But the presence of 20 % iron powder contamination in low-alloy through hardening steels reduces the tensile strength attainable. Fatigue properties were also significantly reduced in line with the reduction in elastic limit, the ratio fatigue endurance/elastic limit being constant. If a carburizing-grade material is used, iron powder contamination must be reduced to an absolute minimum to avoid effects upon the hardenability and mechanical properties of the material and to prevent soft spots being present in case-hardened surfaces.

2.13 Effect of Powder Materials on the Properties of Hot P/M Forging

Brown (48) has examined three powder forging carburizing steels with realistic composition ranges. It was illustrated in terms of the mechanical properties, that carbon content can be controlled in processing to give higher or lower strength as appropriate to the stress situation. The powder forged steels showed ductility values higher than those taken with transverse direction in the equivalent wrought steels, but lower than those taken in the longitudinal direction.

Hoffmann et. al (49) have investigated a number of hot formed P/M steels with regard to their alloying elements, individual mechanical properties, fracture analysis and alloying costs. That investigation included alloying systems with elements having a low as well as a high affinity towards oxygen. A new alloying technique of manganese, chromium and vanadium master alloys were combined in complex carbide phases resulting in H/F steels with high mechanical properties and minimum oxygen content.

2.14 Machinability of Hot P/M Forging

Since many sintered alloy steels have a heterogeneous hardness distribution after forging, some subsequent heat treatment would be necessary to obtain a beneficial machinable structure. A research work at AMAX (50) showed that sintered iron compacts containing 0.1 % tellurium improved machinability by 60 %.

Taubenblat et. al (51) have mentioned in their work that forging of Fe - 0.5 % Cu - 0.1 % Te compacts could be accomplished without cracking using standard forging techniques. For various carbon levels all forged pieces showed an appreciable improvement in machinability as compared to specimens without tellurium.

Fletcher also studied the machinability aspect of P/M forgings (52). He suggested that machinability can be improved by microstructural control to obtain coarse-grained proeutectoid ferrite and a large volume fraction of fine pearlite. By furnace cooling on intercritical hold period prior to cooling was recommended for improving the machinability of 40F2, 40F3, 40F4, and 46F2 P/M forgings. The heat treatment consisted of furnace cooling of P/M forgings from their austenititing temperature of 1350^{O}F (732^{O}C), holding for 30 min. to 2 hours, and air cooling.

2.15 Some Theoretical Equations in Hot P/M Forging

Griffiths et al (53) have derived some theoretical equations representing the compatibility behaviour of axisymmetric, plane strain and uniaxial compression conditions during powder forging. They concluded that frictionless axisymmetric and unlubricated plane-strain compressions produce the least and greatest rates of densification, respectively during the upsetting stage of the powder forging process. The actual conditions encountered in practice would lie somewhere between these two extremes. They pointed out that compatability equations could be used for predicting the geometrical change of P/M preforms during forging as well as for the purpose of summarizing and comparing experimental observations.

Leheup (54) et al. have investigated the elastic behaviour of P/M high density products by means of N.D.T techniques. They found that the experimental elastic modulus data were lower than those preducted by the theoretical models which assume spherical pores in continuous homogeneous matrixes. Possible reasons for this discrepancy were attributed to:

a) The pores in real materials were not spherical and have higher stresses associated with them.

b) The matrix contains defects which may have both higher stresses and higher strains associated with them. The elastic properties of an as sintered sample were lower than those of a powder-forged sample of similar density. It seemed likely that the hot working process lessens the effect of microstructural defects, particularly by the consolidation of boundary bonds and oxidation of pore walls.

2.16 Flow Stress of P/M Hot Forging

Donachie (55) et. al. have studied the effect of composition, temperature and crystal structure on the flow stress of P/M forged preforms. They concluded that

a) the flow stress of P/M forging preforms is determined principally by: 1) phase balance; 2) solid solution strengthening. Ferritic structures have lower flow stress than austenitic structures in the temperature range of interest. Molybdenum contributes strongly to solid solution strengthening.

b) The phase balance at which minimum flow stress will occur is dependent upon: 1) the flow stress of each phase; 2) the slope of the flow stress temperature curve for each phase, and 3. The rate of phase change with temperature.

c) The use of low flow stress alloys in P/M hot forgings may increase die life significantly without decreasing the density of the forged part.

2.17 Effect of Deformation Mode on the Properties of P/M Forging

Antes et. al (56) investigated the effect of the amount of deformation on the mechanical properties of that forged powder preforms 6.5 g/cm^3 initial density preforms of Ni-Mo steel plus 0.25 % C were sintered for one hour in a cracked ommonia atmosphere at 1120°C and then hot formed to solid density by either upset forging at 980°C or by re-pressing at 1040°C, followed by quenching and tempering. It was concluded that upsetting gave better tensile and impact properties due to lateral flow of upset samples, whilst the tensile ductility was the same for both methods. Antes tabulated his results as shown in the following table:

Mechanical Properties	Upsetting	Re-pressing
UTS (MN/m^2)	820	710
Yield strength (MN/m^2)	730	600
R.A. %	35	34
Elongation %	21	22
Impact (C.V.N.) (joules)	35	28
Hardness (R.C)	26	26

Fischmeister (57) et. al. studied the deformation and densification behaviour of powder preforms in hot forging. They stated that, 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 and hence lateral flow is almost absent. It was observed to be a beneficial effect in the sense of forming technology. They also mentioned that 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 compactability of hot preforms, as indicated by the pressure-density or strain-density relationships. One other conclusion which they drew from this work was that the process of densification is basically different from that encountered in the cold compaction of loose powders with regard to both pressure requirements and density distribution. In contrast to compacts made from loose powders, forged specimens have a maximum density in the centre owing to the high pressure and strain to which this region is subjected.

2.18 Preform Design in P/M Hot Forging

In one of the earlier works of preform design Griffiths et.al (58) introduced some guidelines to assist in designing preforms for sinter forging and to attempt to minimise the 'trail and error' aproach. They were also in favour of producing preforms with close tolerances to the shape originally evolved by calculation. The advantages gained would have been:

1. A denser central region, caused by prevention of lateral displacement of metal from this region.

2. Reduced peripheral porosity, since use of larger diameters results in less lateral displacement and hence less chilling.

Downey (59) et. al also have stressed the same basic points which were mentioned by above-mentioned authors about the design of preforms for powder forging. They verified the fact that utilizing a preform stage very close to that of the final part (hot re-pressing) minimises the amount of plastic deformation, and essentially eliminates the occurrence of cracking. It was concluded that the correct preform shape and dimensions can be determined from consideration of fracture during the flow, final part properties and die wear.

The influence of preform shape on material flow and residual porosity was studied by Bockstiegel (60) et. al The hot forging of powder compacts into items of complicated shape requires proper preform design. In particular, in cases where items having sections of different lengths in the forging direction are to be produced in tools with individed punches, the risk of cracking overlapping or residual porosity in one or several of the sections is great. In that work some results were presented from hot forming experiments with a variety of preform shapes. They reported that in the design of a preform some important aspects of powder preforms must be taken into account. In contrast to well established principles of plastic deformation of solid metals, the preform volume is not constant but decrease gradually during deformation. The tensile strength is also considerably lower for a porous material than for a solid material of comparable kind. They pointed out that in many cases, porous preforms may have to be manufactured in rigid dies on fast running powder compacting presses. Consequently they must be desinged to match the particular requirements of conventional powder compaction with corresponding limitations of shape. They also found some results from their model experiments about the material flow and densification as follows: Lateral material flow begins to take place at laterally unsupported sections of the preform long before full densification is achieved in these sections. Lateral material flow becomes more intensive in these sections as densification proceeds. Densification begins to take place, not only in regions of maximum compressive stress, but also along planes of maximum shear stress. Cracks do not necessarily occur in unsupported surfaces of the preforms. Although, from a technical as well as economical point of view, preforms should have the simplest possible shape, a successful forging operation may require a more complicated shape in many cases. In a preform design the optimum compromise between these two adverse requirements will have to be attempted.

As a result of recent developments of preform design in hot P/M forging there is a tendency to design preforms with computerized techniques. Pillary et. al (61) studied a computer-aided preform design technique that specifies preform shape and dimension such that, defects are avoided and full density is achieved. This method is said to eliminate most of the trial and error labour involved in finding an acceptable preform design for each type of component.

2.19 Effect of Residual Porosity in Hot P/M Forging

Kaufman (62) et. al. reported that low alloy steel powder forgings, with densities varying from 95% of theoretical to full theoretical, gave poor tensile properties. The forgings were tested in tension to study the effects of residual porosity on mechanical properties. Tensile strength, yield strength and ductility were all found to fall off sharply in their density range lower yield points were observed only for samples with porosity concentration less than 3 %. They found that the appearance of fracture surfaces underwent a drastic change, below about 0.5 % residual porosity. They added that mechanical properties were a function of final pore shape but not of initial powder particle size prior to sintering.

In another work by Kaufman (63) about the role of pore size in the ultimate densification of P/M forgings. Some detrimental effects of residual porosity on some engineering properties were disclosed. Because of their surface energy, pores are resistant to deformation as their curvature increases. They estimated that the smallest pores that can be closed using existing types of equipment were of the order of 5 μm dia. The creation of a second generation porosity which is resistant to further deformation was shown to prevent the attainment of full densification. Shearing type loads were found to be more efficient than compression in ultimate densification. As a result it was concluded that it is not theoretically possible to eliminate all residual porosity by mechanical deformation because of the secondary effect of inhomogeneous deformation.

2.20 Metal Flow and Densification with Regard to Preform Size and Pressing Properties in P/M Hot Forging

In their work on metal flow and densification during the die forging of powder preforms, Guest et. al. (64) have investigated the optimum values of Do/D (die diameter/initial preform diameter) in terms of forging load, energy and density. Their results were in general agreement with those reported earlier, that in the earlier stages of forging the deformation of the preform is largely in the pressing direction and the rate of densification is then at its maximum. In the intermediate stages the metal flow increasingly takes place in the radial direction. The next stage of densification pattern is similar to that found in simple upsetting experiments, and barrelling takes place until the material reaches the die wall. They found that the formation of radial cracks is likely when the diameter of the preform was increased during forging above a critical ratio, which in the authors experience was $\sim$ 1.40. They concluded that, although the use of large deformation is advantageous in healing porosity, the danger of cracking low-density preforms while forging outweighs the advantage of easy flow of low loads. The most effective way of using powder metallurgy, they suggested, it to produce a preform having intricate detail and close tolerances, to which a simple hot coining operation can be applied in forging.

Bosse (65) et. al. also have studied the hot pressing properties of iron powder and preforms. They concluded that densification of iron powders and preforms required less forging loads at about $850^{o}C$ in alpha phase than at higher temperatures in gamma phase. The lowest densities are obtained when hot pressing in the α-γ transformation range.

A minimum density of finished pieces was obtained for preforms sintered at about $1000^{o}C$.

In an investigation into the hot forging of powder preforms of 4600 (Ni-Mo) alloy steel powder. Davies (66) et. al. found that for 4600 powder + 0.5 % carbon the optimum sintering and forging temperature was $1100^{o}C$. A total sintering time of 15 min was recommended. The effect of the initial preform density on the mechanical properties was negligible for preforms forged to the same final density. Mechanical properties, with the exception of ductility, were improved by increasing the radial flow during forging.

P/M preforms of Fe-0.53 C with 6.5 Mg/m^3 density were hot forged in five types of forging modes and compared with hot forged bar stock by Watanabe et. al. (67). Stage change during forging at the P/M preforms was qualitatively similar to that of bar stock, except that the former showed tendency to crack at a lower strain. The influence of the forging made on densification was most pronounced up to about 0.8 GPa mean punch pressure; above 0.8 GPa the mean density of all specimens in closed die forging reached 99.5 % of theoretical. Density distribution in P/M forged specimens were found to be non-uniform during the early stages of the forging process, but became uniform at the final stage of die filling.

In a work concerning porosity and densification in forged ferrous powders, Majundar et al. (68) found that surface cracking occurred when the available cavity was not filled; when the inserts impeded material flow some surface crack closure occurred and surface oxide and porosity were trapped as sub-surface defects. Six different methods of iron powder preform forging were investigated by Skelly (69). Among various combinations of sintering, cooling, cold forging, resintering and hot forging, he concluded that combining the sintering and forging preheat steps in one operation and then forging was the most promising procedure with respect to economy, simplicity and properties comparable to those of sintered and hot forged material.

2.21 Forces in P/M Hot Forging

Forces encountered during the forging of iron powder preforms was studied by Huppman (70). He found that in forging of iron powder preforms very low stresses are required if deformation takes place just below the α-γ transformation temperature.

2.22 Fatigue Response of P/M Hot Forging

Ferguson (71) et al. investigated the fatigue of iron base P/M forgings. The data obtained from the tests were examined in light of empirical design relationships for the prediction of fatigue response. They concluded that the axial fatigue resistance of fully dense 4620 P/M forgings in both the normalized and tempered martensite condition is enhanced by lateral flow during densification. There was no apparent effect of lateral flow on fatigue behaviour in rotating bend fatigue. Fatigue strengths in the P/M forged material were comparable to those of 4620 forged bar stock they found that the structural integrity and hence fatigue resistance of a prototype P/M part can be enhanced by incorporating lateral flow of material during densification.

The effect of surface treatments (carburizing, nitriding, carbonitriding etc.) on fatigue of powder forged steels was studied by Usmani et. al (72). They found that in fatigued surface treated steels cracks were initiated at inclusions and at pores located at interfaces between core and case. Cracks were also initiated by link-up of fine pores below the compound layer.

2.23 Cost Analysis of P/M Hot Forging.

A number of investigators have studied on the economical aspects and cost analysis of P/M hot forging.

In addition to the authors in ref. (1, 2, 3 and 8) Nelderman (43) and Hoffman et. al (49) have introduced some results of the evaluation of cost. An evaluation of mechanical properties of lower cost powder forged 4100 and 1500 type alloy steels by Neiderman (43) has shown that 4100 and 1500 type powder can result in approximately a \$ 0.10 per pound cost savings when compared to Ancersteel 4600 V (1.8 % Ni, 0.50 % Mo, 0.25 % Mn).

In the investigation of Hoffman et. al (49) certain relations between individual mechanical properties and reduction of area and optimum composition of H/F steels regarding mechanical properties, fracture mechanism and alloying costs were established. They found an optimum alloy of (0.2 % MN, 0.2 % V, 0.2 % Mo, 0.6 % C) having UTS of 900 to 1000 N/mm^2, endurance limit of 340 N/mm^2 and impact energy of 13 joules at only 0.33 DM/kg (7 cents/lb) cost.

2.24 Composite Structures

Composite structures of two or more different materials have elicited much interest, because of their generally improved mechanical and physical properties. They may also be designed to optimize mechanical property to weight ratios for specific applications.

Composites can be of various types. In conventionally produced types, some of them may consist of alternate laminar (i.e. sheet or foil) of different materials.

Coated materials may also be considered as laminar composites.

2.24.1 Cladding in P/M Applications.

Cladding is another way of producing composite materials. The clad sheets are produced by using either a mild steel or stainless steel wrougth sheet on which the iron powder is compacted. They are then sintered together. In the work of Nakogama et. al (73) a thin layer of iron powder (about 1mm) was spread uniformly on a cleaned wrought sheet, and sintered together at 1175°C. As a result, the iron powder layer adhered to the sheet and provided a rough and active surface. On this surface the iron powder after spreading uniformly was compacted. A good adhesion between the wrought sheet and the iron layer was achieved.

2.24.2 P/M Composite Parts Produced by Liquid Phase Sintering

Powdered composites of two or more different material have been produced for a wide variety of industrial applications. The process was foreseen in 1916 by G.L. Gebauer (74) of Cleveland, Ohio, who applied for patent in which he proposed a process whereby a mixture of finely powdered metals (or in colloidal form) could be molded into the desired size and shape under pressure. The compacted material was then subjected to a predetermined temperature which was above the temperature of the lowest melting point constituent, which when molten, penetrated the pore volume of the compact. Freezing this constituent then produced excellent banding of the solid phase. The product could than be pressed, stamped, rolled or worked into various shapes.

Gebauer (75) also used a combination of tungsten and platinum for production of contact points with tungsten as the powdered "skeleton" and platinum as the brazing metal.

An example of gear production by infillration in American Electro Metal Corp (75) consisted a Cu-base slug in the bore of the gear and infiltration by melting of the Cu alloy. During infiltration, controlled atmoshpheres were used so as to prevent any surface oxidation which would impede close surface contact. The infiltration process enabled the production of parts with complicated shapes and involving variations in cross-section such as doughnut-shaped items and gears.

By means of liquid phase sintering, silver-graphite contact materials also can be produced from powders. Graphite with its good lubricating properties provides sliding contacts with very low contact resistance. As a make-and-break contact silver/graphite with approximately 5 wt% graphite is resistant to welding both in switching to an existing short circuit and for closed contacts: (76).

A number of works on two-layer powder compaction and sinter infiltration for electrical contacts were carried out by Schreiner (77, 18). In his experiments two powder layers were compressed together, forming a two-layer compact. The powders, one filled on top of the other were compacted by or without prepressing of the first layer. A smaller intermeshing depth was obtained when a precompressed powder layer was simultaneously compressed with a second layer. After sintering and repressing a two-layer sintered, structural part was obtained. The infiltration was produced by dipping the porous skeleton into the liquified material. By capillary action, the liquid metal penetrated into the pore volume. The infiltrated material was heated to a temperature above its melting point, which resulted a practically pore-free composite material. Composite compacts consisted of tungsten powder as the contact layer and silver, copper and little Ni powder as the supporting layer. Through an overdose of infiltrated material, a second layer was formed on the textured reverse side of the contact, resulting in faultless, joining without costly extra procedures.

2.24.3 Permanent Magnets

Composite forms of sintered powder materials have found applications in permanent magnets. As was explained in Frank's report (78) on permanent magnets for certain applications small magnet assemblies, consisting of a magnet attached to soft iron pole pieces, are required.

In some cases composite parts consisting of separate layers of magnetic alloy and soft iron powder can be pressed and sintered.

2.24.4 Composite P/M Parts in Electrical Applications

One of the first applications of powder metallurgy in the electrical industry was for contacts. (79, 80). Powder metallurgy was used because compositions of most of the necessary materials are such that they could not be made by conventional methods of melting, casting, rolling, drawing etc.

Selection of the grades of contact material depends upon the application. For example if a contact is to be used as an arcing tip, absolute resistance to welding or sticking is required. If the material is to be used as a current-carrying contact, and is not required to make or break the circuit, factors such as current-carrying capacity and low contact resistance are indicated. In most cases however, the contact must combine both of these qualities; such that the material must be able not only to carry the necessary electrical load but also to interrupt or break the circuit when desired. Only by means of powder metallurgy is it possible, to produce contacts that combine high electrical load-carrying properties with ability to interrupt the electrical circuit when short-circuit conditions arise. (79)

Composite materials i.e. those from the systems of silver-metaloxide and silver-nonmetal, have favourable engineering properties. However they show an insufficient wettability when compared with liquid metals. Since hard soldering or welding these materials with the support metal introduces many difficulties, double-layer contacts with a well solderable or weldable second layer are obtained by means of powder metallurgy methods (80).

The metallic bond between powder particles is significantly improved by a shaping operation after sintering. This applies particularly to hot shaping such as hot rolling or extrusion. Apart from the advantage of quality, these processes also offer the possibility of the manufacture of ductile, semifinished articles. In this way it is possible to produce economically small mass production parts from wire strip and sheet using conventional processes. These semi-finished materials also serve as laminating materials and for the automotic welding to support metals.

2.24.5 P/M Composite Metal Plastic Parts:

Sintered composite parts can also be made up of metal and plastic in a variety of forms. In many cases, only a combination of these two materials will produce the special range of desired properties. Sintered metal/ plastic composite parts have advantages over their conventional counterparts which are made by binding, attaching adhesion etc, methods, in such a way that the combination of the two materials is facilitated by mutual penetration of the porous sintered metal and the plastic. Therefore problems such as thermal stresses due to different coefficients of expansion and moduli of elasticity, mechanical stress peaks, extra surface treatment, difficulty in maintanance of dimensional tolerances (which would other wise arise in the use of conventional composite materials) can be avoided (81)

2.24.6 P/M Composite Parts in Medical Applications

Recent developments in P/M are being extended into many areas which in turn P/M has found successful applications in medical implant operations.

Dustoor et. al. (82) have reported a medical use of P/M. The replacement of the femoral head of the hip joint with an artificial prosthesis of P/M composite materials have been elucidated in their work. In their experiments the ball and stem of the hips were isoslaticaly compacted at 45000 lbf/in^2 pressure, using a 325 mesh stainless steel powder in the ball and a relatively coarser (as-received 100/325 grade) mix in the stem. A gradual transition was made from the fine to the coarse powder in the neck region, by using various blends of the two powders.

CHAPTER THREE

BASIC PRINCIPLES OF SINTER FORGING

As it was mentioned in the Chapter (1), sinter forging or (powder forging) is a combination of three distinct processes; preparation of powder preforms, (2) sintering of the preforms in a controlled atmosphere furnace and (3) hot forging of the preforms after sintering.

In this chapter a brief description of each of the above mentioned processes will be given.

3.1 Preparation of Powder Preforms

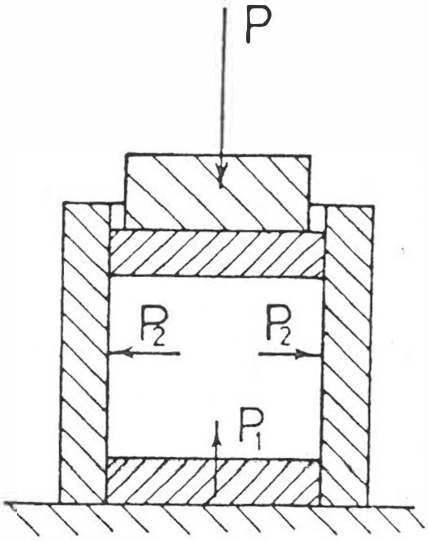
3.1.1 Compaction of Metal Powders

Powder compaction is the very basis of the entire powder forging process. A more detailed coverage of powder compaction can be found in any standard powder metallurgy textbook. (19,20).

Although there are some special purpose methods which do not rely on the application of pressure for powder compaction, the process itself is mostly involved with pressure on the metal powders. The pressure is usually applied by either (1) unidirectional pressing or (2) isostatic pressing. The three main methods of unidirectional pressing are ; (a) single action pressing, (b) double action

$$P_2 = X.P$$

$$P_1 = K.P$$



Where

P   Pressure applied from top

P_1 pressure applied to bottom

P_2 Side pressure applied to wall

X   Compaction conditions factor

K   Friction conditions factor

SOME VALUES FOR (X)

Type of metal powder	Powder pressed to density		
	40 %	60 %	80%
TUNGSTEN	0·08	0·12	0·16
IRON	0·16	0·23	0·31
COPPER	0·22	0·32	0·43

FIG. 3.1   SIDE PRESSURE IN UNIDIRECTIONAL PRESSURE COMPACTION.

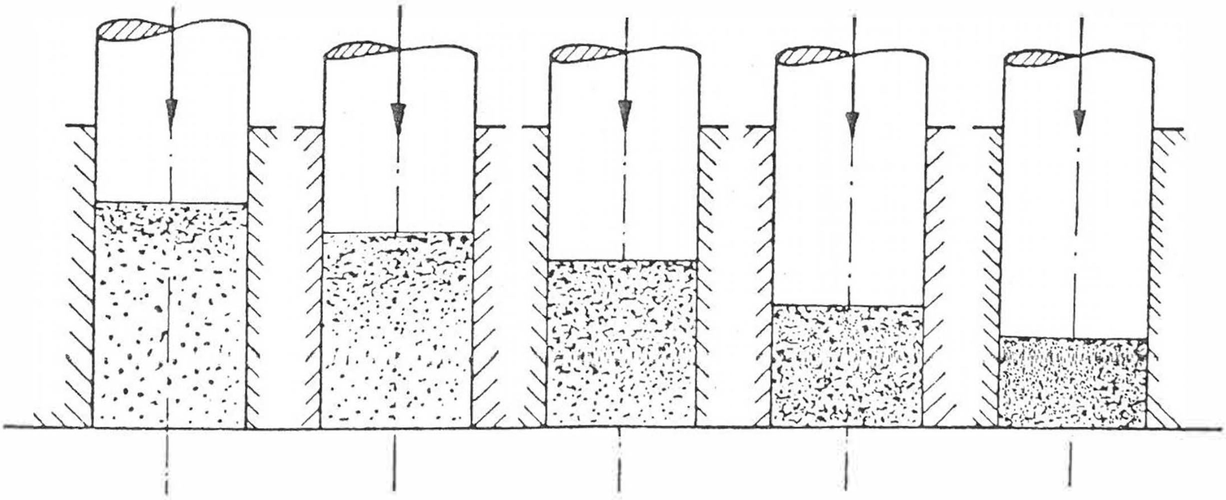


FIG. 3.2   SINGLE ACTION PRESSING.

pressing, and (c) double action pressing with moving die body. Figs. 3.1 and 3.2 show the single action technique in which the die and lower punch are stationary through out the process. Because of the frictional losses there are always differences between the applied pressure at the top and the transferred pressure to the lower punch. This pressure difference results in a non-uniform density distribution of the powder compact but as far as the final properties are concerned, this deficieny is offset by the subsequent hot forging operation which produces near theoretical density. Double action pressing, which is illustrated in fig. 3.3, is carried out by the simultaneous movement of upper and lower punches and this provides a more uniform density distribution in the compact. The floating die compaction method which is shown in fig. 3.4 is a somewhat modified type of double action pressing. In this method the die itself is supported on light springs. As the punch enters the die and begins to compact the powder, it sets up a frictional force on the die wall which overcomes the spring reaction, thus moving the die downwards. Since the spring forces are only just strong enough to withstand the weight of the die, the process is almost identical to the double action technique shown in fig. 3.3. Thus, the floating die method provides the advantage of double action pressing by using only a single action press.

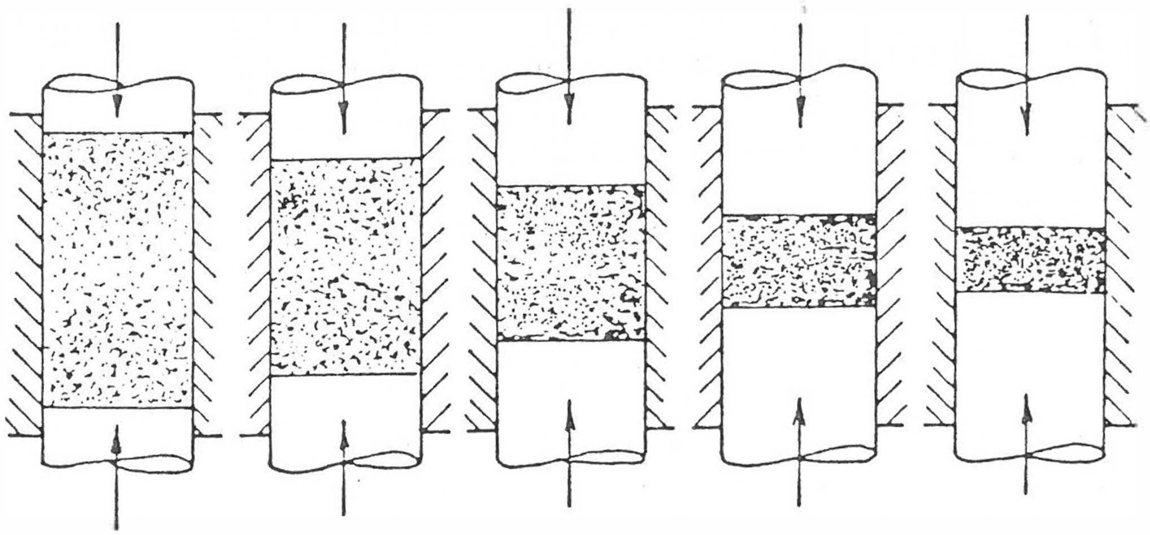


FIG. 3.3 DOUBLE ACTION PRESSING.

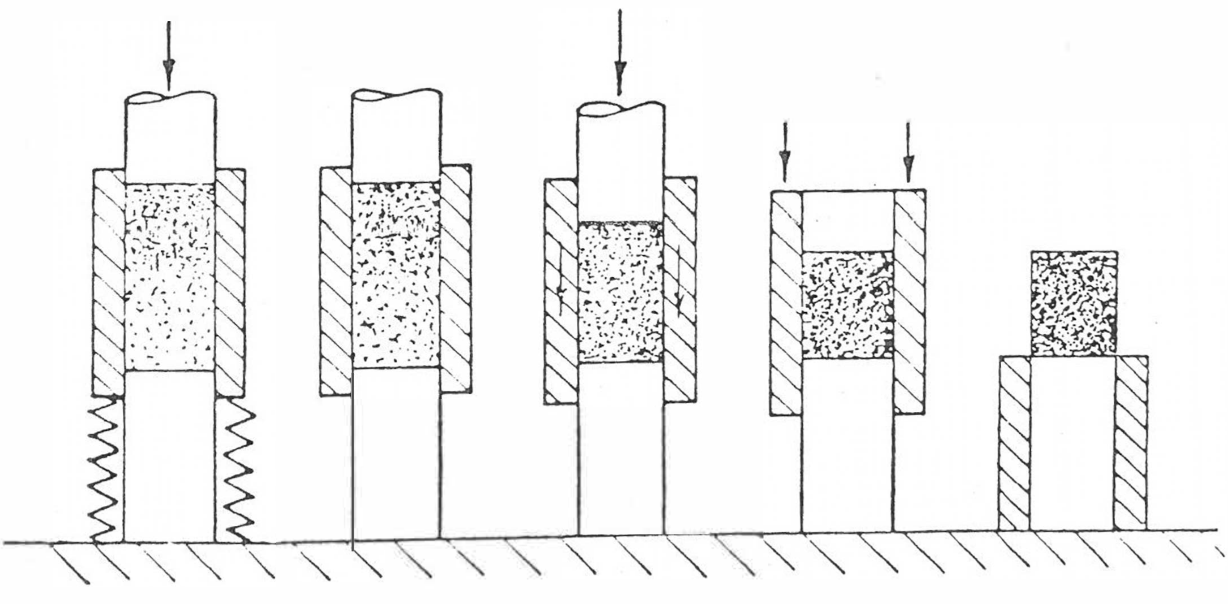
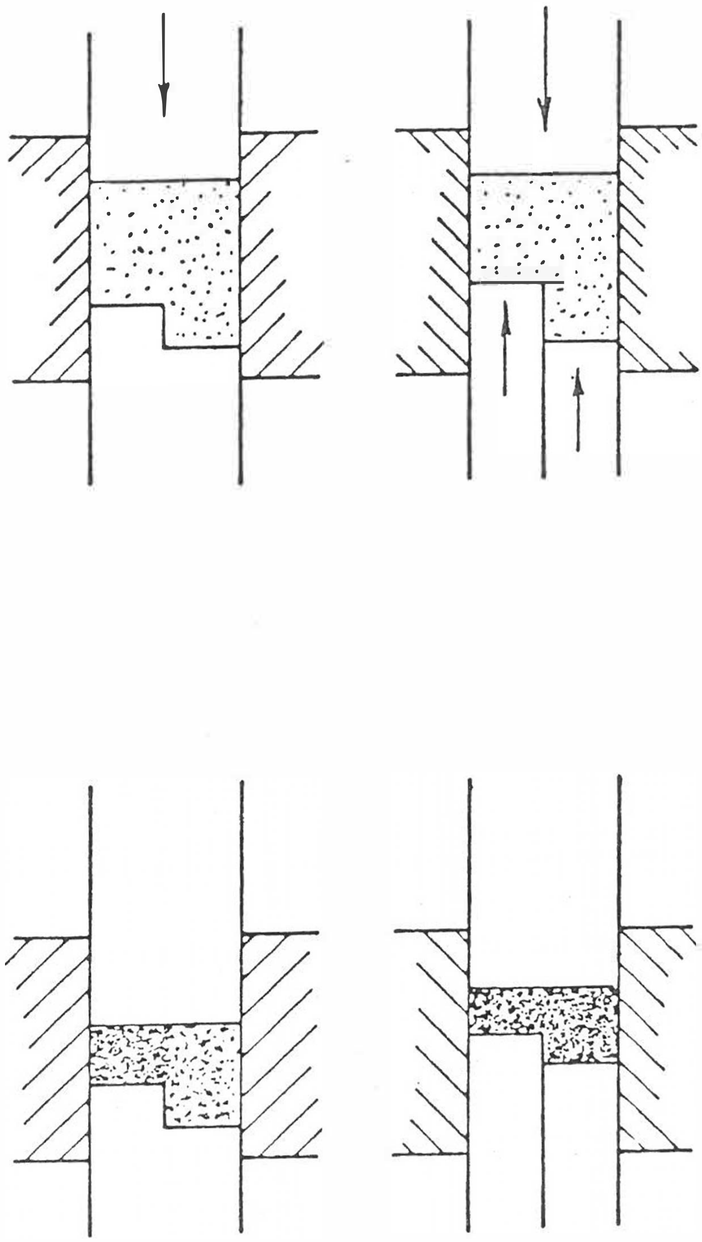


FIG. 3.4 DOUBLE ACTION PRESSING WITH MOVING DIE BODY.



(a) (b)

FIG. 3.5 DENSITY VARIATIONS IN MULTILEVEL COMPACTS.

The required compact in most cases will not be a simple cylindrical shape, but rather a multi-level part. If a simple contoured lower punch is used to press a multilevel compact, the differences of pressure distribution results in different densities in the various sections. The shorter length will be compacted first to a high density thus preventing an equivalent densification of the longer length as shown in Fig. 3.5a. To overcome this problem, synchronised split dies are used so that the density remains equal throughout the multilevel component during the entire compaction operation. A schematic view is given in fig. |3.5 (a, b)|.

Isostatic compaction is the main alternative to the conventional unidirectional compaction technique. This was developed in order to overcome the problems of density variation and the barrier of size and shape in conventional presses. The technique consists of filling a reversible mould, such as a rubber bag with metal powder and placing the powder filled bag into a fluid chamber. Hydrostatic pressure is applied to the fluid chamber, including the sealed rubber bag. If the sealed bag is a separate part of the fluid chamber, the system is called 'wet bag'; if the bag is attached to the pressing chamber, it is called 'dry bag'. The obvious advantages of the isostatic compaction process are; (1) the required compaction load is less than that required for conventional die pressing for the same green density, (2) less expensive tooling and equipment, (3) improved density distribution when compared with conventional unidirectional compaction and, (4) the capability to compact different sizes and shapes which are considered to be difficult or impossible by conventional compaction.

The main variables of the powder compaction process are; (a) method of compaction, (b) compaction pressure, (c) compaction rate, (d) powder grain size, and (e) method of lubrication. Because of die limitations conventional mechanical compaction machines very rarely enable parts to be produced whose density is more than 80% of theoretical, whereas high speed compaction presses such as Petro-Forge, can easily produce compact densities of 90% or more. Details of the high speed compaction technique is given in ref (83). Generally with the use of high speed compaction, properties of the compacts are marginally better than the conventionally produced compacts at the equivalent density.

The grain size of the powder is an important variable which effects the final properties. With small grain sizes, the surface area for contact is increased and a compact of greater strength is obtained. Increased contact area causes a greater amount of frictional energy which is largely converted into heat, thus providing increased plasticity in the grains.

The reasons for the use of lubricants in compaction are to obtain a more uniform pressure distribution during compaction and to ease the ejection of the part from the die after compaction. The compaction lubricants are either applied as die wall lubrication or mixed with metal powders.

'Acrawax' type mixed lubricants are burnt off completely. Detailes of the 'Acrawax' lubricants are given in ref. (84). However, some other mixed lubricants are not burnt off completely and leave some deposit.

3.1.2 The Theories of Powder Compaction

The mechanisms of interparticle bonding have been explained by number of theories one of these theories is the 'interatomic attraction'. Atomic forces have an important role in the powder compaction process. The degree of bonding between two contacting crystals is dependent on the mobility of the atoms and their ability to rearrange themselves in the surface layers. The adhesive bond between atoms of two different surfaces is not identical with, but is related to the cohesive bond between atoms in a lattice, and this in turn is not necessarily identical with, but is related to the forces of chemical affinity (24). In general the mobility of the atoms and their tendency to change places are related to the temperature and pressure. The higher are these factors the more likely it is that greater bonding will be achieved. When particle surfaces are in actual contact and the atoms of the surfaces are within interatomic distances of each other, the forces would be of an interatomic nature and magnitude, in which attractions are exerted at distances slightly greater than the normal interatomic spacing and strong repulsions at less. (24). As it is seen from the fig. 3.6, two approaching metal surfaces are initially

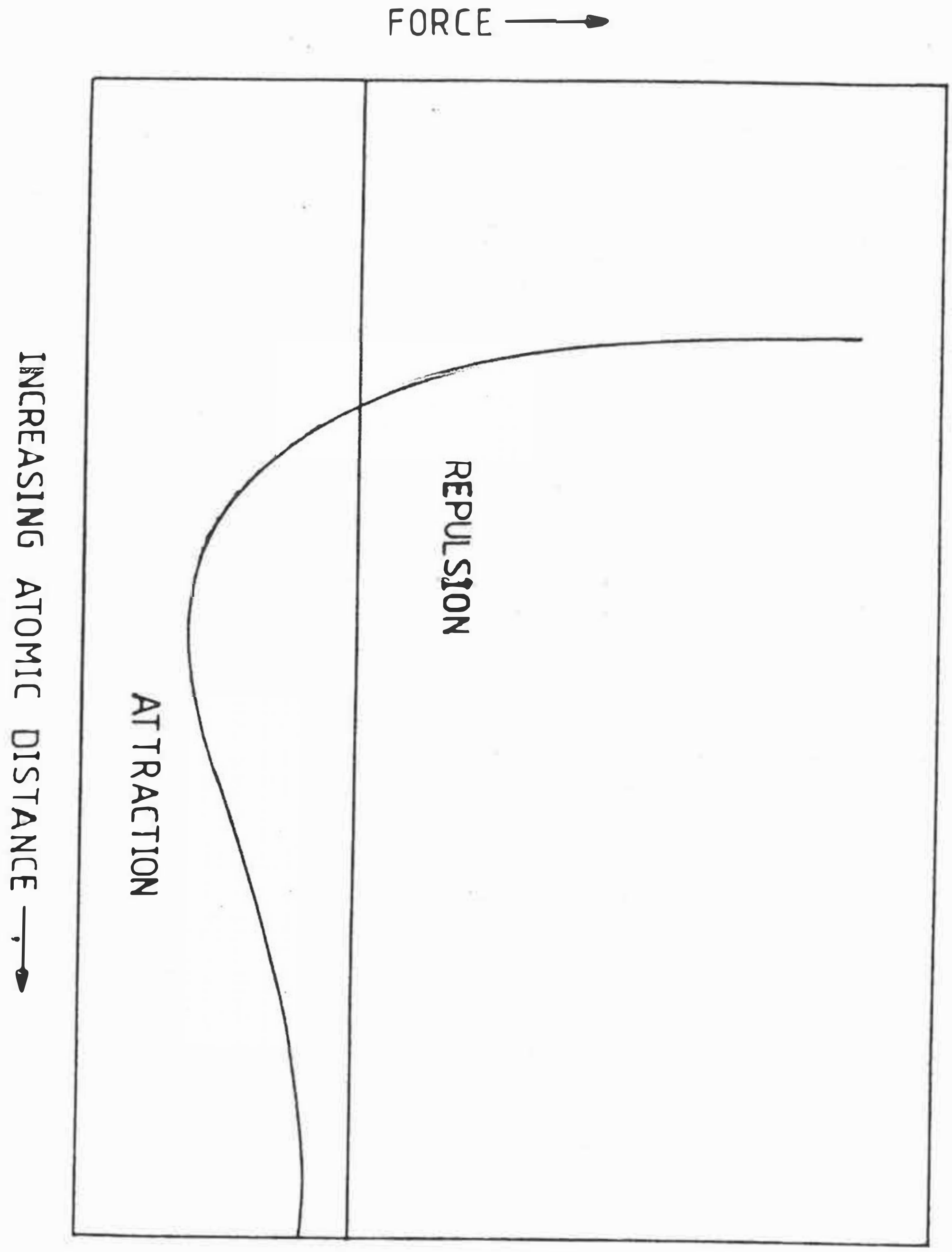


FIG. 3.6 SCHEMATIC DIAGRAM SHOWING FORCES OF INTERACTION BETWEEN TWO SURFACES AS A FUNCTION OF THEIR DISTANCE APART.(24)

attracted, and then, on real contact, exhibit the resistance to compression customary with solid metals. Thus the more points of intimate contact that are achieved the larger will be the net cohesive force holding the grains together. This argument implies that the surface condition of the grains plays a very important role in achieving strong bonding of the atoms. The surface layers that are greater in depth than a few Angstrom units would be sufficient to obstruct the interatomic bonding. Surface irregularities are resistance to plastic deformations are also additional factors preventing surfaces coming into complete atomic contact. The most common form of interference is that of oxide films which are formed by reaction with the atmosphere during manufacture, storage or handling of the powder. In fact, most metals at room temperature rapidly acquire an oxide film. As well as oxygen some metals such as Fe, Pb, Cd, Al may absorb other constituents from the atmosphere such as some layers of sulphide molecules, H_2O and Co_2 molecules and other gases. Although it has been pointed out that oxidation could occur as a result of the friction produced during compaction (85), for many powders, including sponge iron powder, the oxide films are immediately pierced or rubbed away during compaction unless they are very thick.

3.2 Theory of Sintering for Metal Powder

Although the process of sintering is not fully understood there are some well known theories on which a consensus agreement has been made. A more detailed consideration of the various theories may be found in almost any standard powder metallurgy textbook (19, 20, 24). In the following paragraphs a brief explanation will be given with reference to the main theories.

Under suitable circumstances, particles of metal powders when heated will adhere to each other, and, as the time or the temperature of heating is increased, the extent of adhesion increases. Temperature has a greater effect than time. The first reaction of a powder mass introduced to a particular temperature is increased strength and increased thermal and electrical conductivity. Hence a loose mass of powder becomes a porous solid having a certain physical strength. The increased strength is accompanied by a decrease in porosity. Grain growth may occur at certain conditions of temperature and time.

Sintering may be defined as heating a particular body, below the melting point of at least one major constituent, in order to cause interparticle bonding. In addition to causing interparticle bonding, sintering can also lead to the following important effects: (a) chemical changes, (b) dimensional changes, (c) relief of internal stresses, (d) phase changes and (e) alloying. When a compacted powder body is sintered a simple measure of particle bonding is its macro-strength. Figure 3.7 shows schematically how strength and some other properties may vary with sintering temperature.

The mechanism of sintering is dependent on the free surface energy. The tendency for a system to assume its

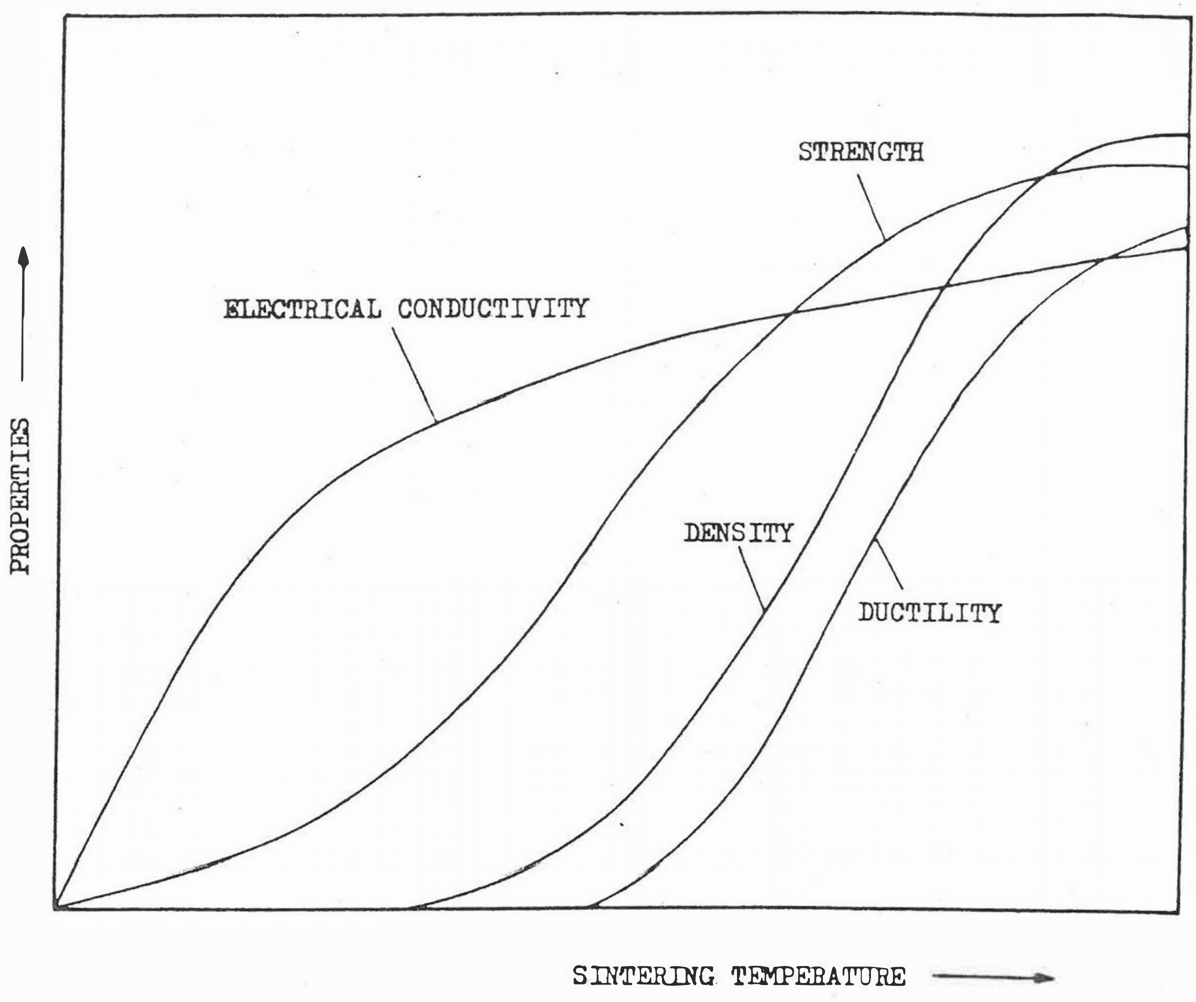


FIG. 3.7 THE INFLUENCE OF SINTERING TEMPERATURE ON THE PROPERTIES OF A TYPICAL SINTERED BODY.(88)

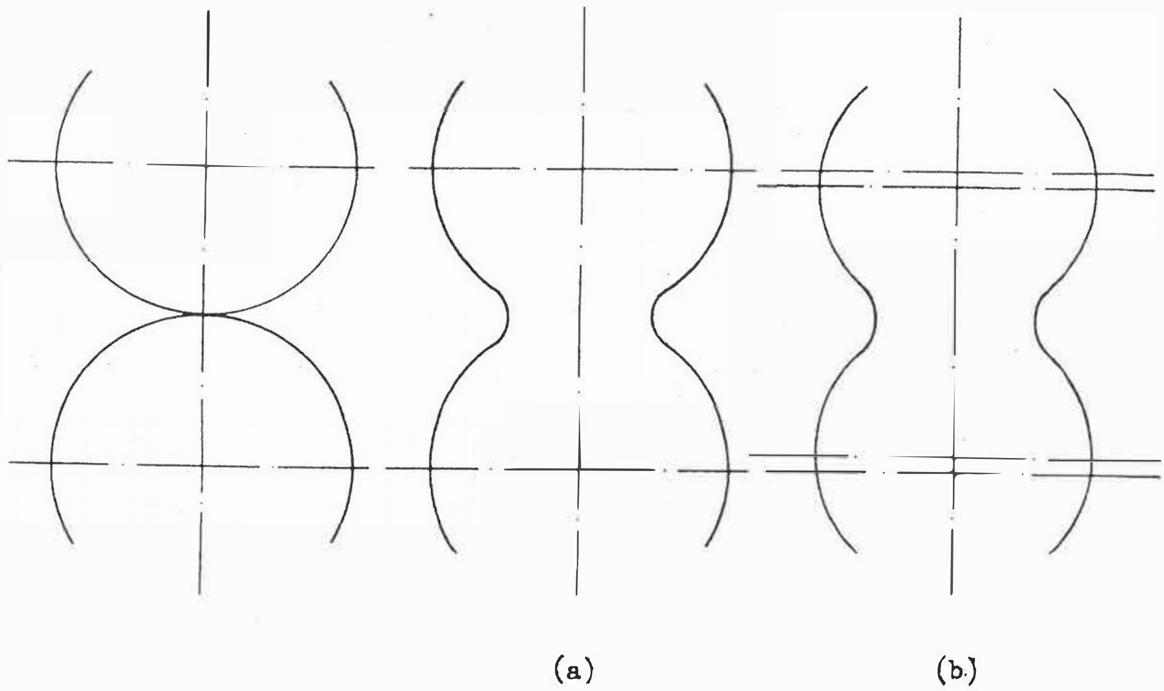


FIG. 3.8 FORMATION OF WELD NECK DURING SINTERING:

(a) WHEN THERE IS NO OVERALL SHRINKAGE AND

(b) WHEN SHRINKAGE OCCURS.

state of lowest energy is the driving force for sintering. In a powder mass there is excess energy due to the large free surface. The amount of excess surface energy, however is not large. Kuczynski (86) has calculated that one gram-mol of 1 μm diameter spherical copper powder will have only about 10 calories of excess energy. During sintering, the surface area is decreased by an increase in interparticle contact area and by smoothing of the particle surfaces. Thus the energy available to continue the process becomes less and the rate of change slower. This may well continue until the solid density is reached where the total surface is equal to the external area of the compact.

One other approach has been put forward by Alexander et. al. (87) who includes some mathematical analyses of the suggested concepts. This model of the sintering process is that of formation of weld necks between the particles and growth of these necks until overlapping occurs. The network of interconnected channels then becomes a series of isolated pores, and finally a considerable increase in density occurs. With the occurrence and extent of particle bonding and dimensional changes, the porosity of the compact tends to decrease. To illustrate this phenomena, first consider the case of two spherical particles which are just touching, the shaping process not having caused any particle deformation. Fig. 3.8 shows two cases of a neck forming between particles. In fig. 3.8 a the particle centres have not moved closed together, and thus the material supplied to the neck has been obtained by dislocation of the particles.

In this case, since the particle centres are still at the same distance apart, a gross dimensional change has not occurred. The situation in the second case of Fig. 3.8.b is different. Here the particle centres have moved closer together and, if these spherical particles had been part of a larger powder mass, macroscopic shrinkage would have been achieved by either evaporation and condensation, surface diffusion, volume diffusion or plastic flow. Once intimate contact has been established at the junction between two grains, there exists a stress which is acting in a manner to cause the neck to grow. Kuczynski (86) has considered these stresses which may arise in the neck formed between two spherical particles due to surface tension forces, and has also shown their implications. He correlated the surface tension to the equilibrium vapour pressure, condensation, evaporation, and vacancy diffusion processes. Evaporation and condensation is only possible from a convex surface to a concave one. Furthermore, such a material transport is only feasible over very short distances. The effect of surface tension could also be a reduced density in the neck hence favouring volume or surface diffusion into the neck. The continued growth of these necks progressively leads to the formation of isolated pores in the material. As the closure of pores proceeds, the effects of evaporation and condensation, and surface diffusion ceases. The influence of volume diffusion becomes diminished as it requires the existence of vacany gradients which are less likely as the channels are now closed and the pores more spherical in shape. For the explanation of the final part of densification another hypothesis of grain boundary diffusion is suggested. The idea is that recrystallisation causes the boundaries to more along gathering up the pores and transferring them to the surface as vacancies along the grain boundaries themselves. The final stages of densification are explained mainly by two concepts: (a) diffusion down the grain boundaries from the external surface or from the boundaries themselves (b) diffusion through the volume of the metal, either from the external surface or from the boundaries.

3.3 Forging

3.3.1 Principles of Conventional Forging for Solid Metals

Conventional forging as a metallurgical manufacturing technique has been practised for reasons of producing lower-cost products with high mechanical properties compared with machining or casting.

Most of the conventional forging processes involve with flash formation in upsetting, edging and fullering on the simple geometrical shape open dies (89). The forged part is generally in a semi-finished state and requires subsequent machining for the desired final configuration. As it was mentioned in the Introduction, the material untilisation during the production of parts by conventional forging is often very poor, because of the flash formation. This, when added to the material which is machined off, means that the weight of the finished part is often only 30% - 50% of the raw material used (7,8).

Conventional forging starts with a fully dense forging blank and hence produces a fully dense component. Dimensional accuracy, however is poor. Therefore, in order to produce precision forgings with a minimum of material wastage, blocking dies (closed-dies) are generally used. In closed-die forgings, the forging billet is usually first fullered and edged to place the metal in the correct places for subsequent forging operations. The preshaped billet is then placed in the cavity of the blocking die and rough forged closed to the final shape. It is then transferred to the finishing die where it is forged to final shape and dimensions. The final step in making a closed-die forging is the removal of the flash with a trimming die. In addition to these sequences, the provision of a flash and gutter around the circumference of the precision die shape is necessary. In order to ensure complete die filling, a slight excess of metal is normally used. As the dies come together for the finishing step, the excess metal is forced out of the die cavity as a thin ribbon of metal called flash.

PROCESS FLOW DIAGRAMS

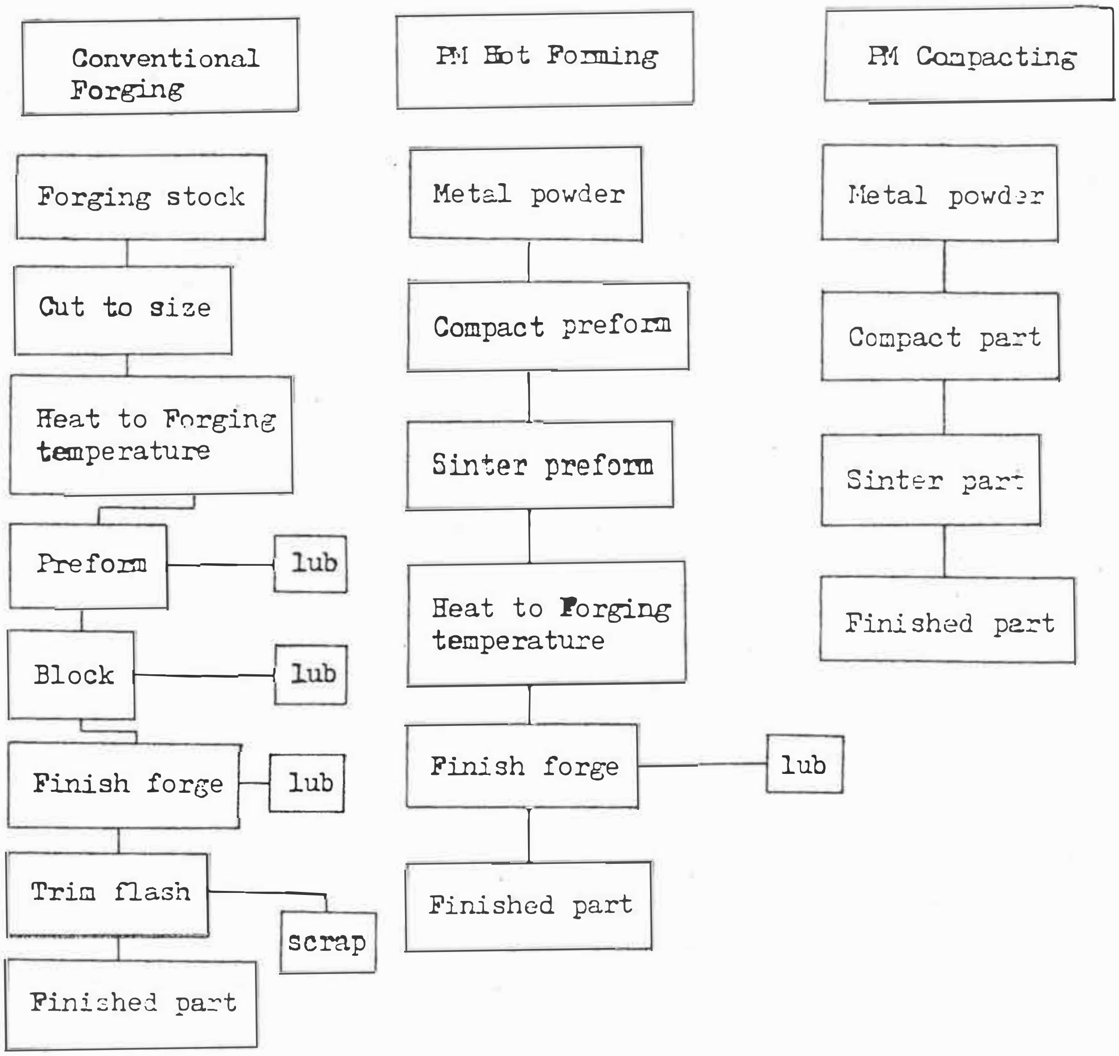


FIG. 3.9 STEP BY STEP FLOW DIAGRAMS OF CONVENTIONAL FORGING, P/M HOT FORGING AND P/M COMPACTING.(90)

When compared with P/M hot forging, conventional forging, because of the number of subsequent operations such as blocking, trimming and possibly piercing, requires more energy.

An important consideration is that the process of conventional forging must overcome the inherent problem of high tool stress, which is the limiting factor in forming intricate components from the solid and results in high tool wear and attendent difficulties in holding close tolerances over a large number of components.

Because of the various die sets, excess material and considerable machining expenses, the overall production costs of the conventional forging become quite high. In Fig. 3.9 for comparison, process flow diagrams of conventional, P/M hot forging and P/M compacting are given.

3.3.2 Powder Preform Forging

The use of a well controlled quantity of powder to mould an accurate preform from which a finished precision forging is produced without the formation of flash has several advantages. Precision flashless components formed from powder preforms can often be produced with a single blow in closed-dies, with an immediate reduction in forging cost. The fully closed-die P/M hot forging technique will result in the highest possible component density and material saving. With careful preform design, therefore P/M hot forging is virtually scrap free. The tolerances of high precision P/M hot forging should approach the tolerances achieved in sintered-part technology, but without the traditional drawback of residual porosity. In addition to material utilization, P/M hot forging also enables very high production rates to be obtained and lends itself to automatic control. (91).

The mechanism of deformation in hot P/M forging differs significantly from that in conventional hot forging, requiring less lead on the tooling and hence longer die life. Also, parts are hot-formed at lower temperatures, which results in lower die wear. Much of the final shape is developed during the early stages of the forming when the loads required are relatively low. At the high forging load very little metal flow occurs, and the predominant process is that of hot compaction, from which the fine detail of the finished forging is achieved. From these arguments, it will be inferred that the hot P/M forging process contributes considerably to reduced die wear. Hot formed parts generally have better as-formed surface finish than their conventionally forged counterparts due to well planned preform design and consequently uniform material displacement throughout the part.

The above mentioned features of P/M hot forging stresses the great potential for the production of components having intricate detail.

As it was mentioned in Chapter 2, great care must be taken the design of preforms, their manifacture as well as the control of the forging variables. Since the most beneficial outcomes of P/M hot forging depend on correct preform design, this main problem needs to be carefully elucidated. From the economic as well as component strength point of view, preforms would have the simplest possible shape, while from the aspect of a successful forging operation a more complicated shape may be required in many cases. Although the compromise between these two requirements, can be found by a trail and error approach, recently there is a tendency towards preform design through computerized techniques. Summarizing all these features, forging of powder preforms can be considered as the synthesis of advanced P/M and advanced forging techniques.

CHAPTER FOUR

DESCRIPTION OF PRESSES, APPARATUS AND PROCEDURES

In the following section, machines, apparatus, equipment and materials used in the powder forging experiments are described. Descriptions of the relevant testing procedures are also given below.

4.1 High Speed Petro-Forge Mk.II Forging Press:

Hot P/M forging of sintered iron-base billets were carried out on the high speed Petro-Forge Mk.II press. The Petro-Forge is a type of high energy rate forming machine which has been developed in the Department of Mechanical Engineering at the University of Birmingham. These high speed machines have been described by Chan et. al. (92). The nominal rating of the model used is as follows:

Rated energy at 0.84 MN/m^2	(13,560 Nm)
Charge pressure	
Rated velocity	(10.67 m/sec)
Rated Stroke	(228.6 mm)

Cycle time	less than 1 sec
Dwell time	1 - 10 m.sec
Moving weight	(224.3 kg)

The ejection procedure can be timed to start as soon as the ram begins to retract.

4.2 Hydraulic Presses:

The presses used in the powder compaction were the Denison TIA/MC 300 tonf. (3000 KN) and Denison T42 B4 500 KN hydraulically powered testing machines. The maximum die closure rate for the Denision TIA/MC was 1.27 mm/sec and for the Denison T42 B4 was 3.33 mm/sec.

The two presses were used for compaction and ejection of the preforms, as well as for load cell calibrations.

4.3 Sintering Furnace

4.3.1 Sintering Atmosphere

The sintering atmosphere of cracked ammonia gas was selected because of its wide industrial use and well suited feature, particularly for sintering iron based powder. Oxygen free nitrogen was used first to purge the furnace of all air prior to the introduction of the cracked ammonia gas. The mixture of three parts hydrogen to one part nitrogen not only prevents oxidation of the powder but is a very effective reducer of any oxide that should happen to be present. To a great extent this is dependent on the sintering time and the degree of access of the gas to the inside of the compacts. For the sintering of iron base or copper materials there is no danger of nitriding, even though there may be such a risk, leading to hardness and embrittlement, with stainless steel compacts. A flow rate of 0.39 cu. meter/hr was maintained throughout the sintering processes at a pressure of 1.05 kg/sq. cm. The gas was burnt off in air at the end of the furnace tube.

4.3.2 Construction

The sintering furnace employed in the experiments consisted basically of an impervious mullite furnace tube of (101.6 mm) diameter and 910 mm in length, surrounded by four axially placed crusilite heating elements. The arrangement was well insulated by refractory bricks and asbestos plates. A water cooled section of the tube enabled cooling to ambient temperature in the furnace atmosphere. The length of the heating zone was 228.6 mm and was suitable for use up to 1300°C, to an accuracy of $\pm$ 5°C. The gases were admitted at one end of the tube, passing through the water cooled section and heating zone before exhausting through 6.35 mm diameter hole in the other end.

The thermocouple was of platinum/rhodium and projected into the furnace tube at the centre of the heating zone. As hydrogen reacts with platinium, the couple was protected by an impervious mullite sheath which was vented to atmosphere. The sintering furnace has the following specification:

Type of furnace	12 kw R.S. Moly furnace
Voltage	400/440 3 phase, 50 cycles
Maximum temperature	$1300^{\circ}C \pm 5^{\circ}C$
Heating elements	Molybdenum wound
Heating zone	228.6 mm
Tube diameter	101.6 mm
Voltage temperature	Automatic control type
Thermocouple	Platinium/Platinium 13% Rhodium

4.4 Powder Material

The principal powders used in the experiments were AHC.100.29 atomized iron powder and ATST-A low alloy atomized steel powder. A list of typical physical and chemical properties are given in Tables 4.1 and 4.2. These types of powders were selected for experiments where combinations of high strength, hardenability and good ductility in composite iron-steel parts were to be investigated.

Carbon was introduced to the steel powder in the form of Rocol-Graphite X 7119. This was blended with the powder in a Y-cone mixing unit (Apex-Blender type 165/CX05). In assessing the correct graphite addition to be made to a mix, it must be remembered that not all of the graphite will be converted to combined carbon in the steel. Carbon dioxide will be formed by the reaction with surface oxide films and inclusions during sintering. The oxygen content of the air in the pores of the compacted part also contributes to the graphite loss. In order to estimate the graphite necessary to achieve a certain carbon content an empirical formula is generally used. (19).

$$G = C + 3/8L.S$$

Where G = percentage graphite to be added

C = final percentage carbon content of the steel

L = percentage of hydrogen loss.

S = shape factor, which varies from one for spherical powders to two for extremely irregular shaped particles.

The powders produced by atomization are generally more spherical in shape compared with the powders produced by reduction techniques. In the majority of powder metallurgical application, the added graphite content is usually between 0.5% / 1%. In the experiments presented here, a reasonably high graphite content of 1% was chosen to investigate its effects particularly on the hardness distribution at the interface of powder layers.

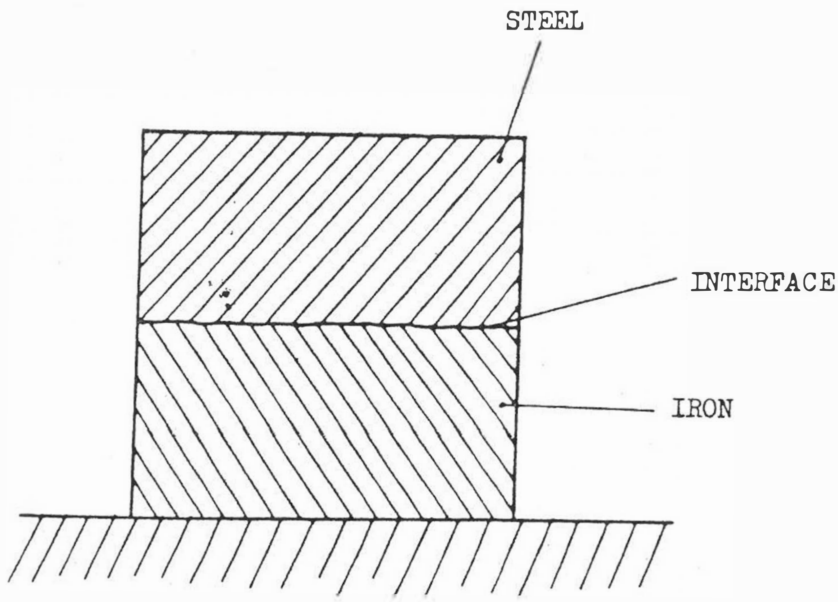
4.5 Tooling

The die sets used in the experimental work can be classified into two categories, (a) compaction dies, (b) forging dies. The design of tooling for both compaction (93) and forging (94-96) was largely based on existing experience. Compaction and forging dies from previous work were utilized although some modifications were made for the forging die set.

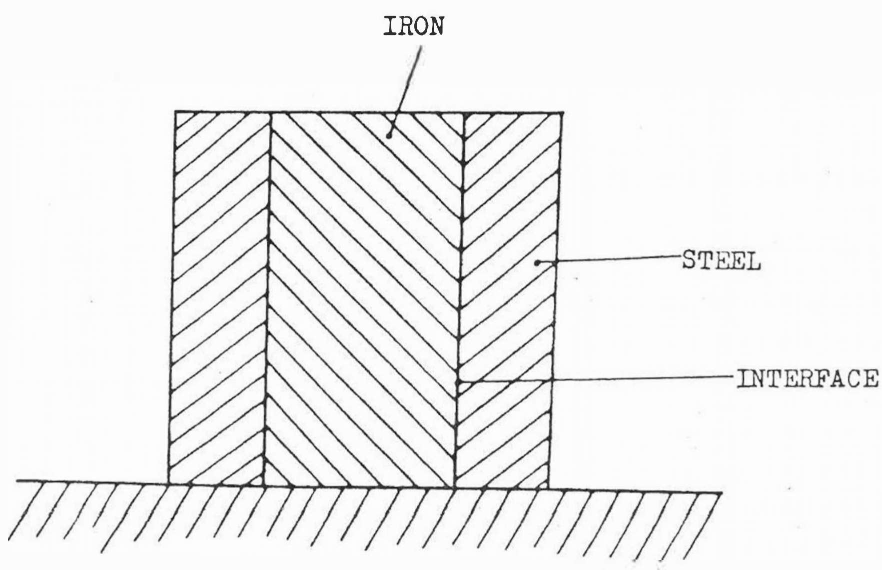
4.5.1 Compaction Tooling

Two cylindrical compaction dies of 25.4 mm and 38.1 mm bore diameters were used in the experimental work to produce powder preforms.

The choice of material for various parts of the tooling is an important aspect for the experimental work. In industrial practice, where production demands are high, the lining and punch faces are made of cemented carbides, but as their use cannot be justified for laboratory experiments, various tool steels were employed. Carr's 695 (containing 1.5% C, 12% Cr, 0.75% Mo, 0.25% V and balance Fe) was used for pressure pads, punches, anvils and dies. The 25.4 mm diameter die was chrome plated to a depth of 0.125 mm to reduce wear. The 38.1 mm diameter



(a) FLAT BOUNDARY.



(b) CIRCUMFERENTIAL BOUNDARY.

FIG. 4.1 LOCATION OF IRON AND STEEL PARTS IN THE TWO-PART COMPACTS.

die was nitrided. The punch, pressure pads and anvils are hardened to 60-62 Rc. The clearence between the various parts of the die assembly is quite critical. They must be small enough to prevent powder flashing and sufficiently large to permit the large volume of air trapped inside the loose powder to escape during the very short time it takes to compact the mass. Therefore a clearance of about 0.025 mm on the diameter was allowed for the 25.4 mm diameter die and 0.040 mm for the 38.1 mm die. Single end compaction was used throughout the tests.

In the compaction of two part composite AHC 100.29 iron - ATST-A low alloy steel parts, a special technique was followed. In the first experiment 150 g. of iron powder were placed in the bottom of a 38.1 mm diameter compaction die and the top surface levelled off. 150 g. of the low alloy steel powder were then added and the whole 300 gm. mass was pressed together at 607.6 N/mm^2. (Fig. 4.1a). The resultant preform did have a density variation, because the iron powder, being more compressible, was compacted to a high density than the low alloy steel powder. As there is always a slight density decrease along the pressing direction, due to pressure transmission losses, it was hoped that by placing the alloy steel above the iron this density variation would be reduced. However, the density variation in the preform may not be regarded to be significant since almost all the porosity is eliminated in the subsequent forging operation.

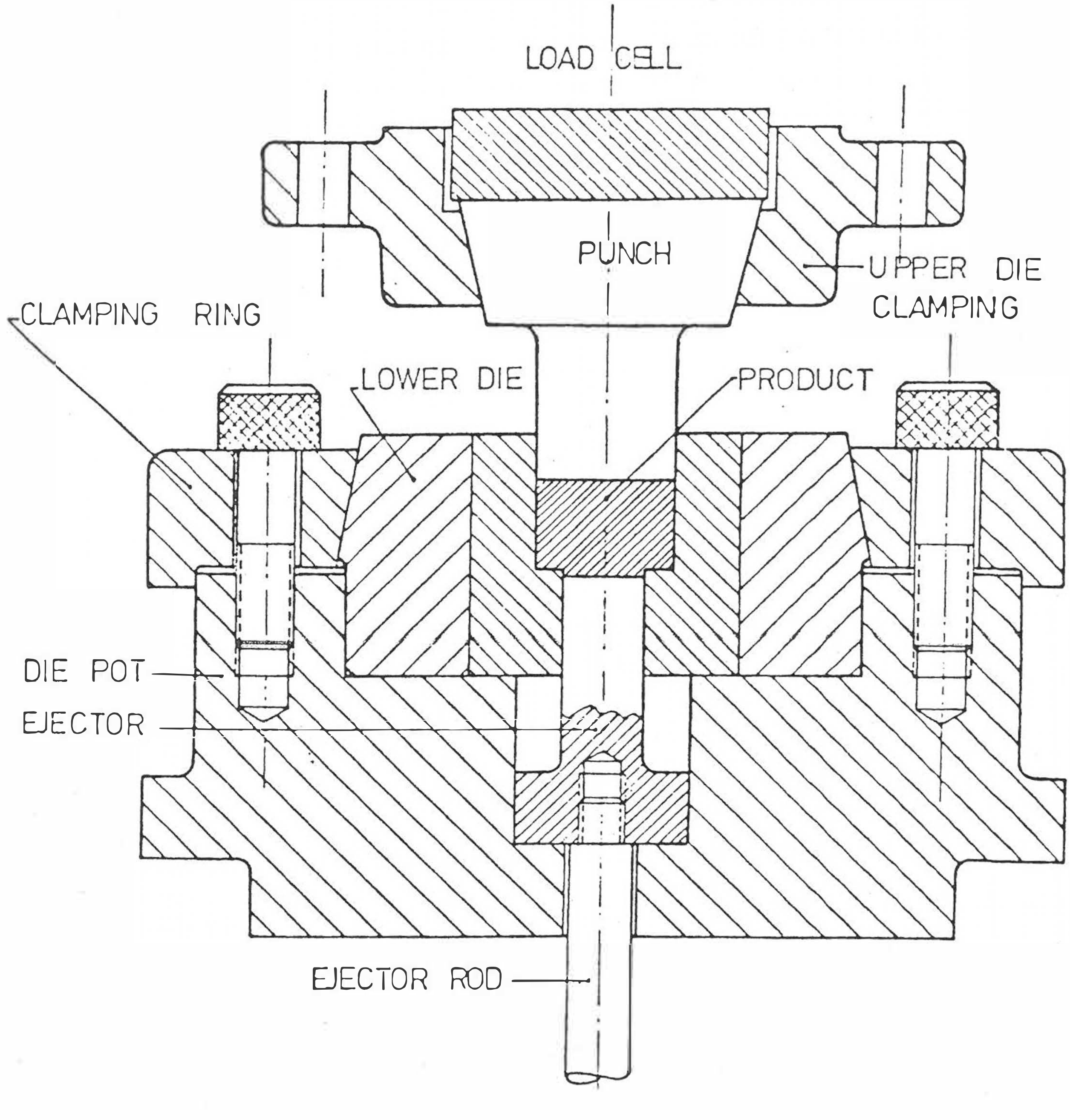


FIG. 4.2 38.1 mm DIAMETER CLOSED-DIE ARRANGEMENT FOR HOT P/M FORGING.

In the other experiment, a 25.4 mm diameter rolled tracing paper was placed in the middle of the die, the central cavity of which was filled with iron powder and the remaining space between the tracing paper and die wall was filled with low alloy steel powder. (Fig. 4.1b). Both powders were simultaneously precompacted by ligth tapping and the tracing paper was taken out. The whole mass was pressed together at 607.6 N/mm^2. Although there was some intermingling of the separate powder masses, the resultant compact consisted of two distinct material layers

4.5.2 Forging Tooling

The forging die set used in the experiments was a 38.1 mm cylindrical close die arrangement, as shown in Fig. 4.2. In production of this die, Jessop-Saville H50 (containing 0.37%C, 1.10%Si, 5%Cr, 1.1%V, 1.35%Mo and balance Fe) die material was used. It is noted for its excellent resistance to washing and thermal shock. After machining, the die was heat treated to a hardness of 45-48 Rc. Strohecker et. al (97) found that dies and punches of H13 (an equivalent material to H50) hardened to 50 Rc are satisfactorily, but the hardness of 40 - 46 Rc gives reasonable die life and permits machining when modification is needed. (97). In the forging die assembly the clearance between the die and the ejector was about 0.05 mm. The die assembly shown in Fig. 4.2 is a good example of the use of die inserts in forging. These can be quickly replaced after failure or interchanged for the production of a new component. The main features of this design are:

(a) The die insert is a duplex cylinder. This provides sustained compressive stresses on the inner ring and therefore offers advantages over a tapered clamping ring which depends for its effect on the tightness of the bolts.

(b) The die insert is tapered at its outer surface which mates with the die pot and the clamping ring. These tapers are designed to provide a further support from the bolster and clamping ring. Furthermore, with the use of the proper tapers, adequate frictional forces can be obtained which prevents the rotation of the die insert about its axis.

(c) The punch is a single unit held within the punch holder by a 7° taper (14° inclusive) that prevents rotation of the punch about its axis. Dies of this type require accurate guidance to ensure perfect mating of punch and die and have been used successfully on high speed Petro-Forge machines.

4.6 Lubricant

4.6.1 Compaction Lubricant

As was mentioned in the section (3), because of its better compaction and sintering response 'Acrawax C' (a product of Glycor-Product Inc) lubricant was used. 'Acrawax C' improves pressure distribution during compaction and facilitates ejection of the part from the die after pressing. Acrawax is burnt off at $750^{\circ}C$ leaving no residue (84). Throughout the experimental work the powders were mixed with 1% Acrawax in a Y-cone blender.

4.6.2 Forging Lubricant

The lubricant used for the forging tooling was 'copaslip' which is a mixture of lead and copper particles in bentane grease. The lubricant was applied to the dies using a small paint brush giving a uniform, thin coating. (98).

4.7 Instrumentation

The forging loads were simultaneously measured on a time base storage oscilloscope and photographed for future reference and analysis. (Chapter 5, Fig. No. 5.Pl)

4.7.1 Load Measurement

Forging loads were measured using a strain gauge ring-type load-cell, that replaced the packing piece as shown in fig. 4.2 This type of ring load-cell is similar to that recommended by Jain and Amini (99) for use in High Energy Rate Machines. The load cell was made from a die steel of Orvar 2 Uddeholms, (containing 0.37%C, 1%Si, 0.40%Mn, 5.3%Cr, 1.4%Mo, 1.0%V) and heat treated to 47 Rc. Eight strain gauges type PL-105-11 (Tokyo-Sokki Kenkyujo Co. Ltd.,) with a nominal resistance of 300 atom and gauge factor of 2.07 were cemented to the outer surface of the ring-cell in a halfbridge configuration

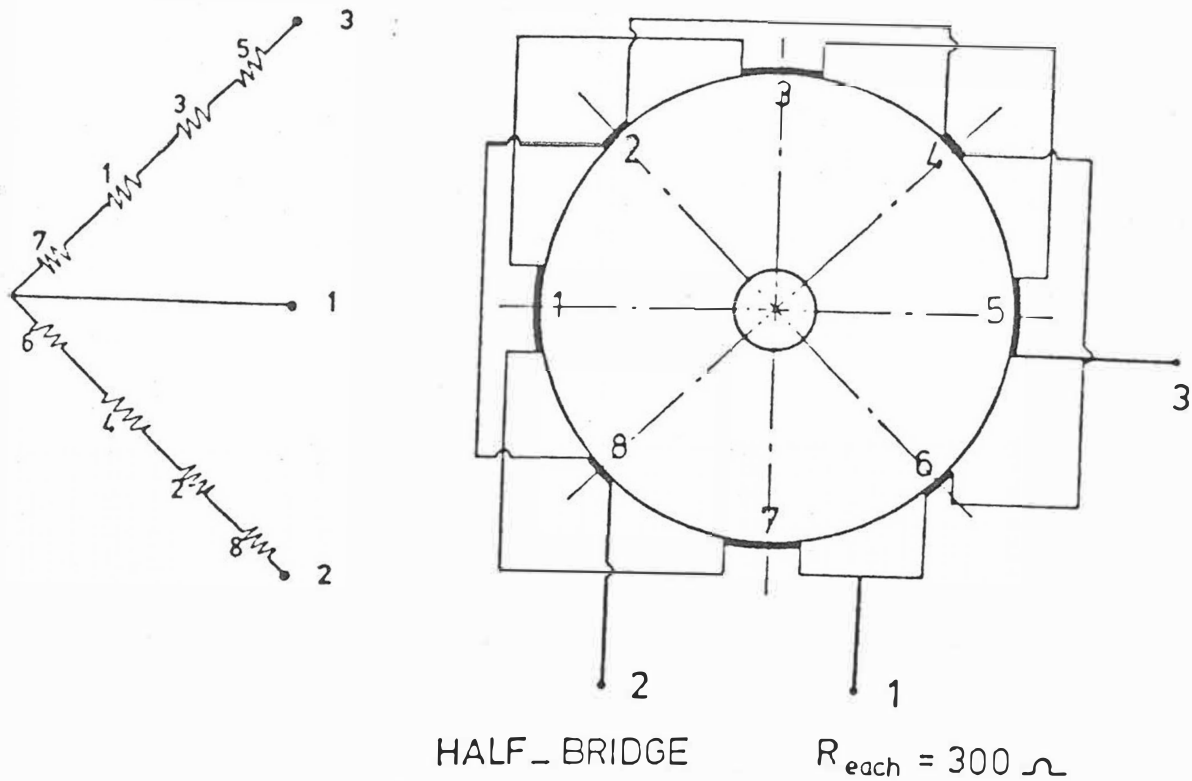


FIG. 4.3 HALF BRIDGE STRAIN GAUGE CONFIGURATION ON LOAD CELL.

as shown in Fig. 4.3. The gauges were placed alternately axially and transversely. The four transverse gauges increased the sensitivity, due to Poisson's ratio effect, as well as providing temperature compensation. The load cell was calibrated using a 3000 KN hydraulic press, and before calibration, the cell was load-cycled for some time in order to reduce or eliminate any hysterisis effects. The load-cell was connected to a Hottinger-Bridge type KWs 11/50 and the output from the bridge fed into one of the channels of a Textronix oscilloscope type 564. The calibration curve shown in Fig 4.4 was obtained by suddenly applying and releasing the load and recording the corresponding voltage output for the maximum load. For loads which are less than 200 KN, the calibration curve was non-linear. This non-linearity, however could be eliminated during experiments by initially compressing the cell using the clamp ring until it gave an output equivalent to 200 KN. By resetting the bridge output to zero, the subsequent readings could be expressed as a linear function of the voltage output. The calibration was repeated several times throughout the experiments.

4.8 Density Measurement

All densities of preforms and forging were measured according to the Archimedes immersion method. The specimens were first treated with a mixture of acetone + 0.1% silicone oil (93) to prevent the ingress of water particularly to the compacted preforms. The

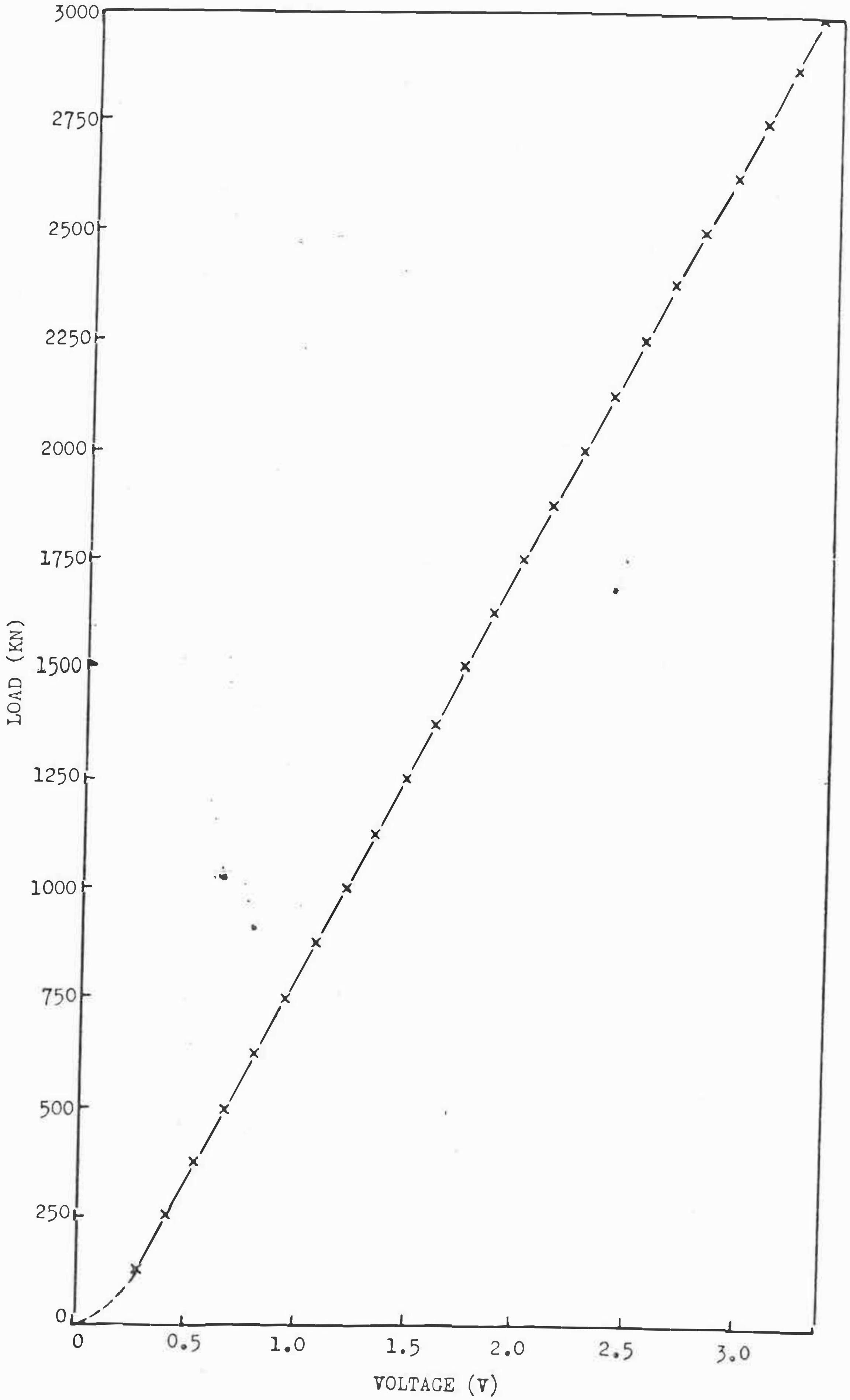


FIG. 4.4 LOAD CELL CALIBRATION CURVE.

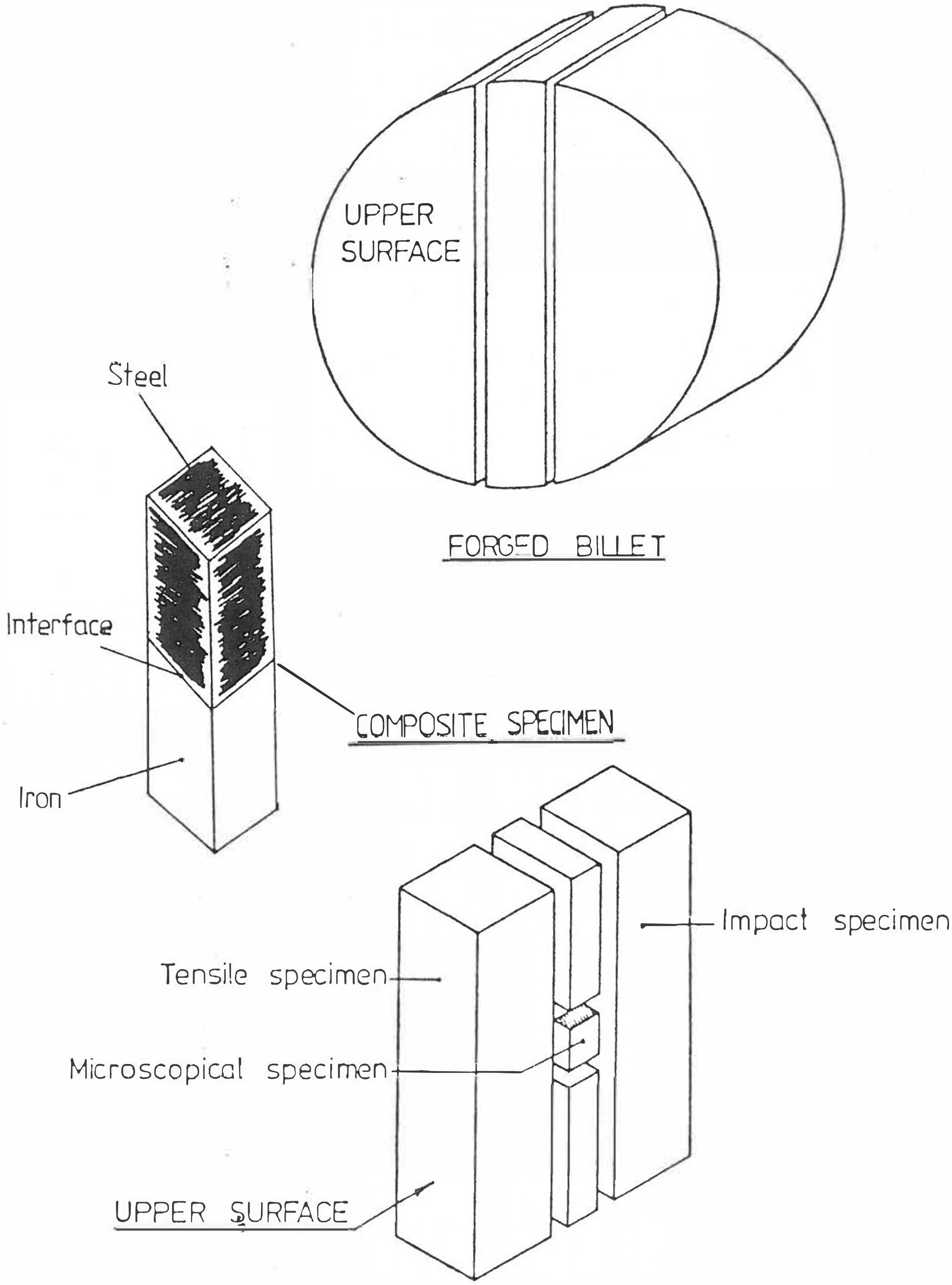


FIG. 4.5 POSITION OF THE TEST SPECIMENS.

resulting density measurements maintained an accuracy within ± 0.0025 g/cm^2.

4.9 Impact Test Machine and Procedures

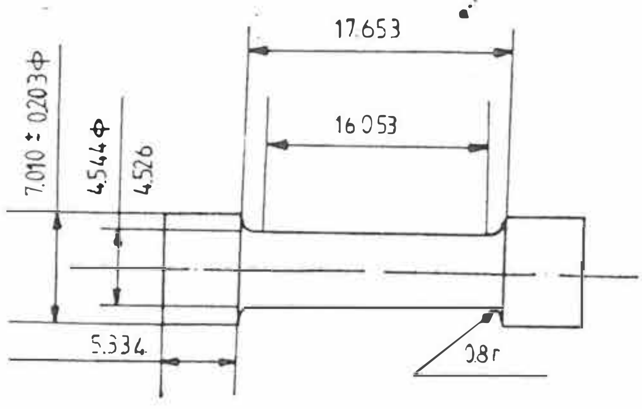
For impact tests a pendulum impact machine type PSW 30 kg (Losenhousenwerk, Dusordorfer. Machineau AG, West Germany) was used. The machine was available for two ranges of impact energy measurements, 294,2 and 147.1 Nm. (217 and 108.5 ft.lbs), and was suitable for tests on several charpy impact specimen sizes. Present tests were carried out at the 294.2 Nm range. The test specimens were made from the lower middle of the forged billet (transverse direction) as shown in fig. 4.5 for the plain iron and low alloy steel specimens. In the composite iron-steel specimens the interface was approximately in the centre of the specimen, as is shown in Fig. 4.5. All the specimens were ground. The specimens were unnotched and had a rectangular cross section of 10 mm height x 8 mm width and a length of 38.1 mm. During testing each specimen was supported at both ends leaving a gauge length of 28 mm unsupported.

4.10 Tensile Test Procedures:

Tensile tests of the hot P/M forgings were carried out on a Hounsfield Tensometer type "w". The single material AHC 100.29 iron and ATST-A low alloy steel specimens were taken from the upper middle of the forged billet (transverse

Twice full size

All dimensions in (mm)



MATERIAL: HOT P/M FORGED AHC 100.29 IRON + ATST-A

LOW ALLOY STEEL.

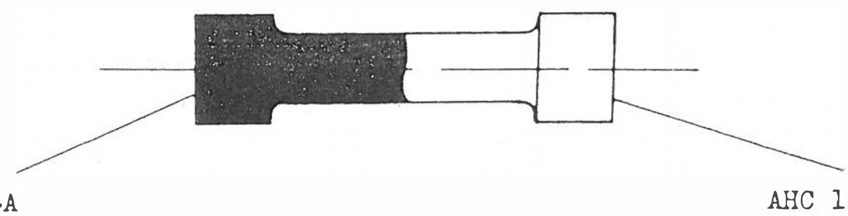


FIG. 4.6 NO.12 HOUNSFIELD TENSILE TEST SPECIMEN.

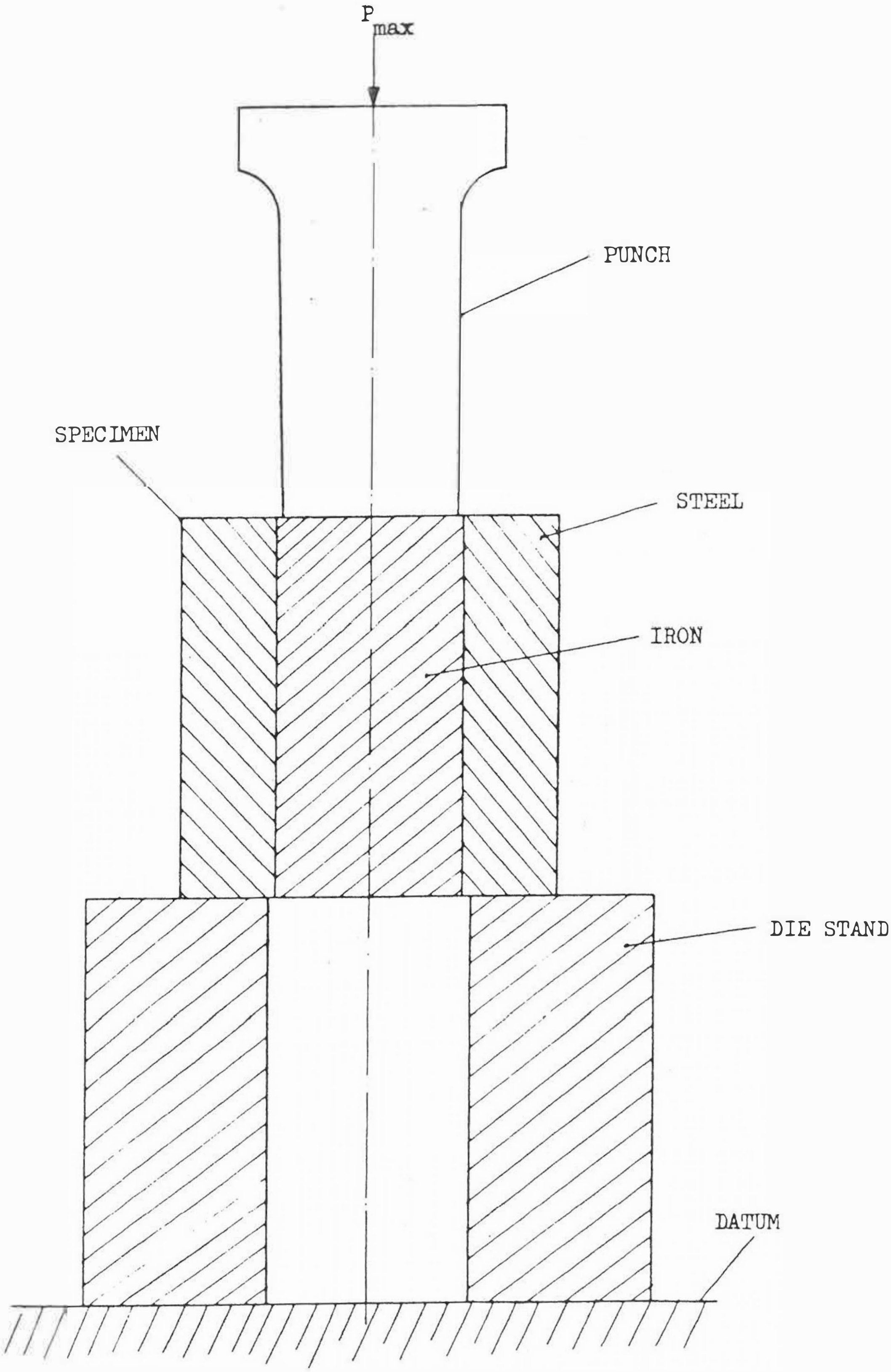


FIG. 4.7 SHEAR TEST DIE SET ARRANGEMENT.

direction) as illustrated in Fig. 4.5. The composite iron-steel specimens were prepared in such a way that the joint was approximately in the centre of the gauge length. The specimen type was of the No. 12 standard Hounsfield Tensometer type 'w' as shown in Fig. 4.6. For three different compositions of specimens, ultimate tensile strength, elongation and reduction in area measurements were made.

4.11 Compressive Shear Test

The die set arrangement for compressive shear testing is illustrated in Fig. 4.7. Composite iron-steel cylindrical specimens of 38.1 mm diameter x 38.1 mm height, were prepared with the top and bottom surfaces being maintained parallel to each other. Cylindrical powder forged specimens consisted of a 25.4 mm diameter inner cylinder of iron and outer ring of low alloy steel. Compressive load was applied to the iron part of the specimens with a nominal 25.4 mm dia. punch. Maximum shear loads were recorded on the chart attached to the testing machine. The Ultimate shear stress was calculated by using the following formula:

$$U.S.S = \frac{P_{max}}{A_s}$$

where,

U.S.S. = Ultimate shear stress

P_{max} = Maximum shear load

A_s = Sheared Area

A_s = π x d_i x h

d_i = inner ring diameter (iron)

h = height of the composite specimen

TABLE 4.1.

Chemical Analysis of Powder Materials

Powder	Chemical Analysis %						
	C	Mn	Ni	Mo	S	P	H_2-loss
ATST-A Steel (Atomized)	max 0.15	0.25-0.35	1.8-2.2	0.45-0.55	max 0.03	max 0.02	
AHC.100.29 Iron (Atomized)	max 0.02						max 0.20

TABLE 4.2

Sieve Analysis & Physical Properties of Powder Materials

Powder	Sieve Analysis %			Physical Properties			
	+65 mesh	+100 mesh	-325 mesh	Compressibility (g/cm^3)	Flow Rate (sec/50g)	Upper Particle Size (mm)	Apparent Density (g/cm^3)
ATST-A Steel	0	max 10	20		25		3.0
AHC.100.29 Iron				min 6.65	max 28	0.17	2.95

CHAPTER FIVE

EXPERIMENTAL RESULTS & DISCUSSION

5.1 Introduction

In order to investigate the behaviour of composite iron base materials during forging, the toolset used was for the recompaction (consolidation) condition. In this deformation condition, the diameter of the preform was nominally the same as that of the die.

Since the number of variables involved in powder forging are quite considerable, an attempt was was made to reduce these to a reasonable amount. The main variables under consideration were; 1. Sintering and Forging temperature, 2. Sintering time. Sintering and forging was carried out at the same temperature.

For the production of preforms, the compaction pressure was approximately 627 N/mm^2, giving an initial average density of approximately 6.65 g/cm^3. The energy level of the Petro-Forge machine was 1400 Nm, corresponding to a final die impact speed of 6 m/sec.

In order to eliminate the effects of variations of cooling rate in the forged billets, they were annealed at 900°C for 1 hour and then furnace cooled.

The experiments are divided into two sections; 1. Temperature range tests, 2. Time range tests.

1. Temperature range tests:

The effect of varying the sintering and forging temperature from 950°C to 1200°C in 50°C increments was investigated. Since for most of the customary powder forging applications, the lower temperature limit for hot working conditions is about 900-950°C, a minimum temperature of 950°C was chosen for analysis. The sintering time was kept constant at 40 minutes.

2. Time range tests:

Predetermined sintering times ranged between 20 ÷ 60 minutes in 10 minute increments.

5.2 The Results of Temperature Range Tests:

The variation of peak load over the sintering and forging temperature range is shown in Fig. 5.1. The varying temperatures on this and subsequent figures are the indicated sintering furnace temperatures. The loss of temperature of the billet during transfer to the forging press was between 20°C and 40°C (95). The rise in billet temperature due to the actual forging operation was also calculated to be about 35°C (96). This means that the temperature loss of a billet is offset by the temperature rise of the billet itself during forging.

Since the structure of the material remained in the γ-phase throughout the experiment, a gradual fall in the peak load was observed as the temperature was increased from $950^{\circ}C$ to $1200^{\circ}C$. This is readily explained by the fact that within the same phase, the hot strength of the iron base materials decreases as the temperature is increased. In his investigation of the forces during forging of iron powder preforms, Huppmann (70) found that a pronounced minimum of forging stress exists below the α - γ transformation temperature. However within the temperature range $950^{\circ}C$ to $1150^{\circ}C$, the forging loads fell gradually. The trend of Huppmann's forging load vs. forging temperature curve agrees well with the present forging load curve which is shown in Fig. 5.1.

The forged billets were sectioned in such a way as indicated in Fig. 4.5. All the specimens were etched in 2% nital. Photographs of structures are shown in Figs. No. 5P (5.6).

Since the hot P/M forging of the iron-steel composite parts was carried out in the γ-iron temperature range of $950^{\circ}C$-$1200^{\circ}C$, as would be expected there was no significant differences in the microstructures. Furthermore, the furnace annealing of the specimens resulted in a somewhat similar, uniform structure.

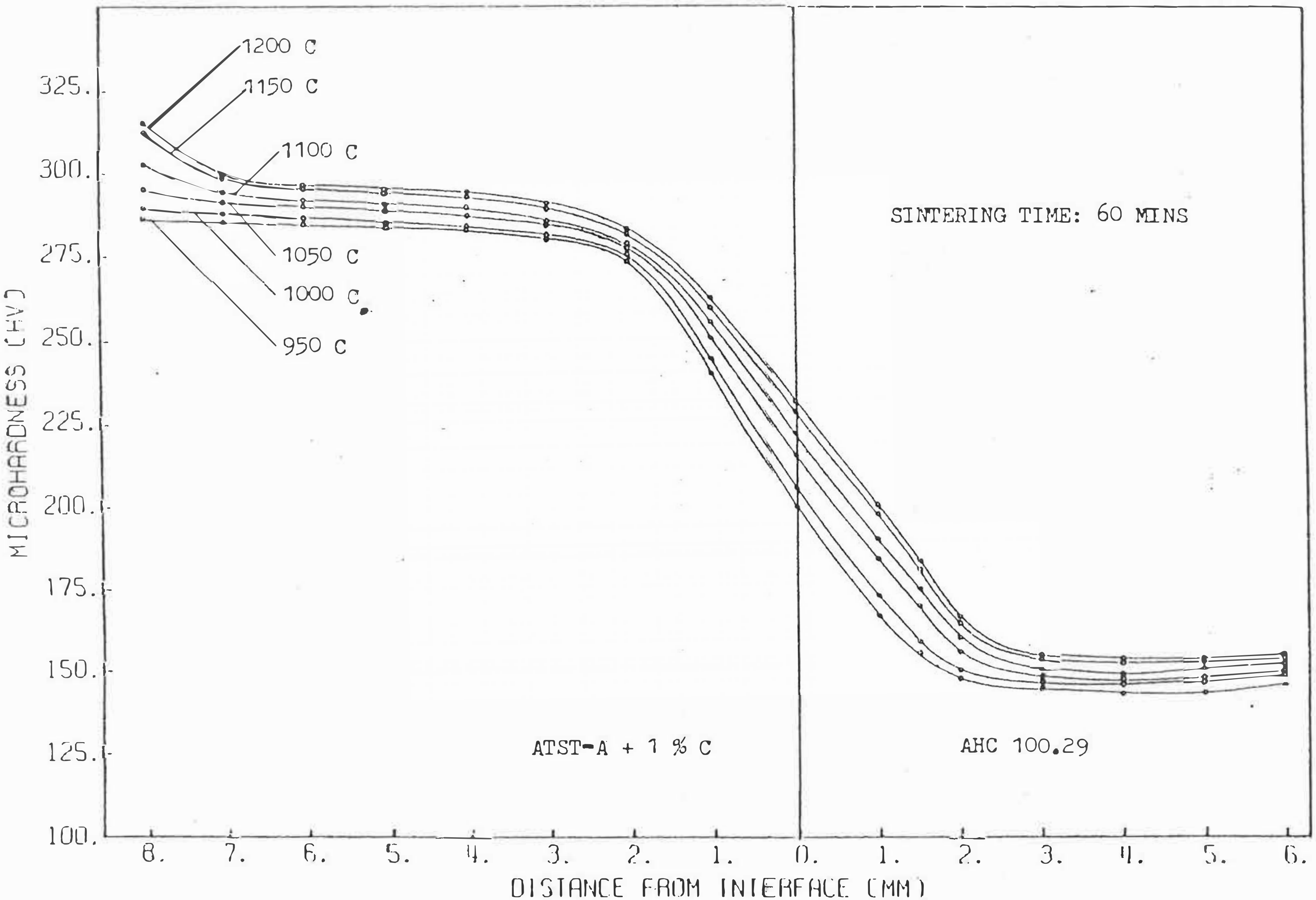


FIG. 5.2 EFFECT OF SINTERING & FORGING TEMPERATURE ON THE MICROHARDNESS VARIATION.

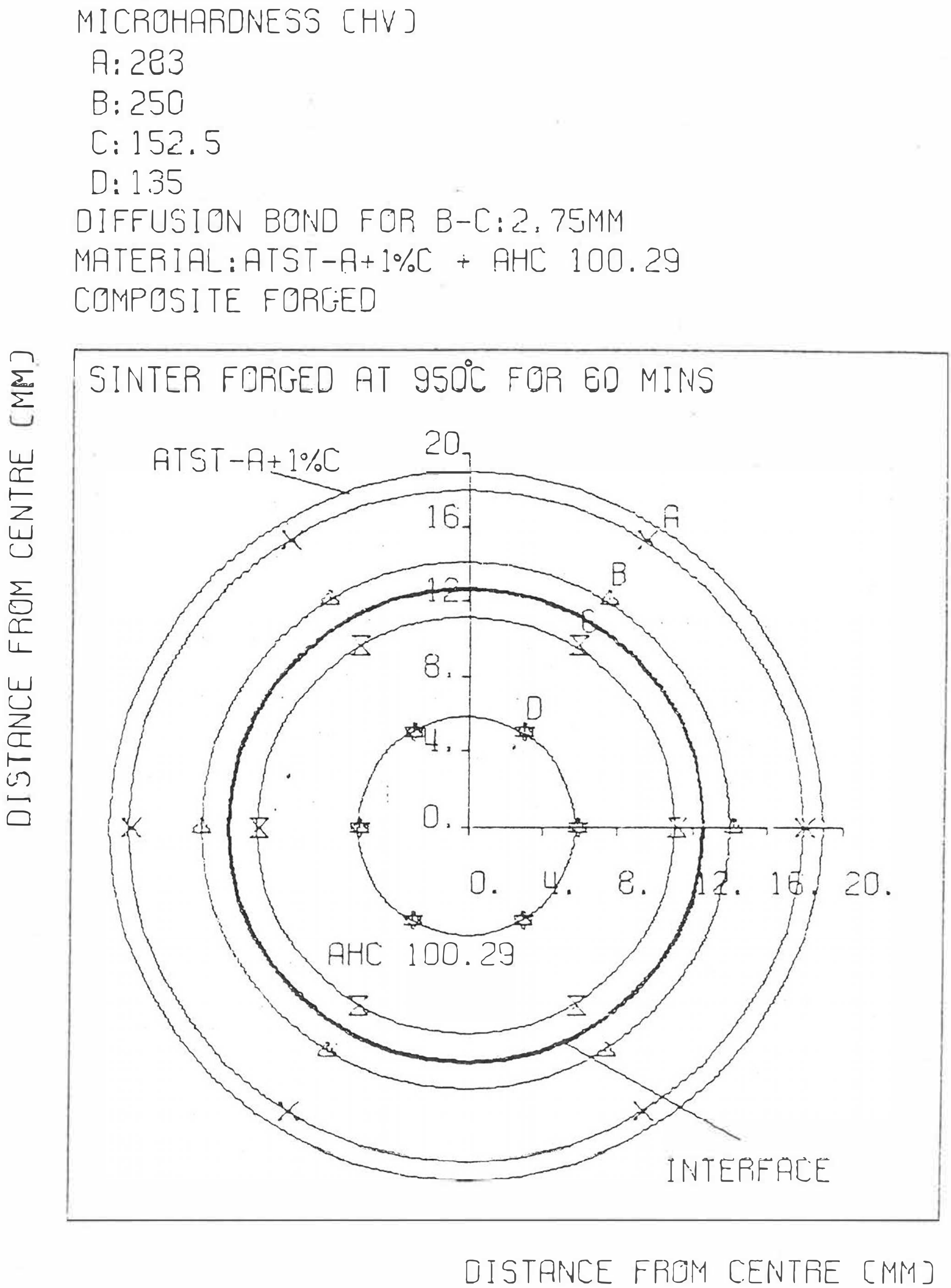


FIG. 5.3 EFFECT OF LOW SINTERING & FORGING TEMPERATURE ON THE MICROHARDNESS VARIATION FOR A CIRCUMFERENTIAL BOUNDARY TWO-LAYER MATERIAL.

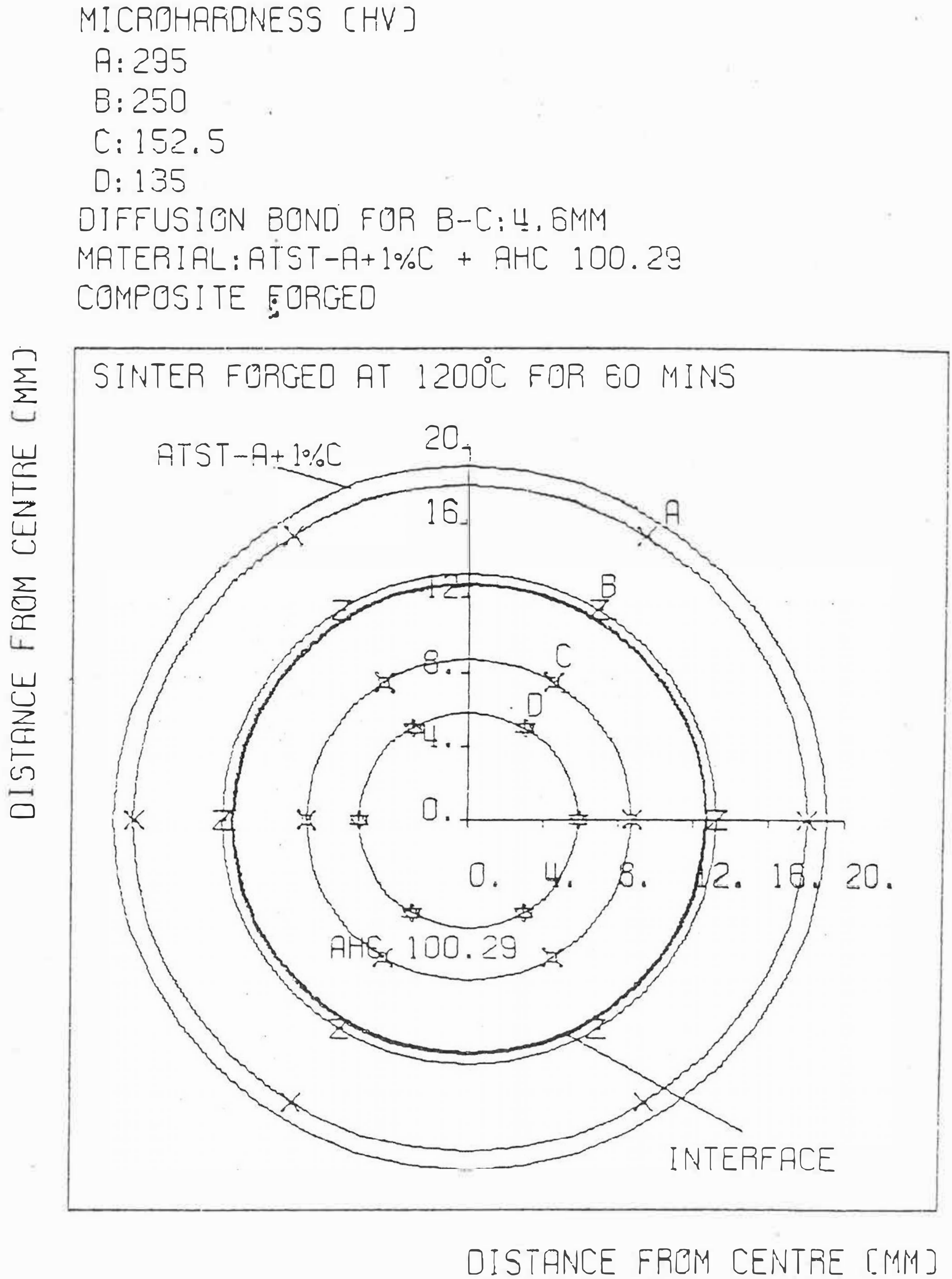


FIG. 5.4 EFFECT OF HIGH SINTERING & FORGING TEMPERATURE ON THE MICROHARDNESS VARIATION FOR A CIRCUMFERENTIAL BOUNDARY TWO-LAYER MATERIAL.

The microstructures are mainly ferrite with some carbides at the grain boundaries of iron part and lamellar pearlite + iron carbide in the steel part. At the boundary between the two materials, there was some evidence of diffusion of carbon from the steel to the iron part which increased in width as the temperature was increased from 950^{o}C and the time increased from 20 minutes.

It was also evident that annealing caused grain growth of both structures.

The microhardness variation curves of the iron-steel composite powder material are shown in Figs. 5.2, 5.3 and 5.4. Particular emphasis was given to the variation of hardness across the bond between the two materials. A sintering time of 60 minutes was chosen to investigate the effect of varying temperatures onthe microhardness. As can be seen from those curves, there was no significant difference in the microhardness of the steel part at the different sintering and forging temperatures.

But as the interface is reached a gradual change of microhardness values occurred, proceeding from the steel into the iron part. A limited region to the right and left of the interface was the critical part of the curves.

The effect of higher sintering temperatures on the diffusion was plainly visible. As the temperature was increased from 950^{o}C to 1200^{o}C, the change of hardness between the steel and iron parts became more gradual due

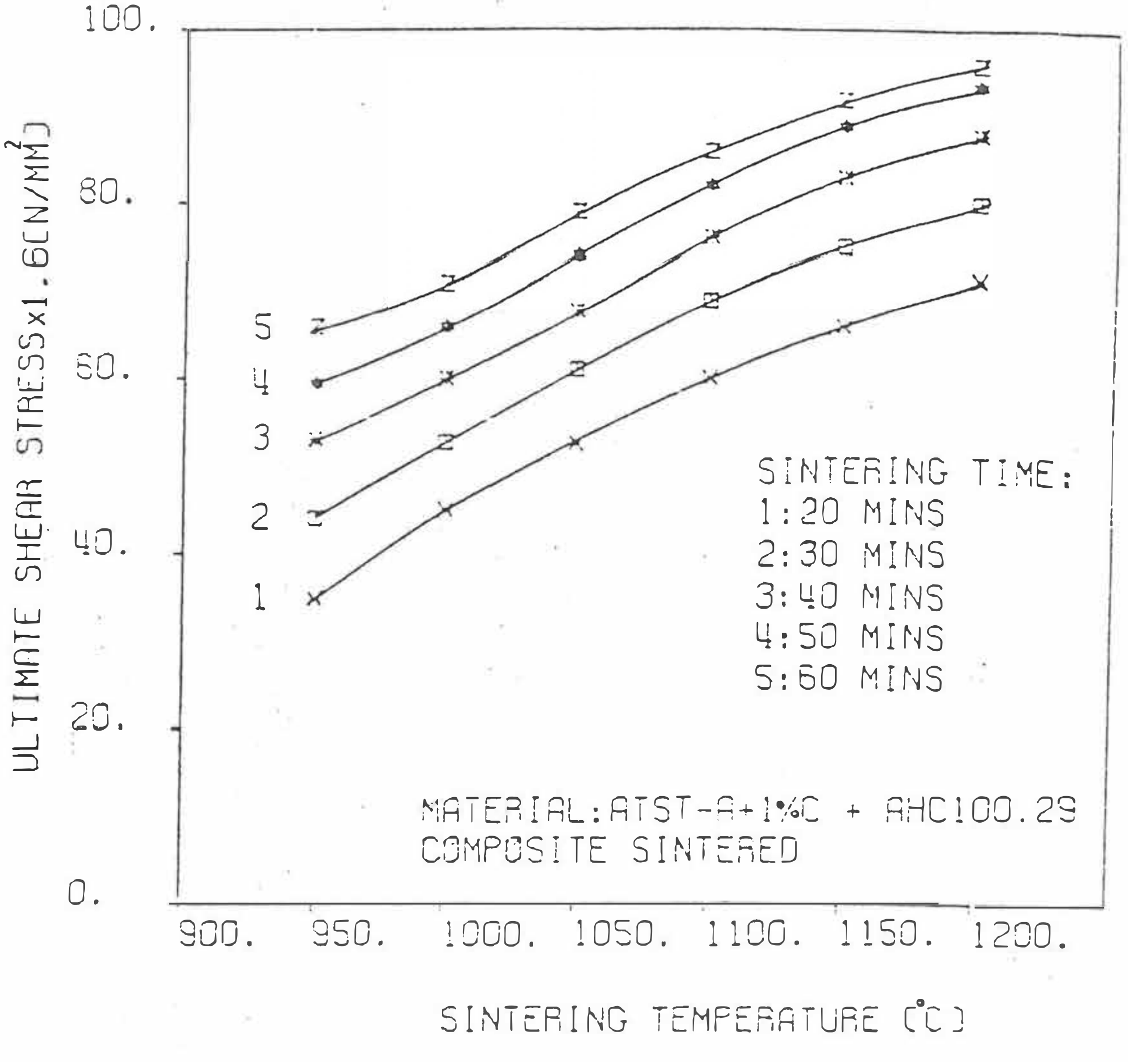


FIG. 5.5 EFFECT OF SINTERING TEMPERATURE ON THE COMPRESSIVE SHEAR STRESS OF AN IRON BASE SINTERED COMPOSITE MATERIAL.

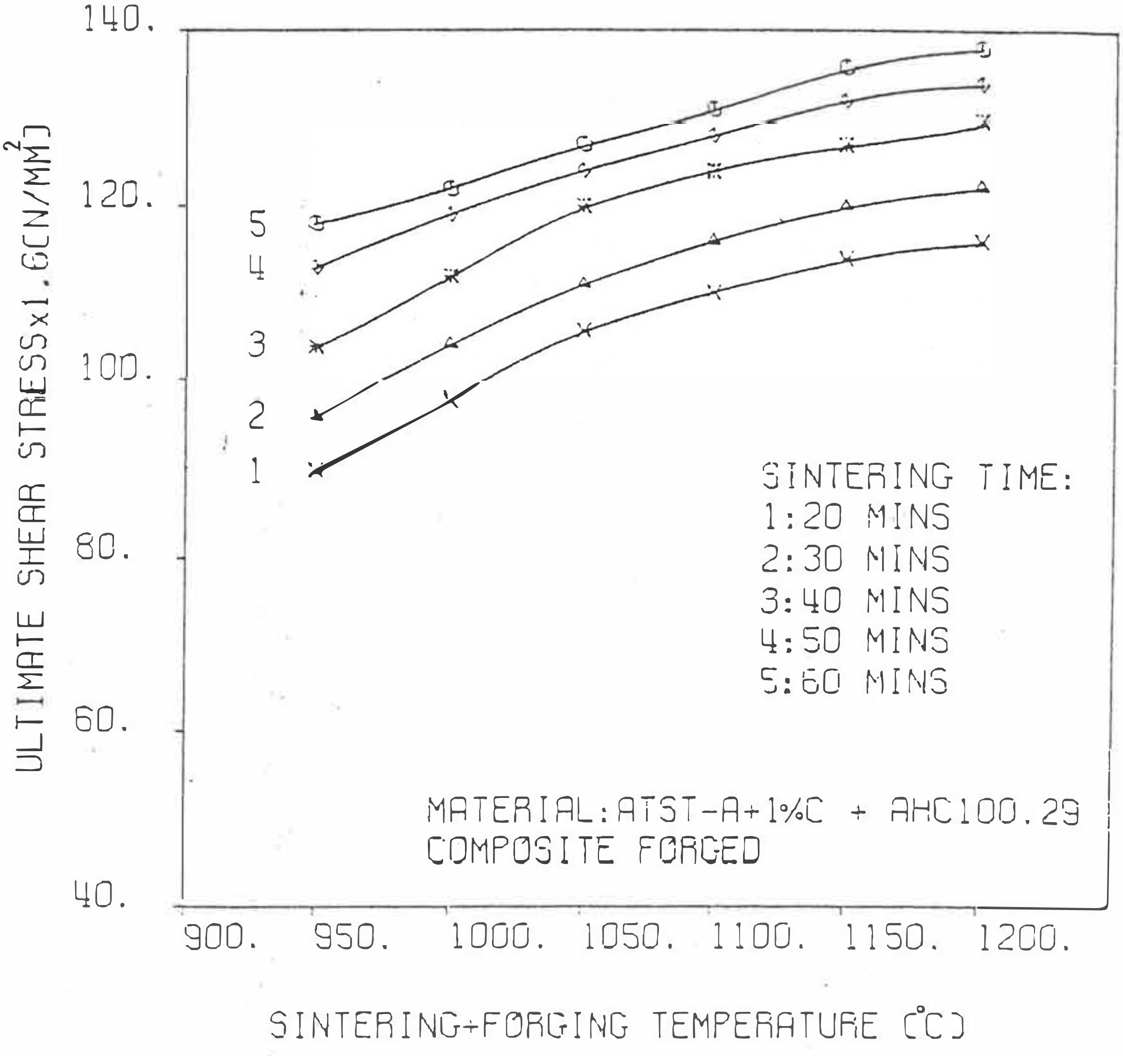


FIG. 5.6 EFFECT OF SINTERING & FORGING TEMPERATURE ON THE COMPRESSIVE SHEAR STRESS OF AN IRON BASE SINTER FORGED COMPOSITE MATERIAL.

to increased diffusion of carbon.

The effect of sintering temperature on the strength of the interface of the two part material was also pronounced as can be seen from the compressive shear test. (Fig. 5.5, 5.6). As a comparison the tests were carried out for both as-sintered and as-forged conditions. The as-forged specimens as expected had higher ultimate shear stress values than the as-sintered parts. In both cases ultimate shear stress values increased as the temperature increased from 950^{o}C to 1200^{o}C. For the different sintering times the regular increase in the ultimate shear stress values with increasing temperature was evident. The sets of shear stress curves became somewhat closer to each other particularly at the higher sintering temperatures.

The Charpy impact curves over the temperature range of 950^{o}C - 1200^{o}C are shown in Fig. 5.7, 5.8) for three different compositions of materials. Charpy impact values increased with increasing sintering and forging temperature. The Charpy impact test results that are shown in the figures for two-part steel-iron parts were compared with both plain iron and low alloy steel specimens. Plain iron specimens of AHC 100.29 atomized iron powder gave the highest Charpy impact values, and these sharply increased as the temperature rose from 950^{o}C to 1200^{o}C. For various sintering times, the curves had a similar characteristic trend. Previous Charpy impact tests (carried out by Huppmann (100)) for atomized iron powder-forged specimens were reported to

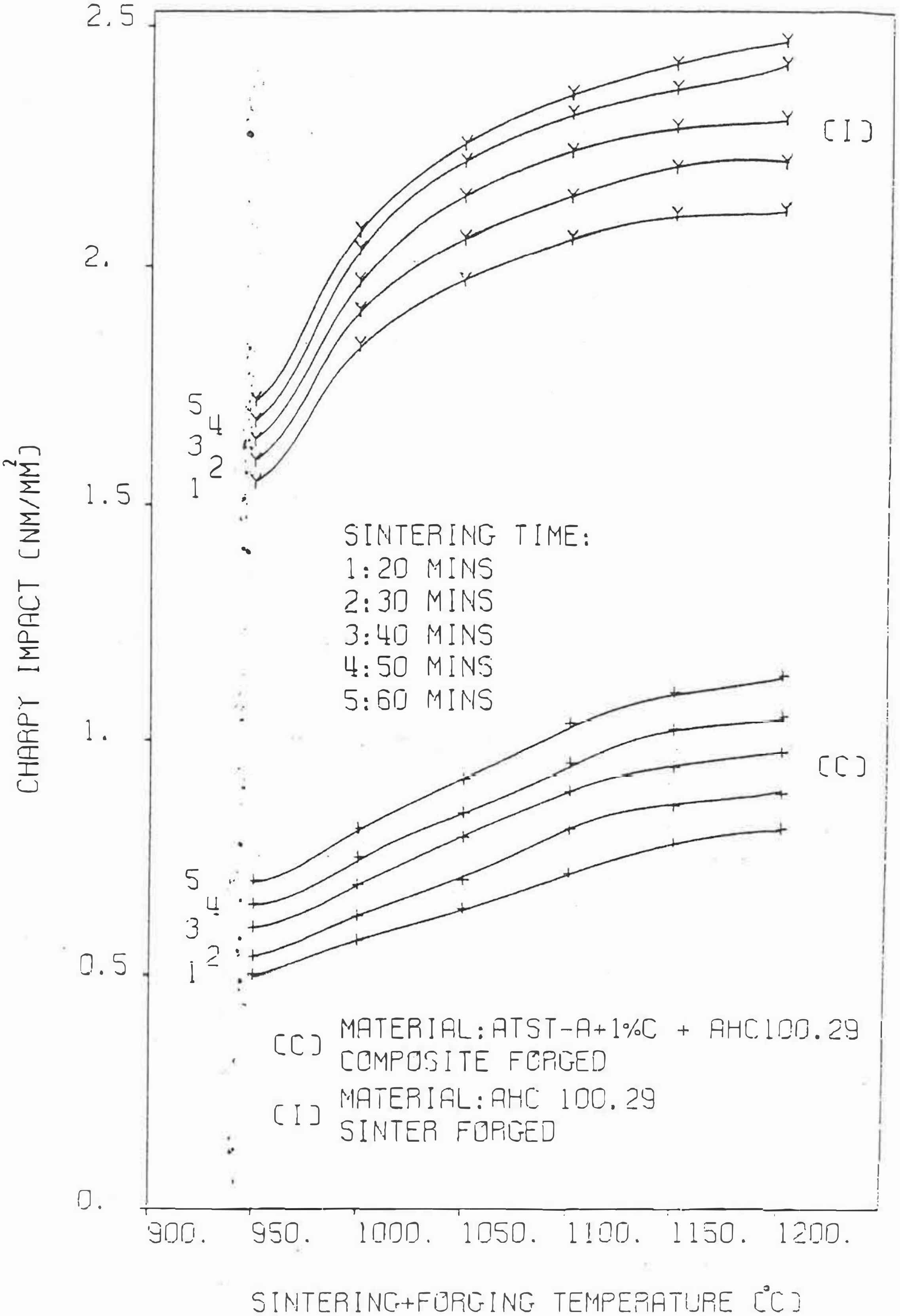


FIG. 5.7 EFFECT OF SINTERING & FORGING TEMPERATURE ON THE CHARPY IMPACT RESISTANCE OF THE IRON BASE MATERIALS.

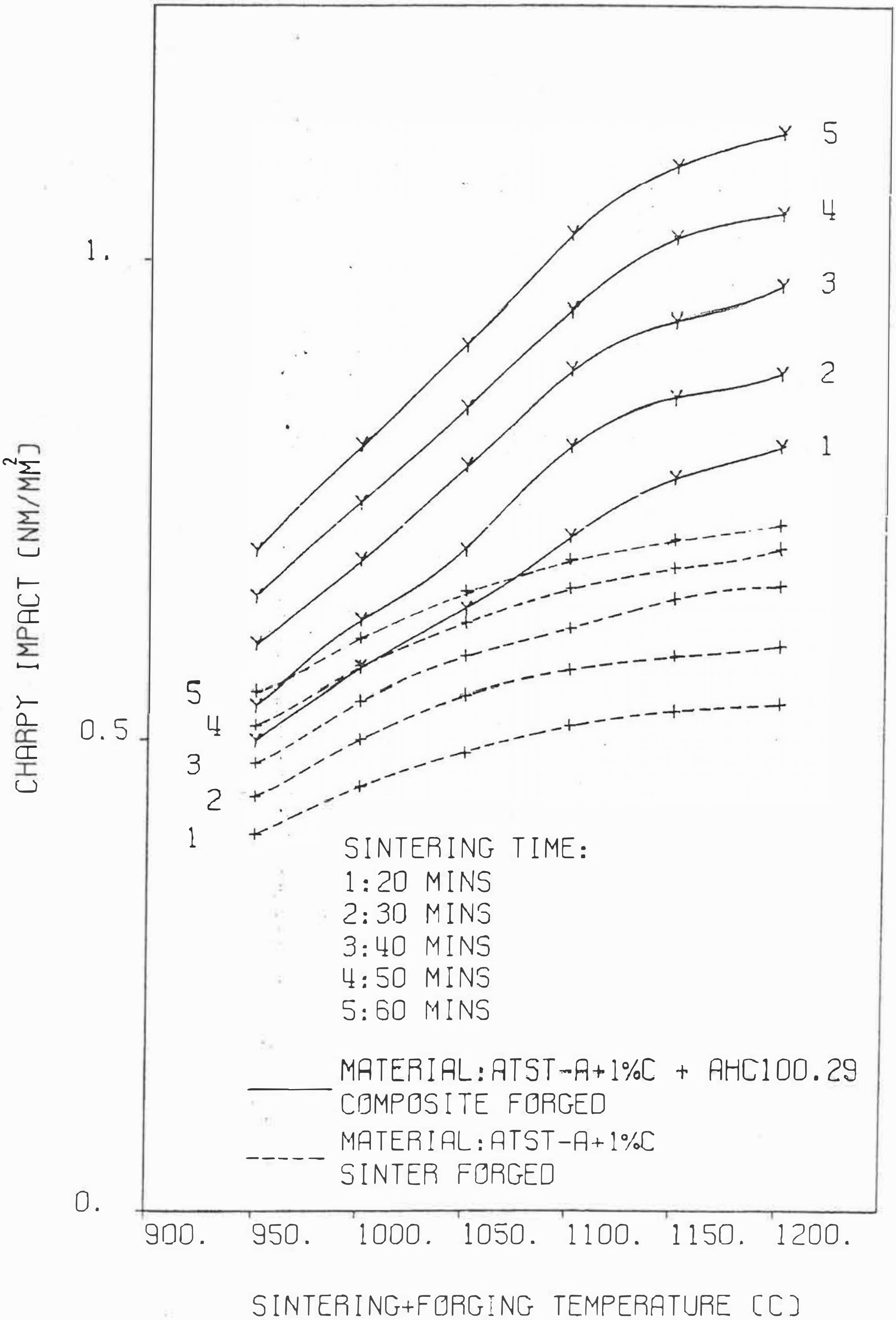


FIG. 5.8 EFFECT OF SINTERING & FORGING TEMPERATURE ON THE CHARPY IMPACT RESISTANCE OF THE IRON BASE MATERIALS.

have higher values than the reduced and spongy iron powders. In this work on the effect of powder characteristic on the sinter-forging process, NC.100.24 sponge iron powder was shown to have lower Charpy impact values than both RZ 150 air-atomized iron and WP type water atomized iron powders. These very low Charpy energies of NC.100.24 sponge iron powder were attributed to the brittle intergranular failure which seemed to originate at the large non-metallic inclusions of this powder. In the present work the higher impact values fo: the atomized AHC.100.29 iron powder agreed well with the earlier reports. Since the impact specimens were unnotched, the tests resulted in somewhat higher Charpy values than would have been obtained from the notched specimens.

The effect of increasing sintering temperature on the impact values was also mentioned by Wang (93). For a number of densities and sintering temperatures he carried out impact tests and found a sharp rise in the impact energy of MH.100.24 sponge iron powder for 7.5 g/cm^3 final density and sintering temperatures above 900°C. Sintering temperatures of 900°C and upwards caused an increase in impact values along with increasing temperature. Although the iron powder used in his experiments was of the MH.100.24 sponge type, the favourable effect of increasing temperature on the impact values was still evident.

It was significant that throughout the Charpy impact experiments none of the plain iron unnotched specimens broke indicating a high degree of ductility.

The Charpy impact values for ATST-A + 1% graphite specimens were considerably lower than for the plain iron specimens. Increasing temperature, caused a small increase of impact values. The lower impact values of the ATST-A + 1% graphite steel specimens may be atributed to the fact that relatively higher graphite content (1% graphite) rendered the material brittle and fairly sensitive to impact. Low alloy steel powder particles might have also entrapped some oxide inclusions during the production process, and as many other authors have reported the presence of manganase might have caused the material to be prone to the formation of manganase oxide which is very difficult to reduce under the normal hot working conditions, particularly at temperatures lower than 1200 $^{\circ}$C. There are some reports indicating the fact that manganase oxide cannot be reduced satisfactorily unless higher temperatures, (> 1200°C) are reached. Crowson et. al. (42) found that the oxide content of manganase is quite difficult to reduce below 1205°C and can adversely affect P/M steel forging properties. In order to reduce this oxide it is necessary to sinter at higher temperatures (1260°C and 1315°C) over longer periods of time. References (40) and (41) also refer to the detrimental effects of the unreduced oxides on the fracture resistance of the P/M forged materials. The effect of increasing carbon content on the impact strength of sinter-forged low alloy steels was investigated by Cundill et. al (30). They found that the impact resistance values of the SAE 4600 Charpy V-notch samples decreased with increased carbon contents. Their

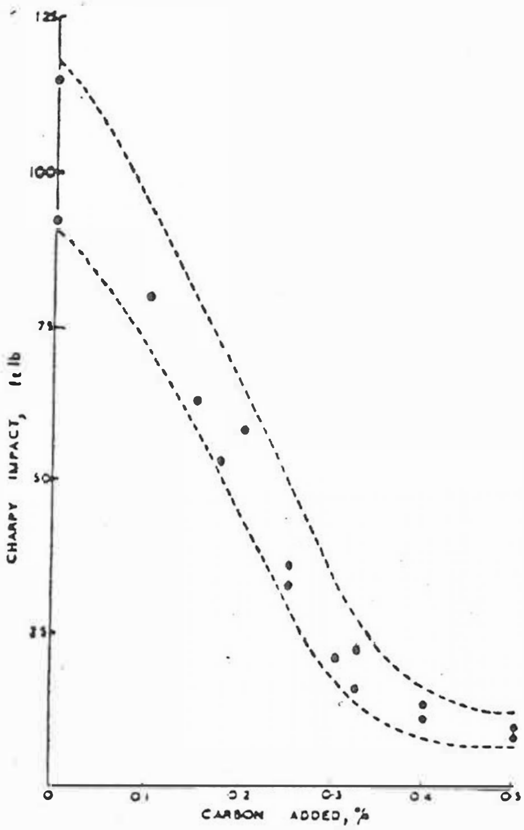


FIG. 5.9 CHARPY V-NOTCH IMPACT-RESISTANCE OF AIR-COOLED SAE 4600 ALLOYS.(30)

results are shown in fig. 5.9 for air cooled samples, where it can be seen that impact values of 70-115 ftlb (12 - 20 da J/cm^2) were obtained from low carbon alloys, but there was a marked decrease with increasing carbon additions. The effect of sintering or forging temperature on the impact strength of iron based hot forged materials were studied by a number of authors. Increased sintering or forging temperatures increased the impact strength of the materials, mainly because of the reduction in the total oxide content. Moyer (101) Cook (44) and Hanejko (102) all reached the same conclusion regarding the beneficial effect of increased sintering and forging temperature on the impact properties.

Plain iron-ATST-A + 1% graphite two-part composite powder forged specimens showed impact values somewhere between the plain iron and low alloy steel + 1% graphite specimens. This was expected because of the nature of the specimens which were a combination of both brittle ATST-A + 1% graphite and ductile plain iron AHC.100.29 specimens. It was also noted that none of the specimens broke at the joint (interface). That was an indication that the joint had a considerable impact strength.

The results of the temperature range tests on ultimate tensile strength (UTS), percentage elongation and percentage reduction in area are shown in Fig. 5.10, 5.11 (a,b) and 5.12, for the three different compositions of materials.

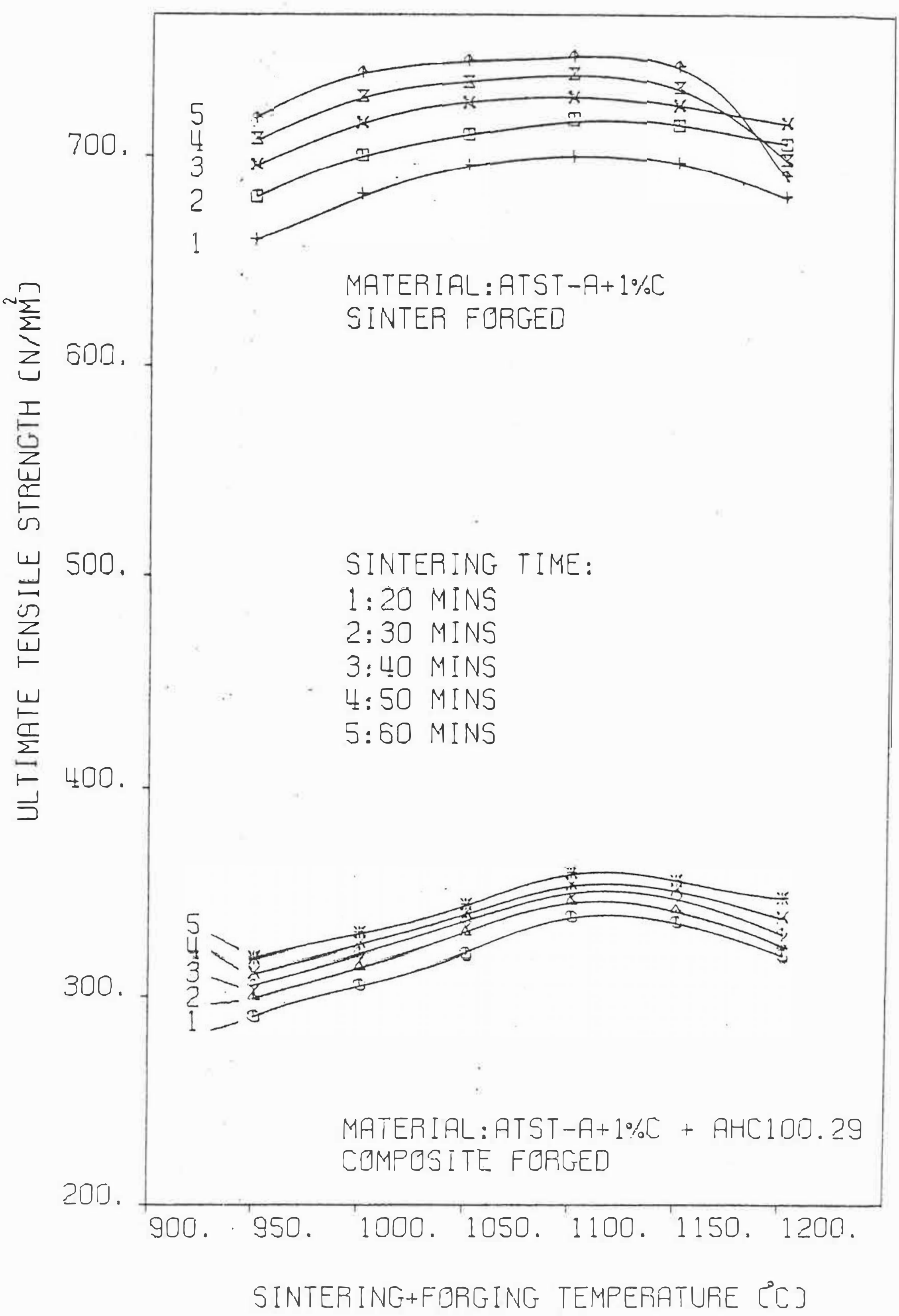


FIG. 5.10 VARIATION OF UTS WITH SINTERING & FORGING TEMPERATURE.

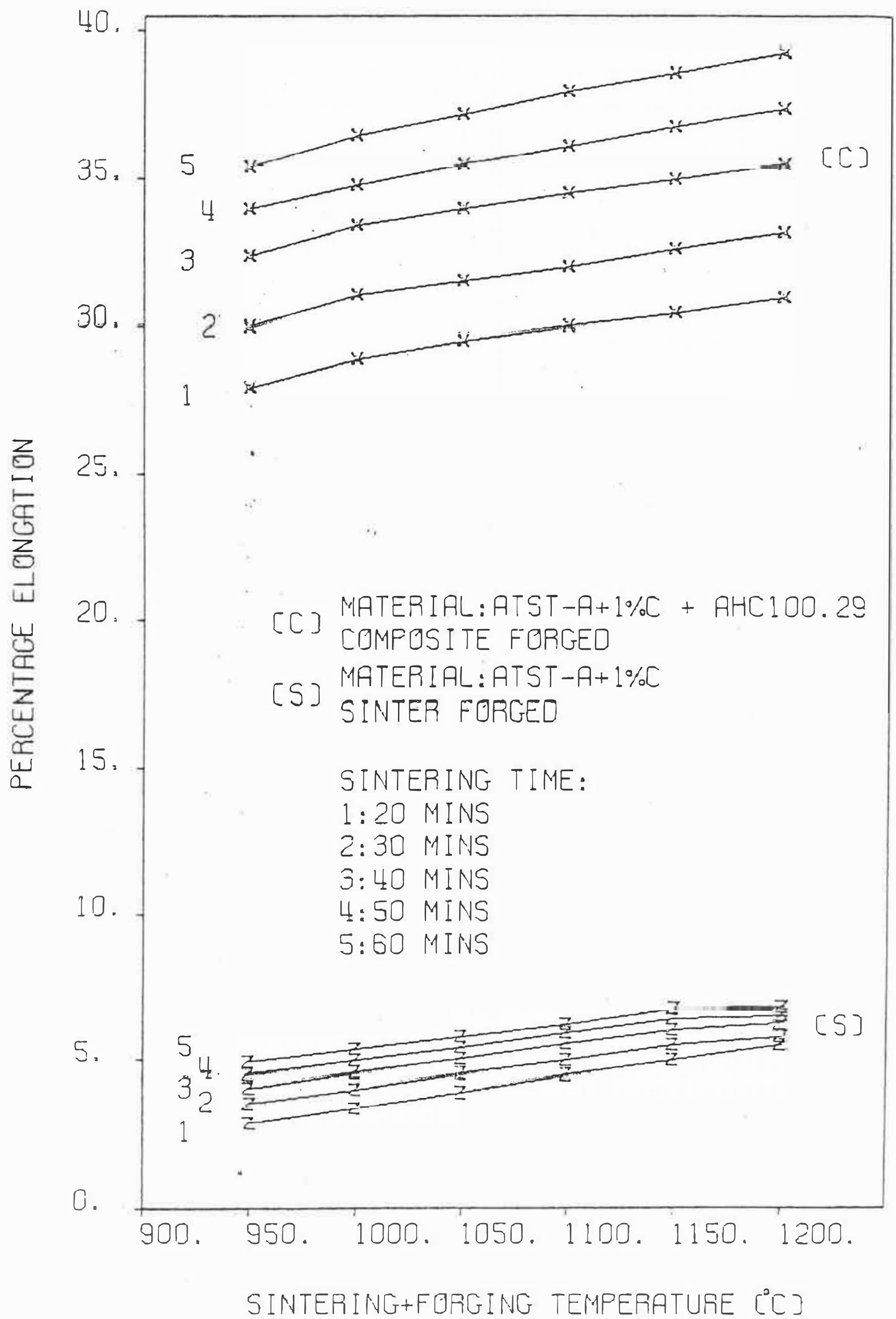


FIG. 5.11.a VARIATION OF PERCENTAGE ELONGATION WITH SINTERING & FORGING TEMPERATURE.

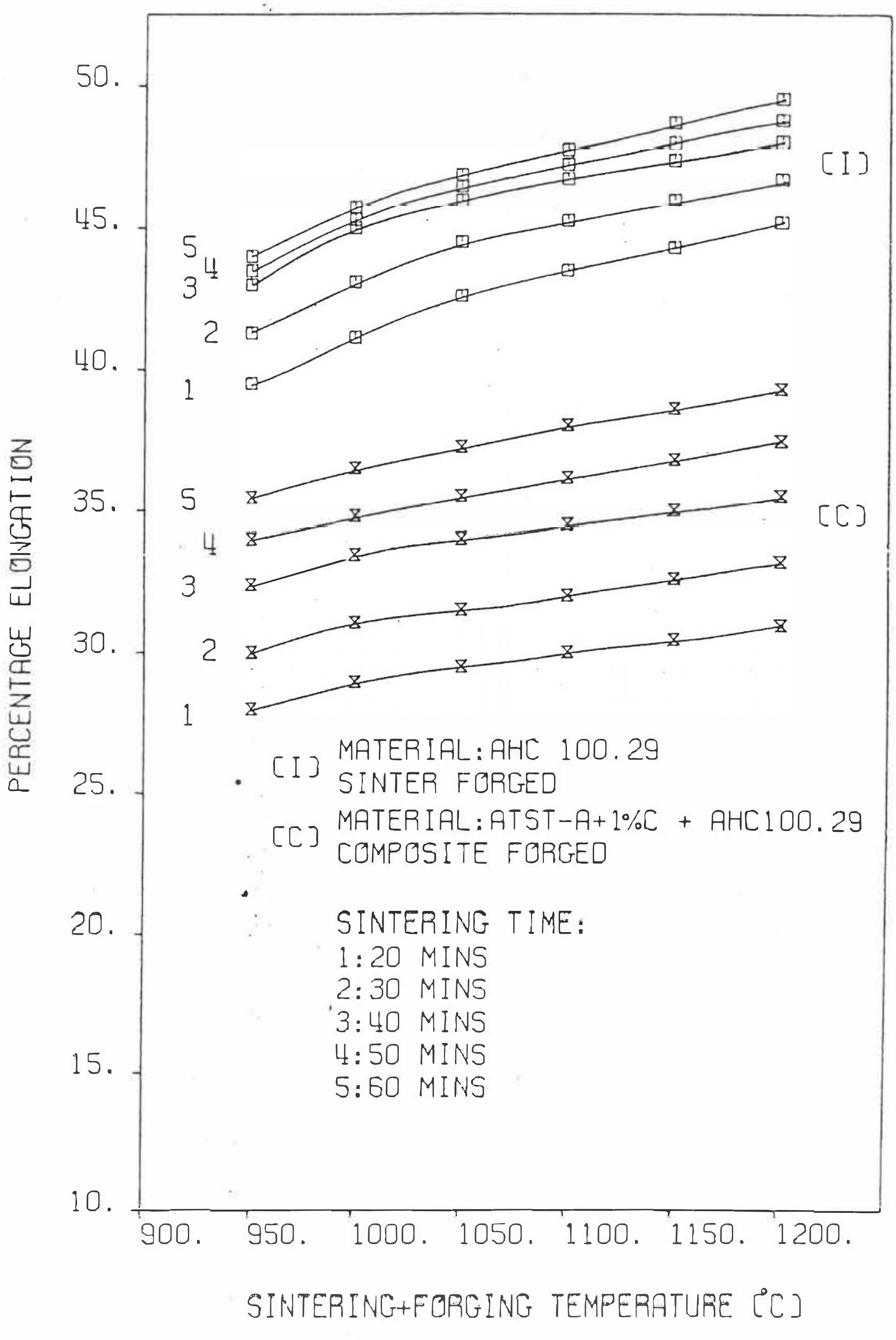


FIG. 5.11.b VARIATION OF PERCENTAGE ELONGATION WITH SINTERING & FORGING TEMPERATURE.

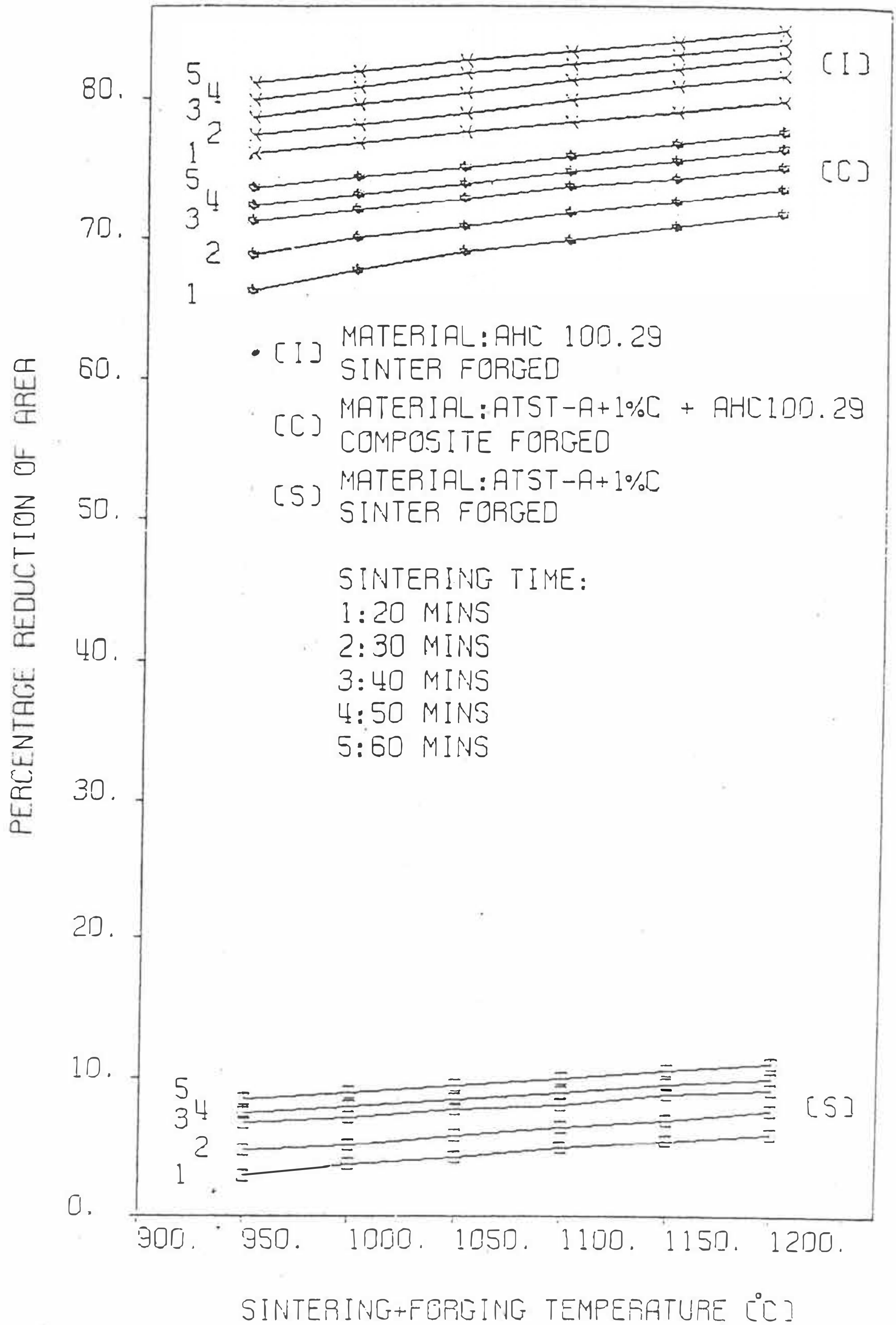


FIG. 5.12 VARIATION OF REDUCTION OF AREA WITH SINTERING & FORGING TEMPERATURE.

Since the UTS curves for plain iron and iron-steel composite parts gave very close results, only the curves for composite specimens are shown in the same graph together with a comparison with the low alloy steel specimens. As is seen from the curves, composite specimens had a tensile strength of just over 300 N/mm^2, which were very close to the values for the plain iron specimens. In the temperature range 950^oC-1100^oC, tensile strength increased slightly and above 1100^oC a gradual fall was observed. Increasing temperature in the 950^oC - 1100^oC range produced better bonding of particles, whilst increased grain growth would probably cause the tensile strength to fall gradually above a temperature of 1100^oC. One other reason for the increased tensile strength between 950^oC - 1100^oC could be the presence of oxidized grains which lead to an increase of tensile strength. Ref. (103). Above 1100^oC, evolution of entrapped gases could also cause a fall in the tensile strength. Ref. (103).

UTS curves for low alloy steel specimens followed a similar pattern as for the composite specimens. UTS values for low alloy steel specimens were more than twice those of both composite and plain iron specimens. Above 1100^oC sintering and forging temperatures the UTS values gradually decreased particularly for 50 and 60 minutes sintering times at 1200^oC. This was probably due to the effect of decarburisation of high C low alloy steel specimens of higher temperatures and longer sintering times.

From the S.E.M. fractographs two distinct fracture types appeared; 1. Ductile fracture of iron parts, 2. Brittle fracture of steel parts.

Ductile fracture is defined [104] as 'dimpled rupture' fracture which consists of the growth and coalasence of voids by plastic flow. The main features of dimpled rupture surfaces is the presence of numerous smooth rounded concave depressions on the surface.

Brittle fracture is defined as 'Cleavage fracture' or transgranular fracture. The cleavage surface fracture has certain distinctive markings, namely 'river markings' and 'tongues'.

Composite and plain iron specimens showed a typical ductile fractures with regular necking, whilst the low alloy steel specimens showed evidence of brittle fracture. Scanning electron microscope (S.E.M.) tensile fractographs of the three different types of specimens are shown in Figs 5P (2,3). Thecomposite and plain iron specimens showed typical ductile fracture with a network of fine dimples. Low alloy steel specimens which have a reasonably high carbon content showed brittle fracture without necking. This is also seen from the S.E.M. fractographs.

Since the composite parts consisted of brittle and ductile materials, it was interesting to note that all the specimens broke in the iron part with necking and not at the joint (interface).

The effects of temperature on the percentage elongation and percentage reduction of area for three different composition of materials are shown in Fig. 5.11 (a,b) and 5.12.

As is seen from the Fig. 5.11 (a,b) percentage elongation increased very little with increasing temperature for the high carbon content (1% graphite) low alloy steel samples and remained lower than the composite and plain iron specimens.

Percentage elongation increased with increasing temperature for the both composite and plain iron parts, which gave higher percentage elongation values than the composite parts.

Fig. 5.12 shows that as would be expected the percentage reduction of area results followed a similar trend to the percentage elongation results. Low alloy steel specimens showed little increase in percentage reduction of area with increasing temperature and gave lower values than composite and plain iron parts. Composite and plain iron parts showed increasing percentage reduction of area with increasing temperature.

In Fig. 5.13 and 5.14, the effect of increasing C content on the tensile properties of SAE 4600 sinter forged material are shown (30) SAE 4600 has a similar composition ATST-A powder. From these curves it is also evident that high values of added C content causes an increase in UTS but a decrease in percentage elongation and percentage reduction of area.

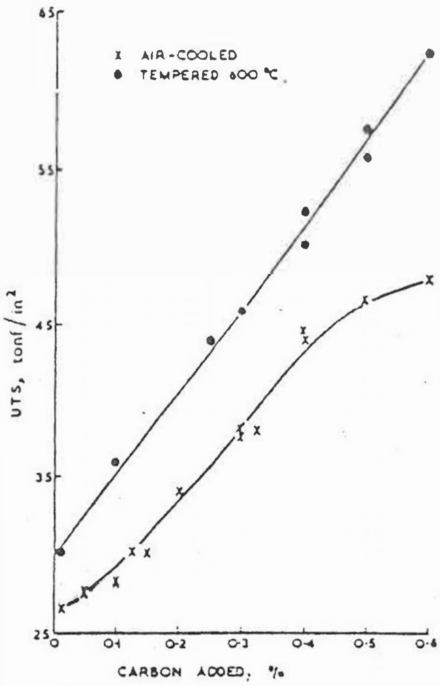


FIG. 5.13 VARIATION OF TENSILE STRENGTH WITH NOMINAL CARBON CONTENT FOR SAE 4600 ALLOYS.(30)

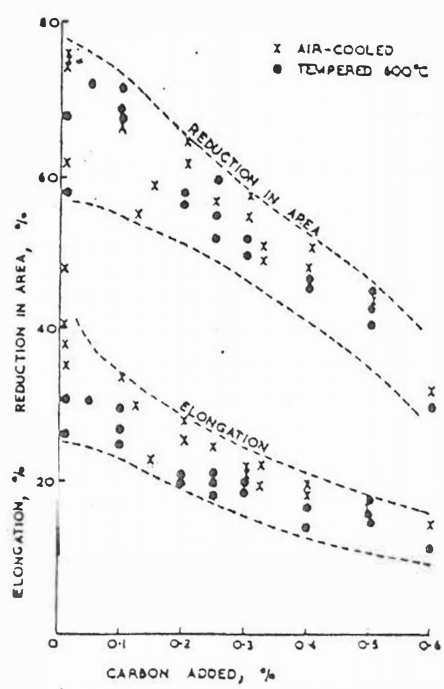


FIG. 5.14 TENSILE DUCTILITY OF SAMPLES MADE FROM SAE 4600 POWDER.(30)

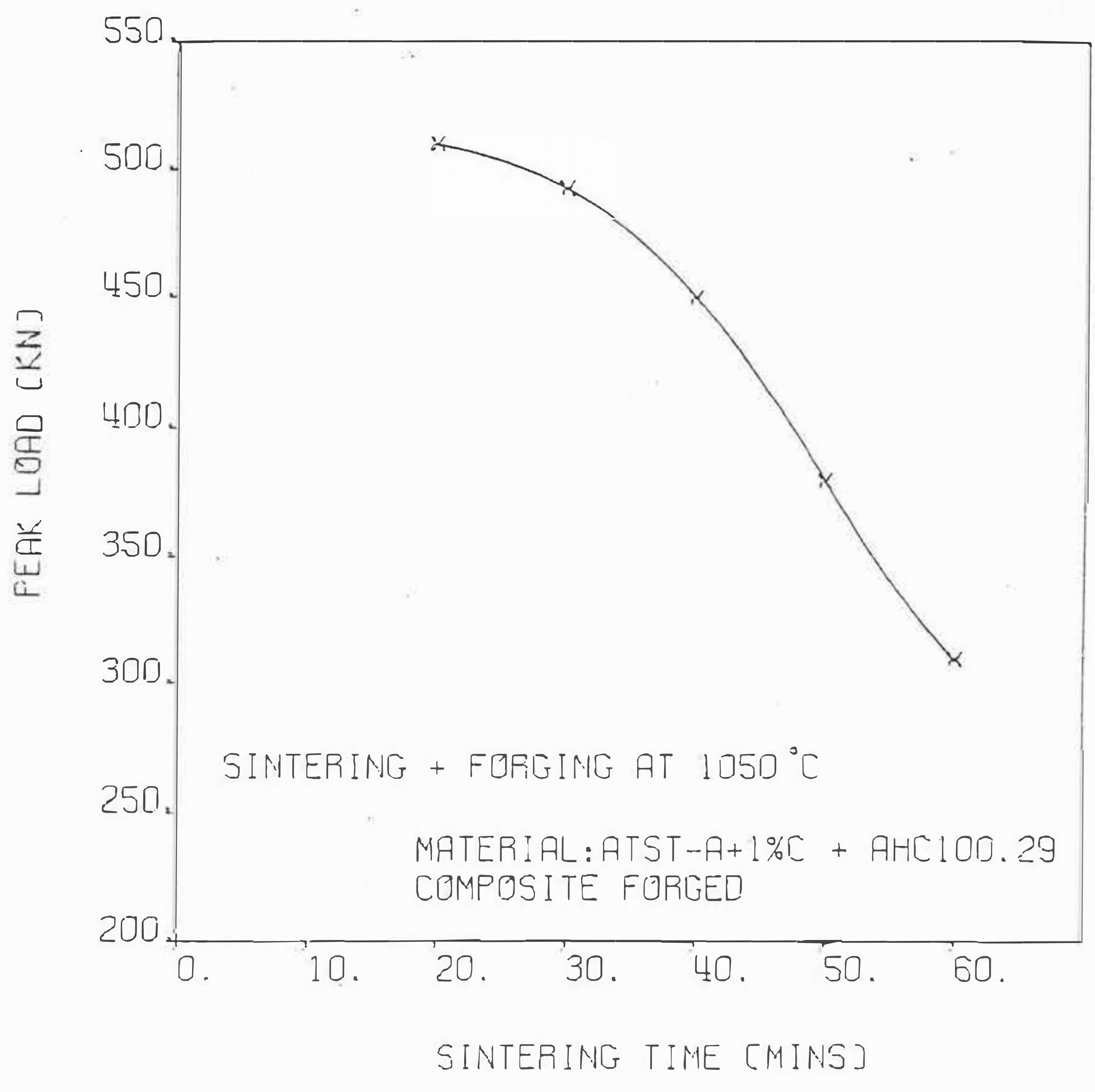


FIG. 5.15 EFFECT OF SINTERING TIME ON THE VARIATION OF PEAK LOAD.

5.3 The Results of Time Range Tests:

The variation of peak load with sintering time is shown in fig. 5.15. In the time range tests for peak load measurements, a constant sintering temperature of 1050°C was used. Sintering times were varied from 20 minutes up to 60 minutes in increments of 10 minutes. For the short sintering times of 20 to 40 minutes, the reduction of forging load with time was gradual, but increased markedly when the time was increased further. The increase in sintering time also reduced the iron base materials resistance to deformation as did increasing temperature.

The variation of microhardness for both low and high temperatures are shown in Figs. 5.16, 5.17, 5.18 and 5.19. For a low sintering temperature of 1000°C the sintering times were between 20 and 60 minutes in 10 minute increments. In the time range tests the hardness showed no significant change for the steel part of the composite iron-steel material as was seen in the temperature range tests. The fall in microhardness values between steel and iron was more gradual as the sintering time was increased from 20 minutes due to carbon diffusion. Microhardness variation for the higher sintering temperature of 1150°C is shown in fig. 5.17 and 5.19. The effect of increasing sintering time on the width of the diffusion bond was found to be more pronounced at higher sintering temperature, compared with the lower sintering temperature. At the higher sintering temperature of 1150°C, longer sintering times caused the amount of diffusion and the strength of

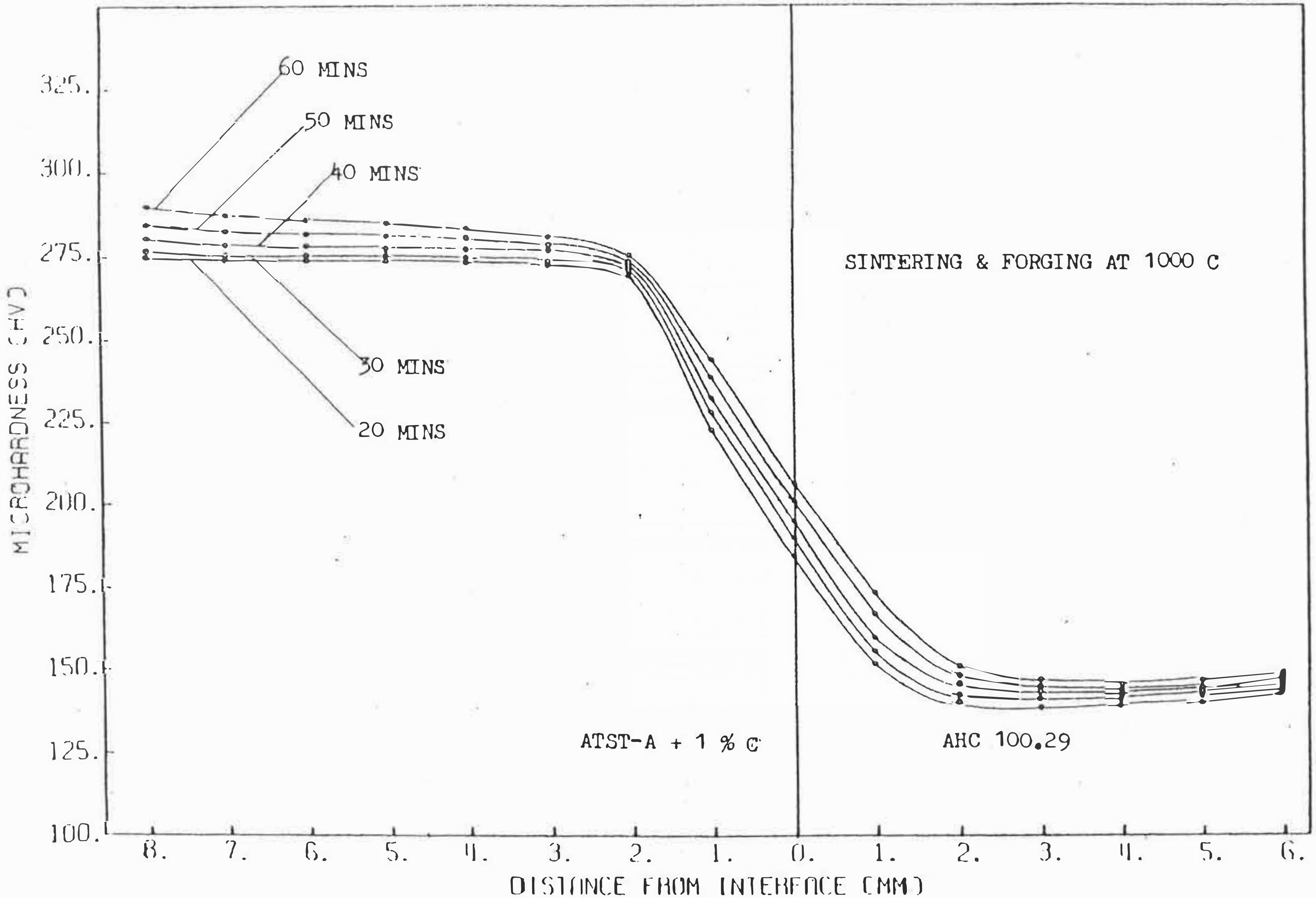


FIG. 5.16 EFFECT OF SINTERING TIME AT LOW TEMPERATURE ON THE MICROHARDNESS VARIATION.

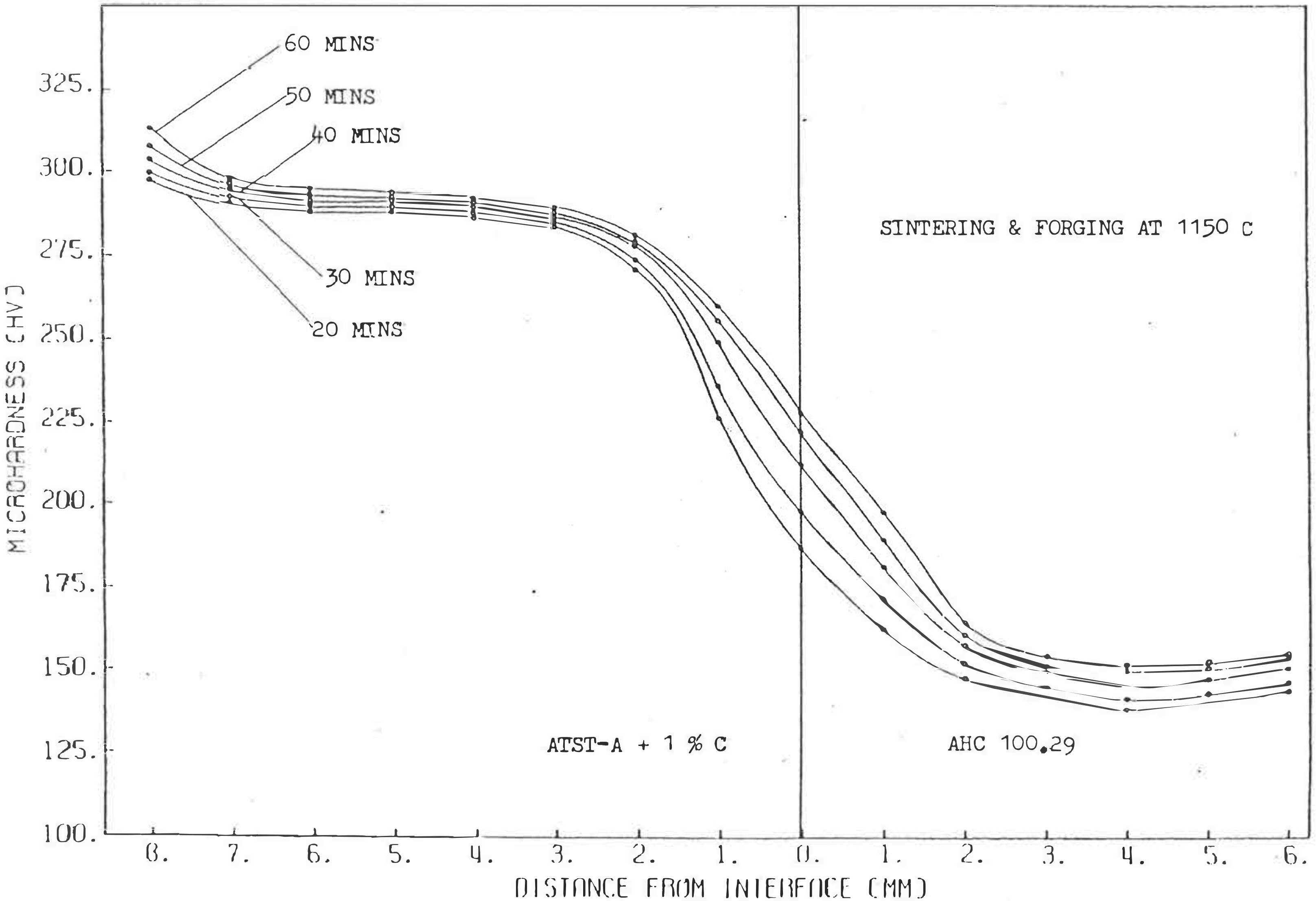


FIG. 5.17 EFFECT OF SINTERING TIME AT HIGH TEMPERATURE ON THE MICROHARDNESS VARIATION.

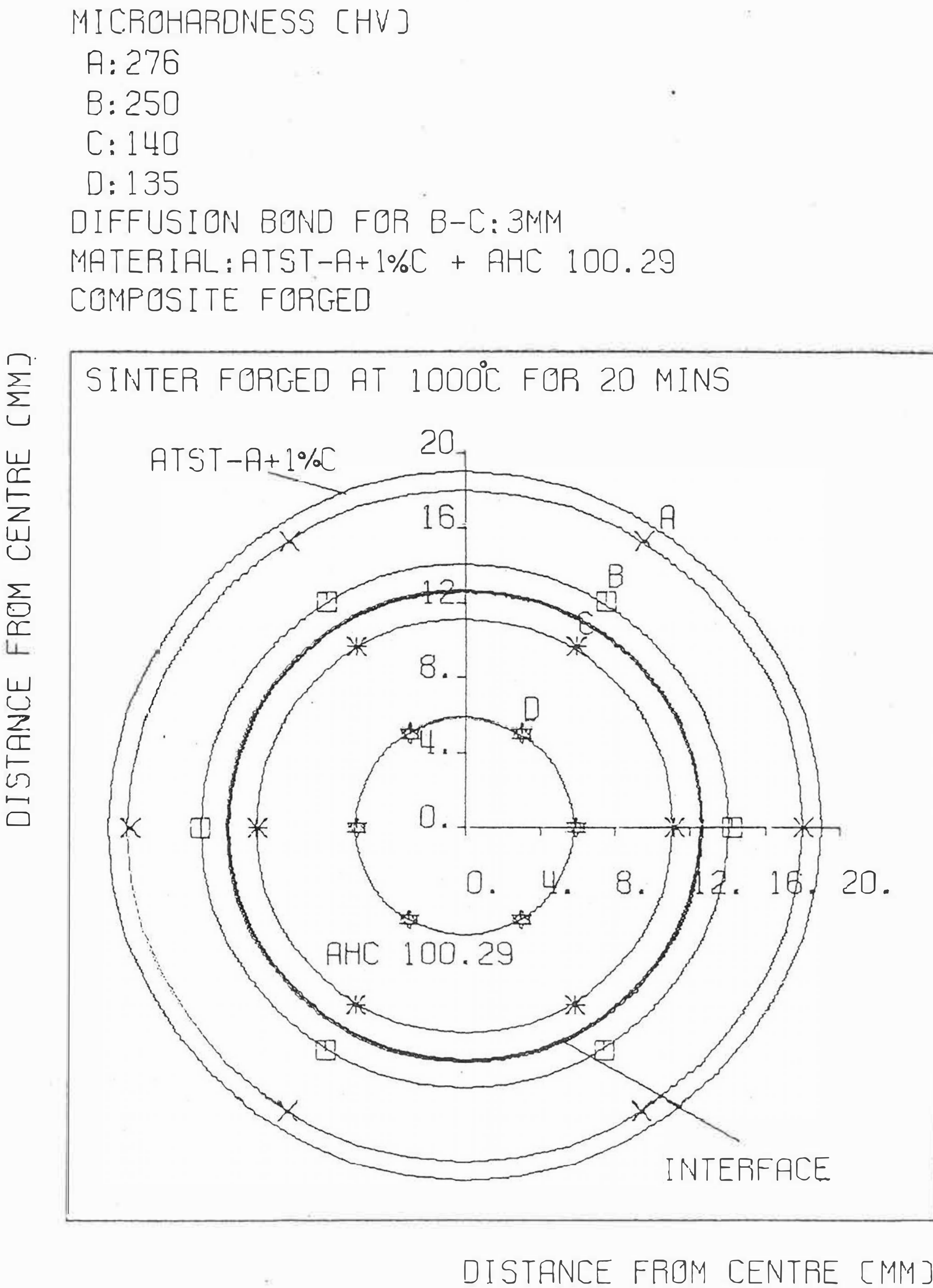


FIG. 5.18 EFFECT OF SINTERING TIME AT LOW TEMPERATURE ON THE MICROHARDNESS VARIATION FOR A CIRCUMFERENTIAL BOUNDARY TWO-LAYER MATERIAL.

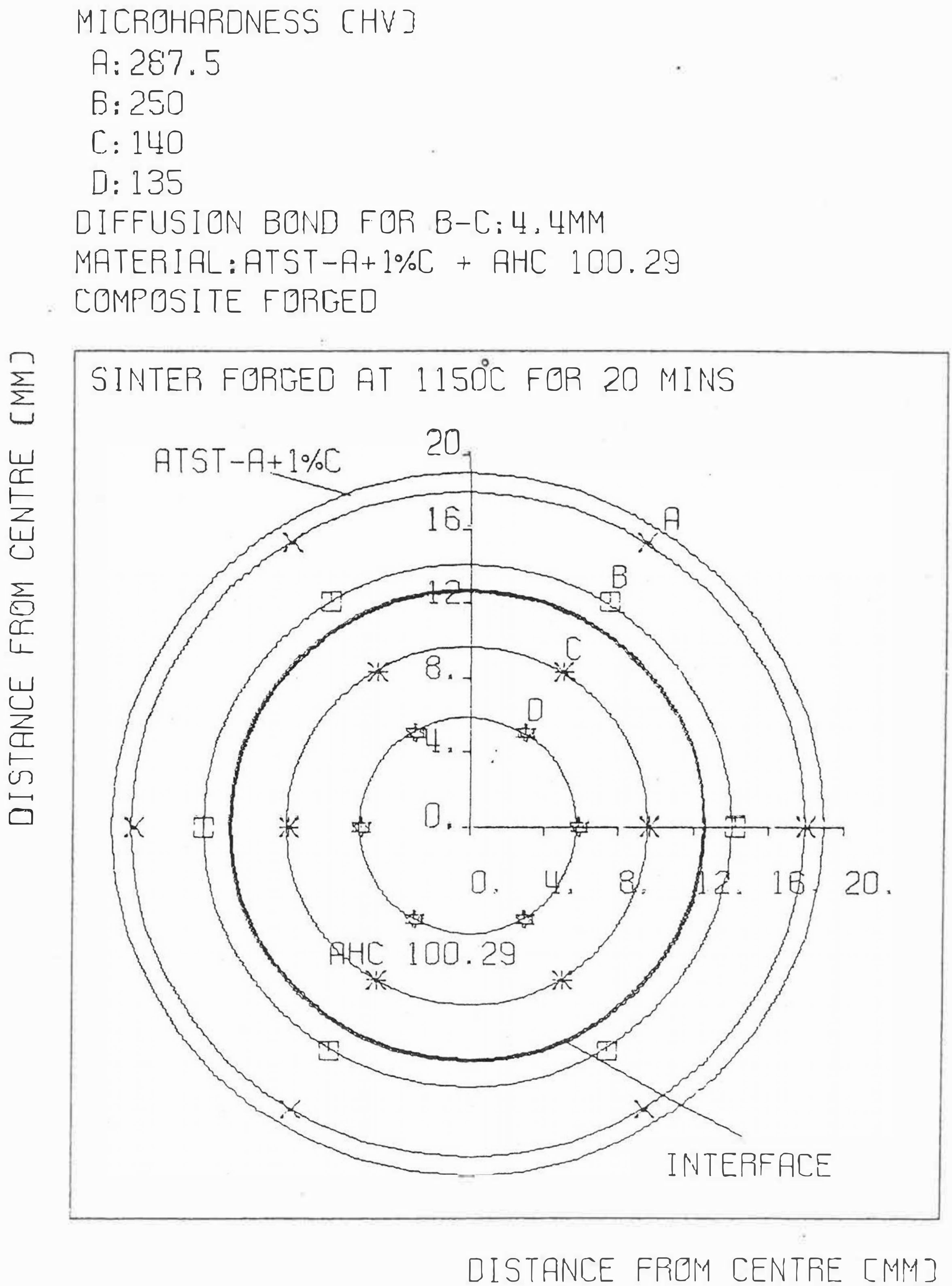


FIG. 5.19 EFFECT OF SINTERING TIME AT HIGH TEMPERATURE ON THE MICROHARDNESS VARIATION FOR A CIRCUMFERENTIAL BOUNDARY TWO-LAYER MATERIAL.

the bond to increase as the sintering times were increased from 20 minutes.

The effect of increased sintering times on the strength of bond of the two part powder composite material followed a similar trend as for the sintering temperatures. In the time range tests the ultimate shear stress tests were carried out for both as-sintered and as-forged conditions. As expected the as-forged specimens again had higher ultimate shear stress values than the as-sintered parts. Fig. (5.20, 5.21) Increased sintering times caused a rise in ultimate shear stress values.

The Charpy impact curves are shown in fig. 5.22 (a,b) for the time range of 20 minutes to 60 minutes for plain iron AHC.100.29 powder, low alloy steel of ATST-A + 1% graphite and composite iron-steel powder forged specimens. Increased sintering times caused an increase in the Charpy unnotched impact values. In the time range tests the highest values of Charpy impact values were obtained with the plain iron AHC.100.29 specimens; this was attributed to the ductile, impact absorbing nature of the plain iron. Unnotched impact specimens of atomized plain iron resulted in somewhat higher results than would be found with notched specimens. The curves of Charpy impact values versus time followed a fairly linear trend except at higher sintering times and higher sintering temperatures. In the time range tests, again none of the plain iron unnotched impact specimens was broken. For 1050^{o}C and

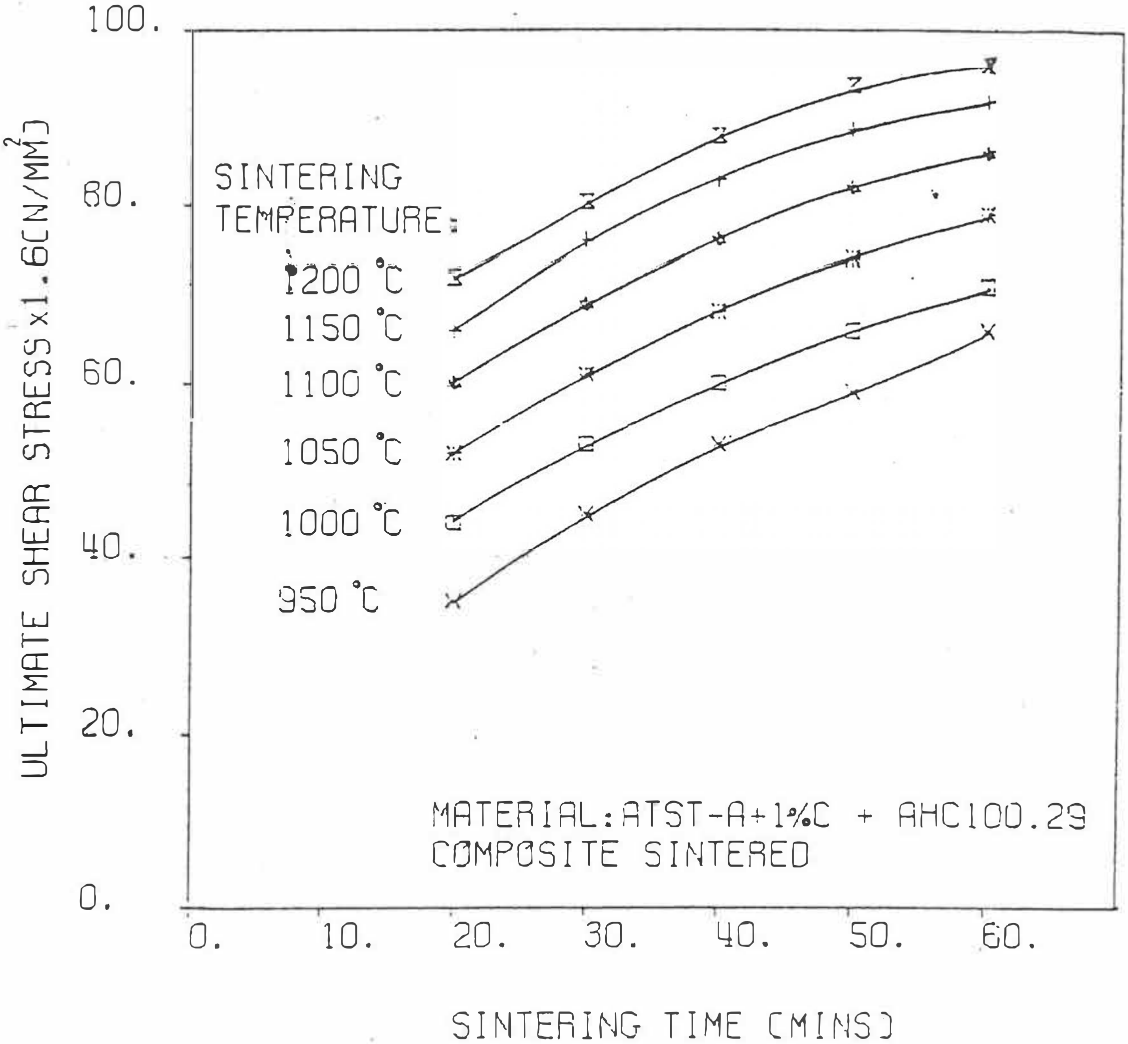


FIG. 5.20 EFFECT OF SINTERING TIME ON THE COMPRESSIVE SHEAR STRESS OF AN IRON BASE SINTERED COMPOSITE MATERIAL.

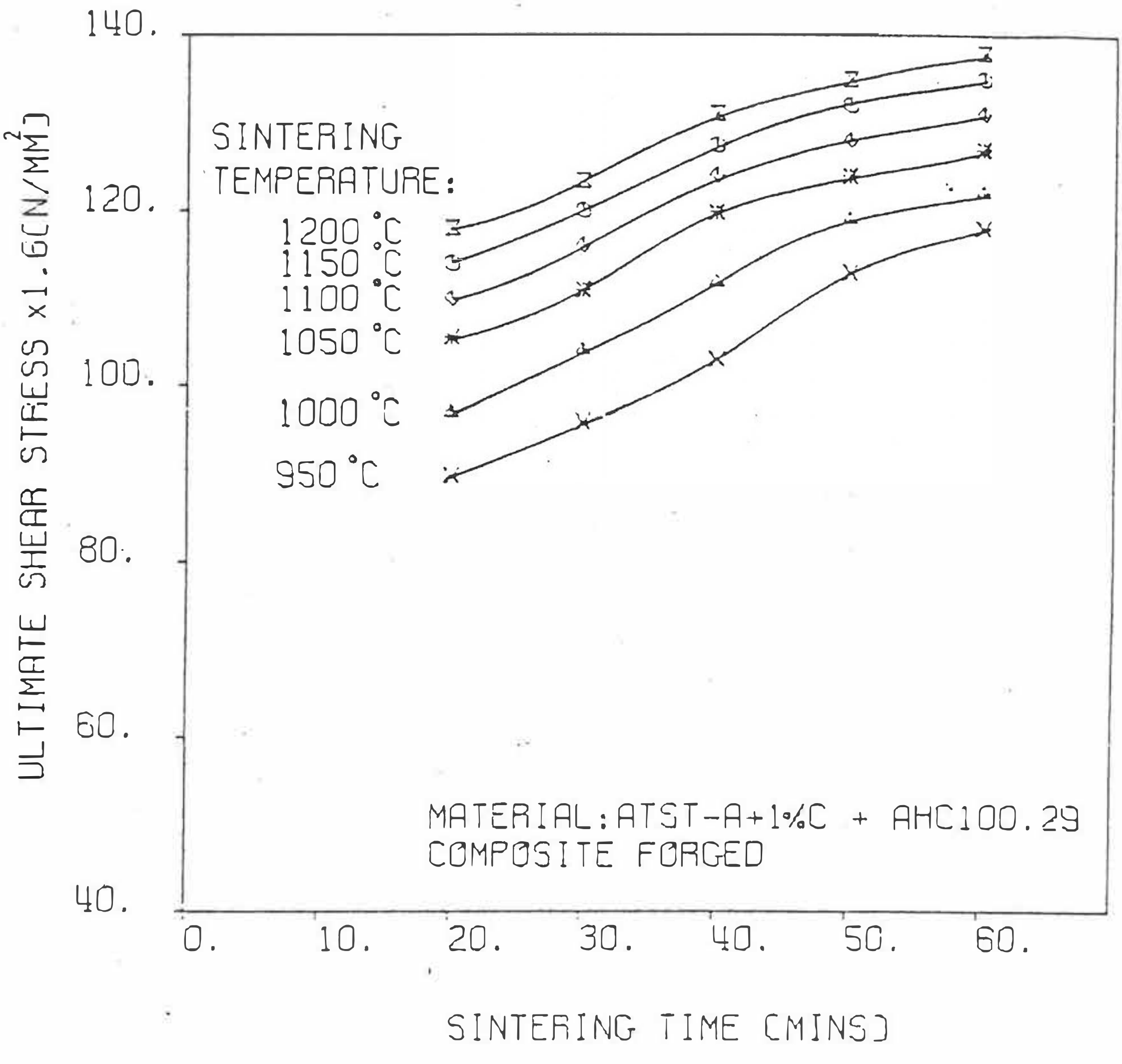


FIG. 5.21 EFFECT OF SINTERING TIME ON THE COMPRESSIVE SHEAR STRESS OF AN IRON BASE SINTER FORGED COMPOSITE MATERIAL.

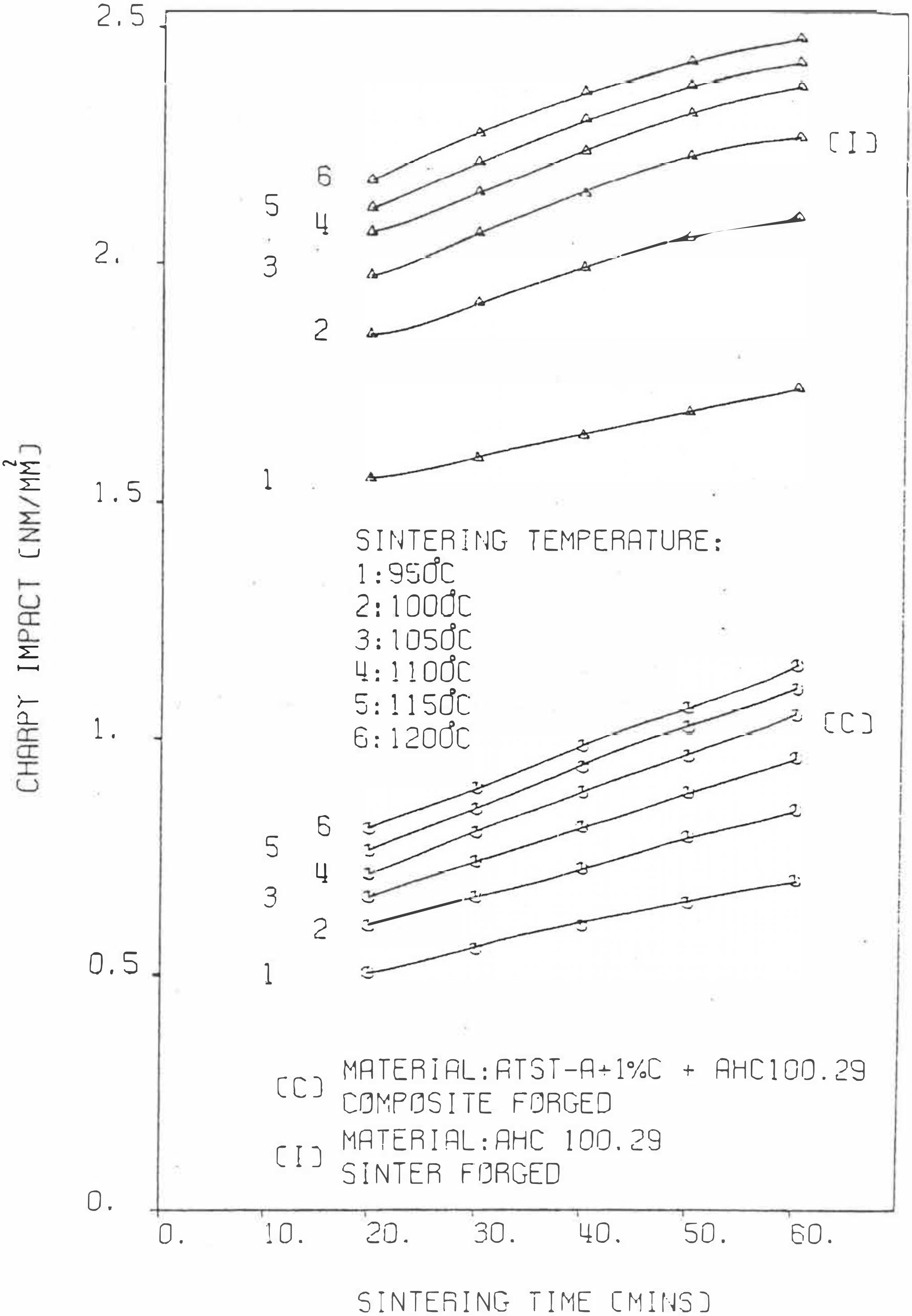


FIG. 5.22.a EFFECT OF SINTERING TIME ON THE CHARPY IMPACT RESISTANCE OF THE IRON BASE MATERIALS.

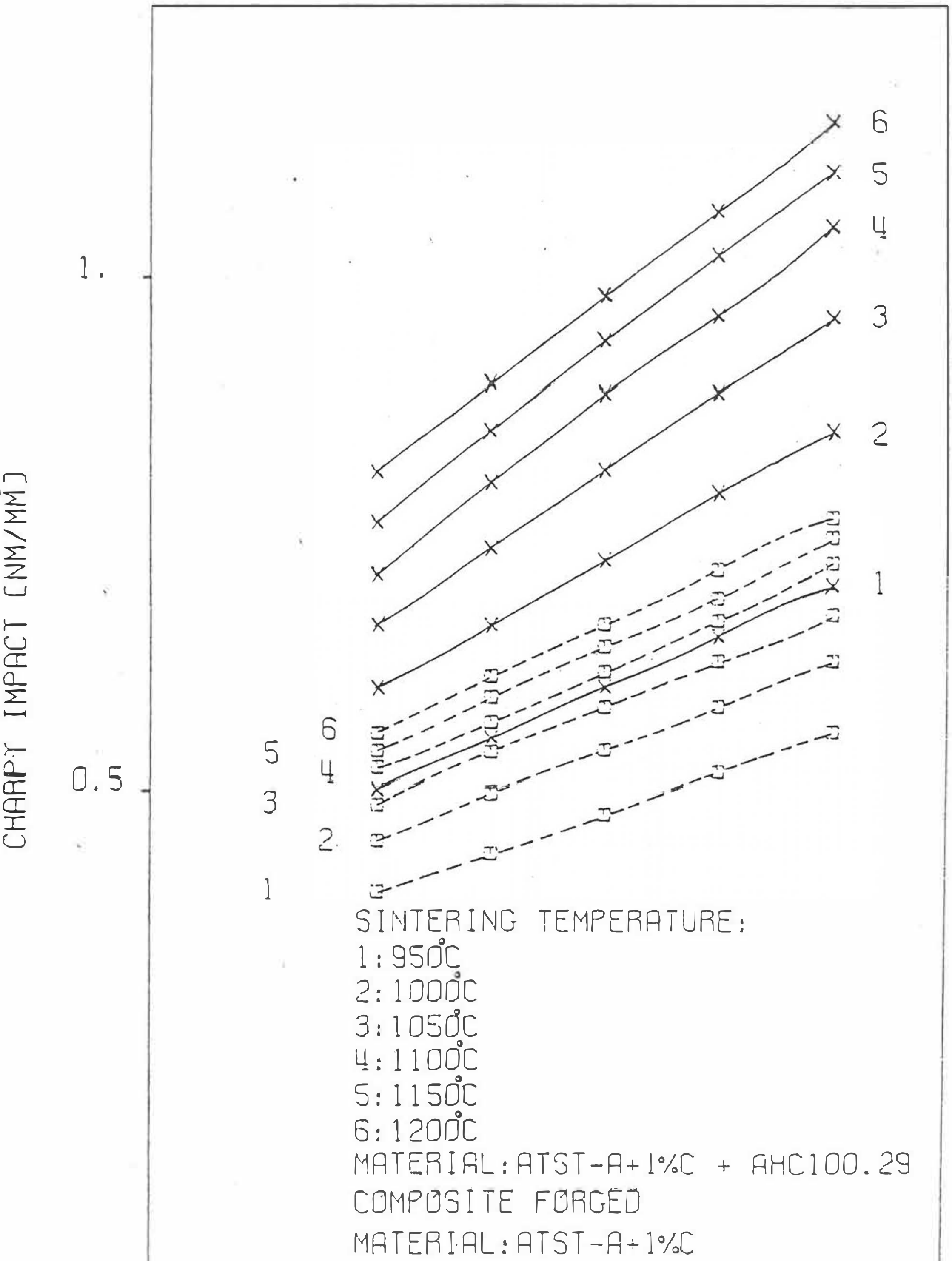


FIG. 5.22.b EFFECT OF SINTERING TIME ON THE CHARPY IMPACT RESISTANCE OF THE IRON BASE MATERIALS.

higher sintering temperatures, the Charpy impact values increased with increasing sintering times. This result reveals the necessity of using higher sintering temperatures for obtaining higher impact values.

The Charpy impact specimens for ATST-A + 1% graphite low alloy steel powder had again lower Charpy impact values than the plain iron and composite specimens. Increasing sintering times contributed little to increasing charpy impact values even at higher sintering temperatures. As was explained in section 5.2, high added carbon content (1% graphite) and unreduced alloying oxides (especially manganase oxide) might have been mainly responsible for the lower impact values.

Plain iron AHC.100.29 - ATST-A + 1% graphite low alloy steel composite powder forged specimens had impact values higher than the steel but lower than the plain iron specimens. Composite impact specimens in time range tests did not break at the joint, which showed that a strong bond could be formed by using two different materials in a powder forged part.

The time range test results of UTS, percentage elongation and percentage reduction of area are shown in Fig. 5.23, 5.24, 5.25 (a,b) and 5.26. In Fig. 5.23 UTS curves for the composite iron-steel and plain iron specimens are shown. Although the curves followed a similar pattern, composite specimens had marginally higher values than the

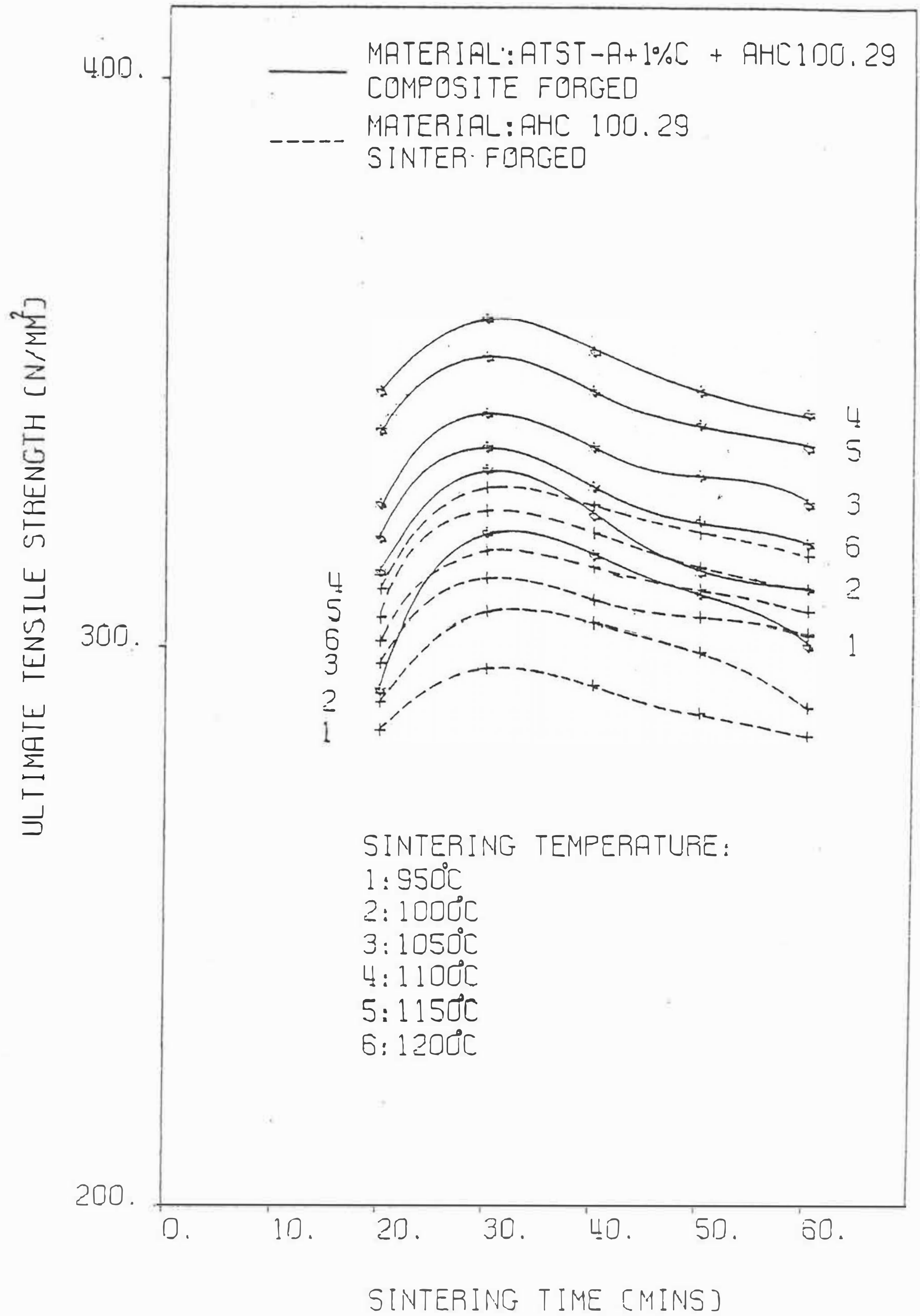


FIG. 5.23 EFFECT OF SINTERING TIME ON THE UTS OF THE IRON BASE MATERIALS.

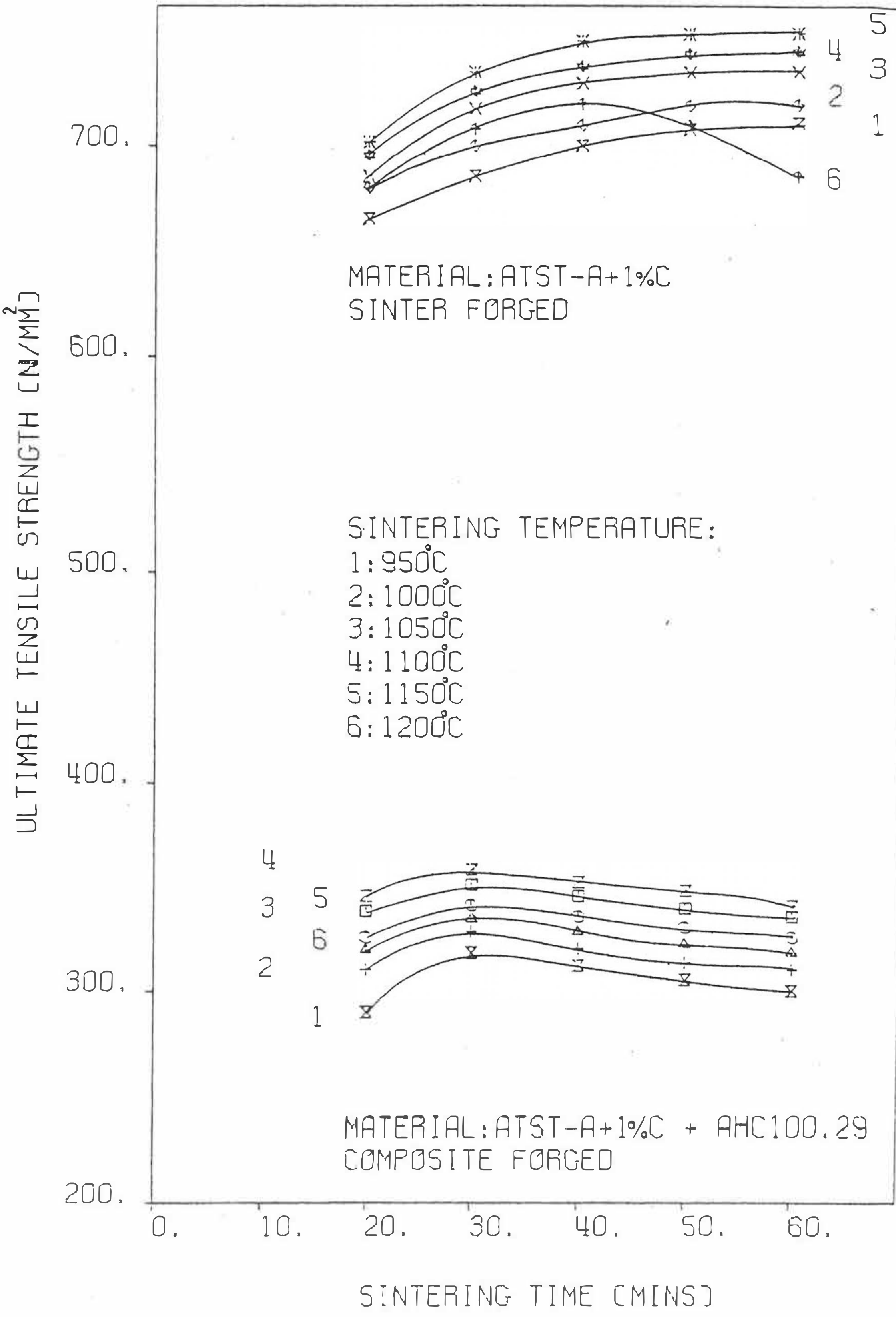


FIG. 5.24 EFFECT OF SINTERING TIME ON THE UTS OF THE IRON BASE MATERIALS.

plain iron specimens. A maximum UTS was observed for a 30 min sintering time. With a further increase in sintering time, UTS values fell gradually probably because of grain growth.

In Fig. 5.24, UTS curves for composite and low alloy steel specimens are given. Low alloy steel specimens had UTS values more than twice those of the composite specimens, UTS values for low alloy steel specimens increased with increased sintering time from 20 minutes to 60 minutes, except for the specimens which were sintered at 1200^{o}C. High sintering temperatures would increase the rate of grain growth and the amount of decarburisation causing a fall of UTS values particularly at high sintering times. The variation of percentage elongation with time is shown in Fig. 5.25 (a,b) for the low alloy steel specimens. The curves follows a similar pattern as for the temperature range tests percentage elongation increased slightly with increasing temperature for the low alloy steel samples. Composite specimens (Fig. 5.25 (a,b) showed an increase in percentage elongation with increasing time. Longer sintering times caused on increase in the amount of ductility. Plain iron specimens (Fig. 5.25(b) showed a similar pattern as for the composite iron-steel specimens. Percentage elongation values of plain iron specimens were higher than for the composite iron-steel specimens.

Percentage reduction of area curves (Fig. 5.26) for low alloy steel specimens followed a similar pattern as

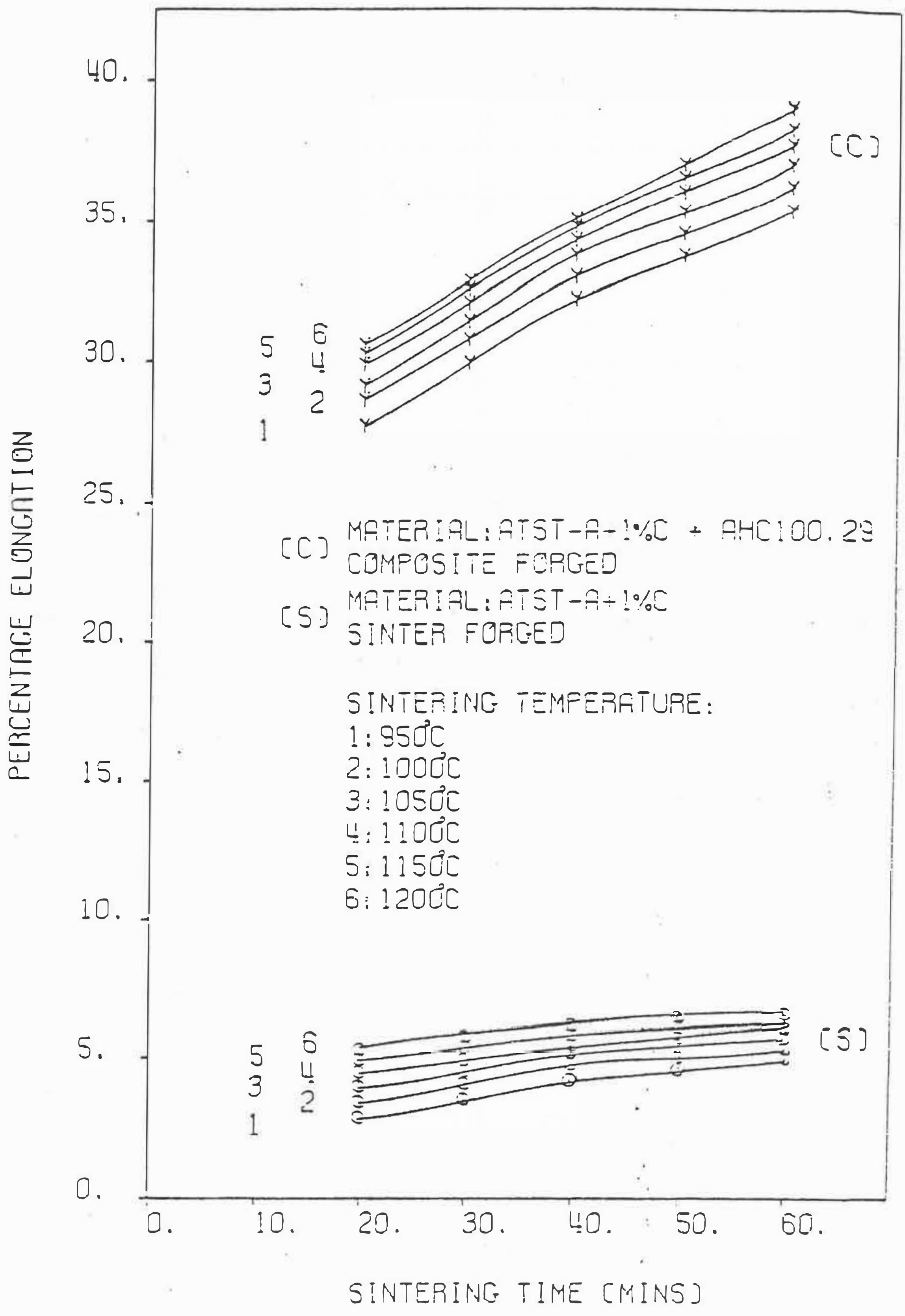


FIG. 5.25.a EFFECT OF SINTERING TIME ON THE PERCENTAGE ELONGATION OF THE IRON BASE MATERIALS.

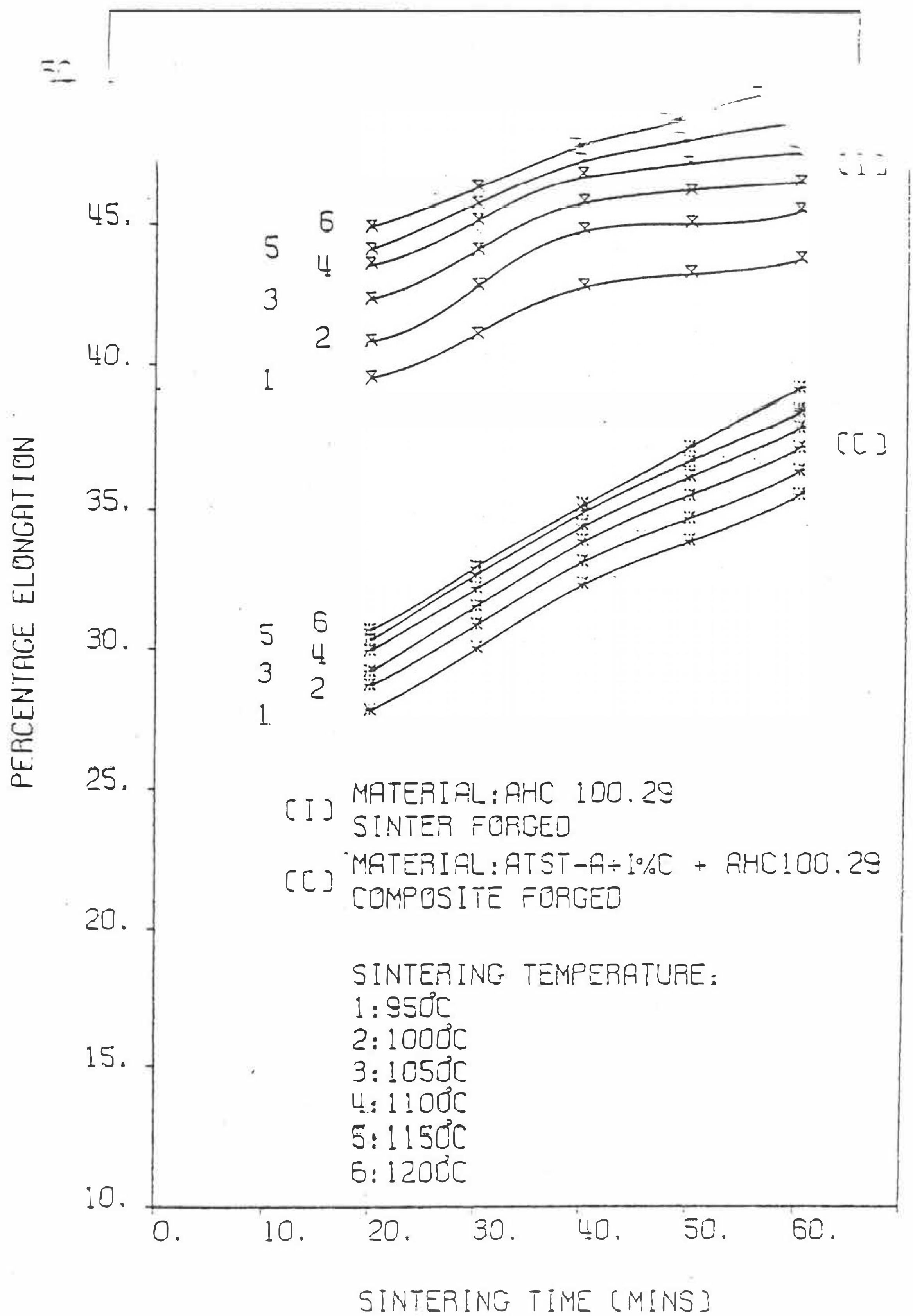


FIG. 5.25.b EFFECT OF SINTERING TIME ON THE PERCENTAGE ELONGATION OF THE IRON BASE MATERIALS.

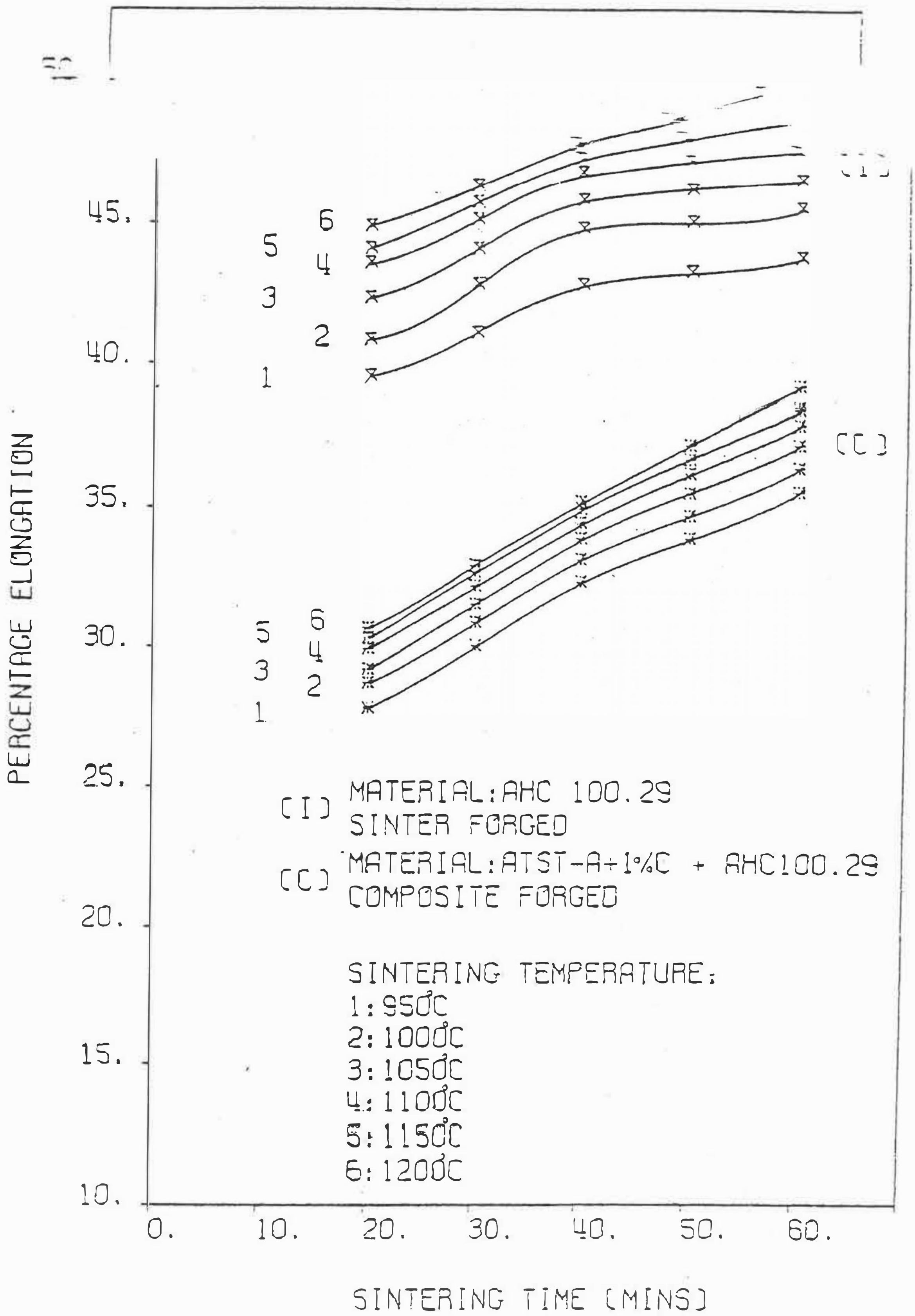


FIG. 5.25.b EFFECT OF SINTERING TIME ON THE PERCENTAGE ELONGATION OF THE IRON BASE MATERIALS.

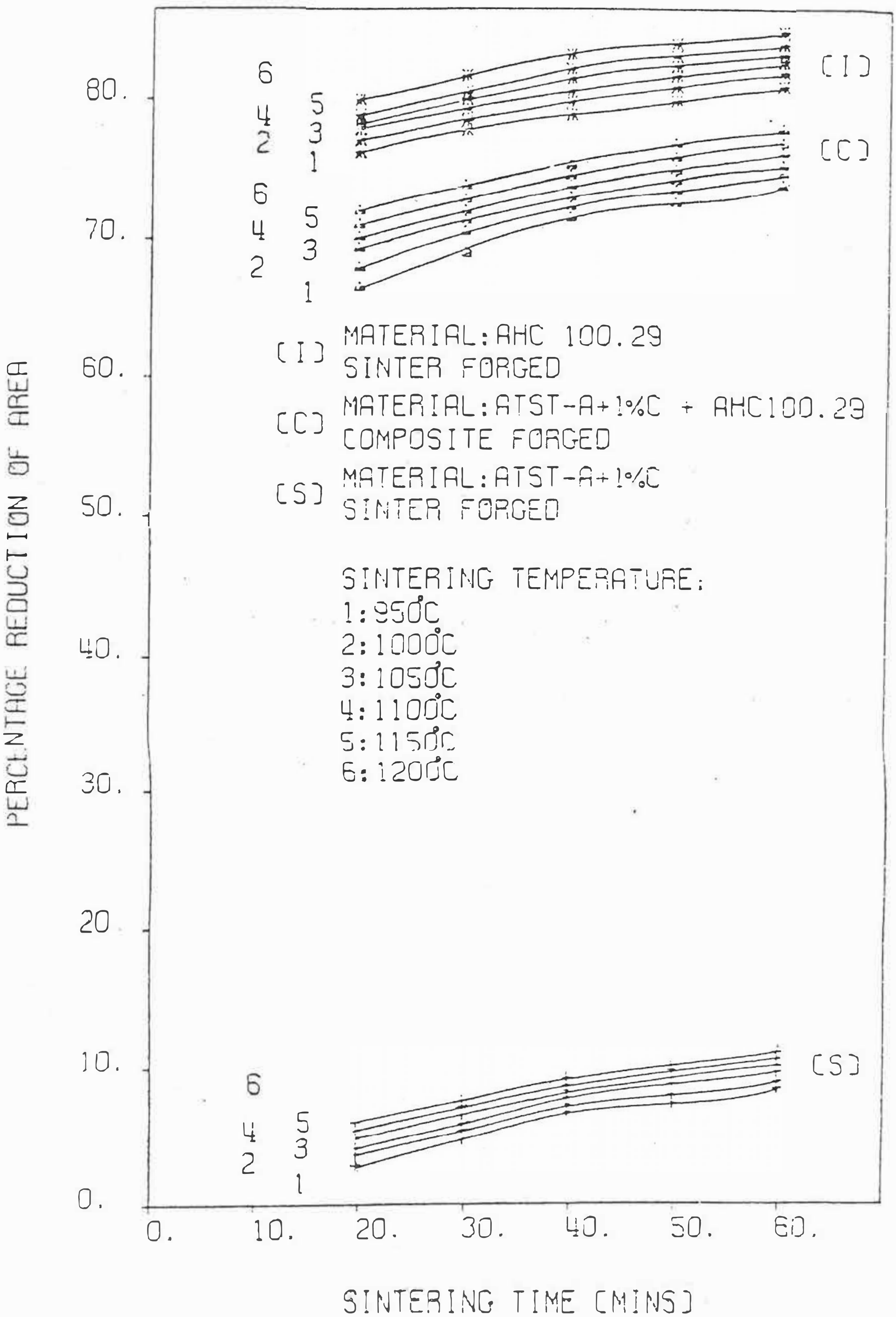
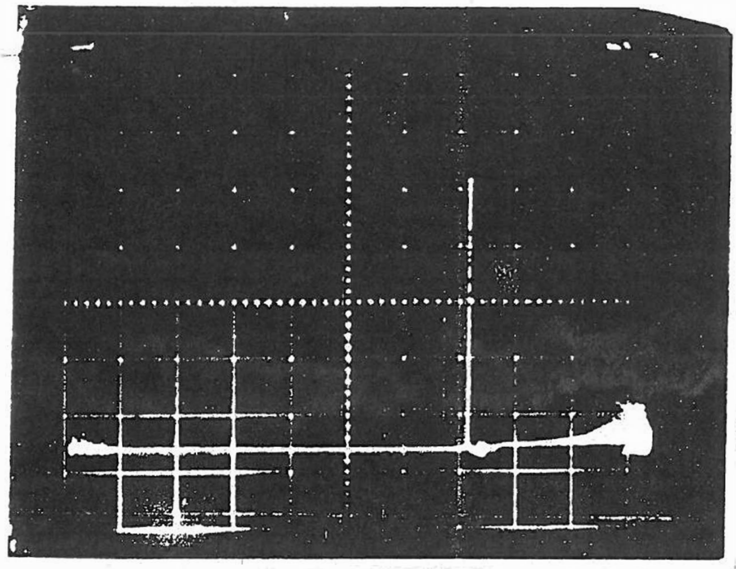


FIG. 5.26 EFFECT OF SINTERING TIME ON THE PERCENTAGE REDUCTION OF AREA.

the percentage elongation curves. With increased sintering times the increase in percentage reduction of area was only marginal. Composite and plain iron specimens followed a similar trend (Fig. 5.26). The values of percentage reduction of area for both curves increased with increased sintering time.

Longer sintering times caused an increase of percentage elongation and percentage reduction of area values for all three compositions of material.

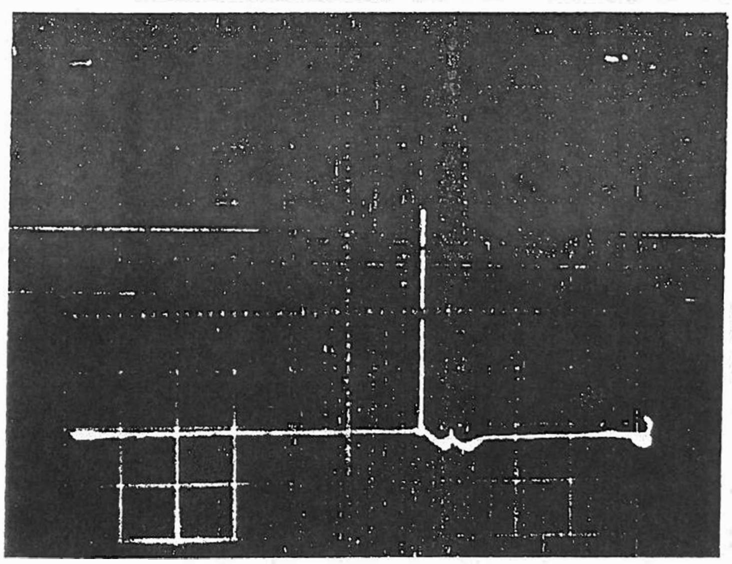
FORGING LOAD
100 KN/DIVISION



TIME (SEC.)

SINTERING TIME: 40 MINUTES
SINTERING TEMPERATURE: 950°C
MATERIAL: ATST-A + 1% C + AHC 100.29

FORGING LOAD
100 KN/DIVISION



TIME (SEC.)

SINTERING TIME: 40 MINUTES
SINTERING TEMPERATURE: 1050°C
MATERIAL: ATST-A + 1% C + AHC 100.29

FIG. NO. 5.P 1   TYPICAL TRACES FOR P/M HOT FORGING AT HIGH SPEED.

(a) x 2000

MATERIAL : ATST-A + 1%C

SINTERING & FORGING AT 1050^{o}C

SINTERING TIME : 60 MINS.

(b) x 2000

MATERIAL : ATST-A + 1%C + AHC 100.29

SINTERING & FORGING AT 1050^{o}C

SINTERING TIME : 60 MINS.

(c) x 2000

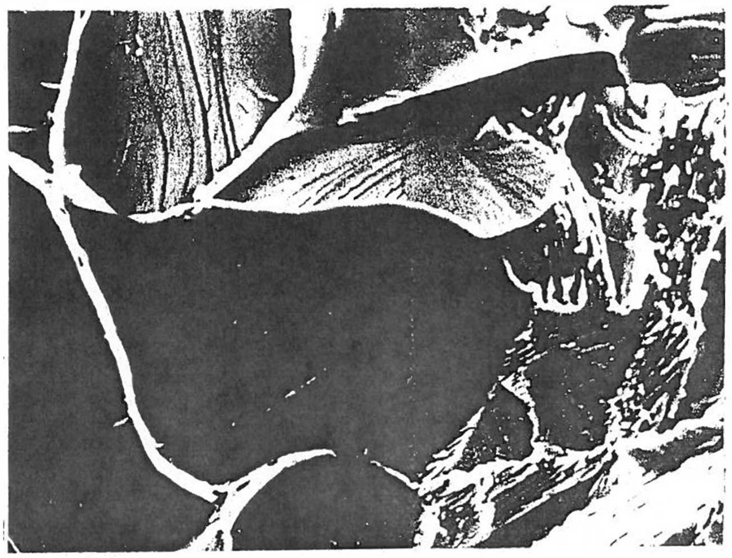
MATERIAL : ATST-A + 1%C + AHC 100.29

SINTERING & FORGING AT 1150^{o}C

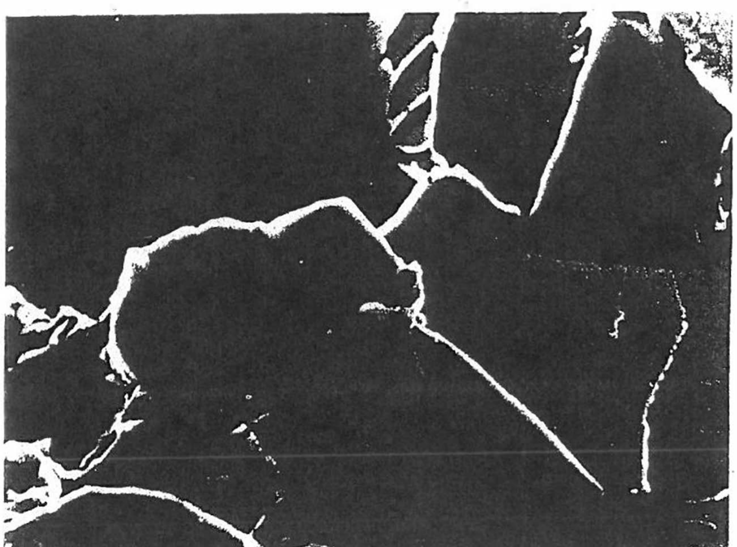
SINTERING TIME : 60 MINS.

FIG. NO. 5.P2 S.E.M. IMPACT FRACTOGRAPHS

(a)



(b)



(c)

(a) x 50

MATERIAL : AHC 100.29

(TYPICAL NECK FORMATION)

SINTERING & FORGING AT 1150°C

SINTERING TIME : 60 MINS.

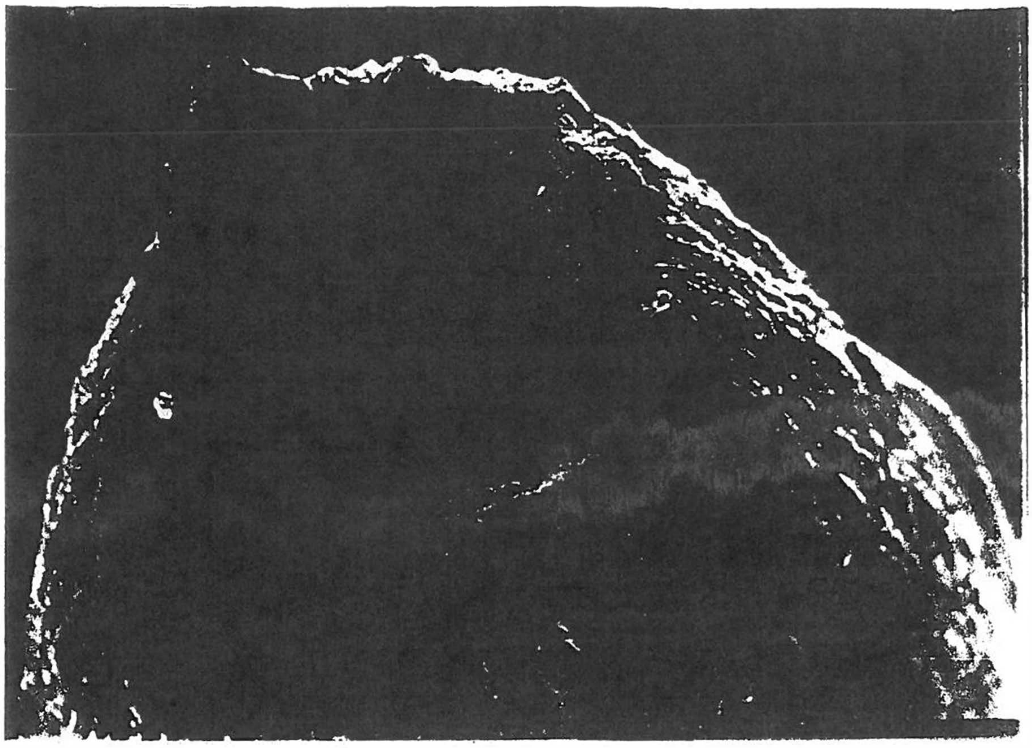
(b) x 2000

MATERIAL : ATST-A + 1%C + AHC 100.29

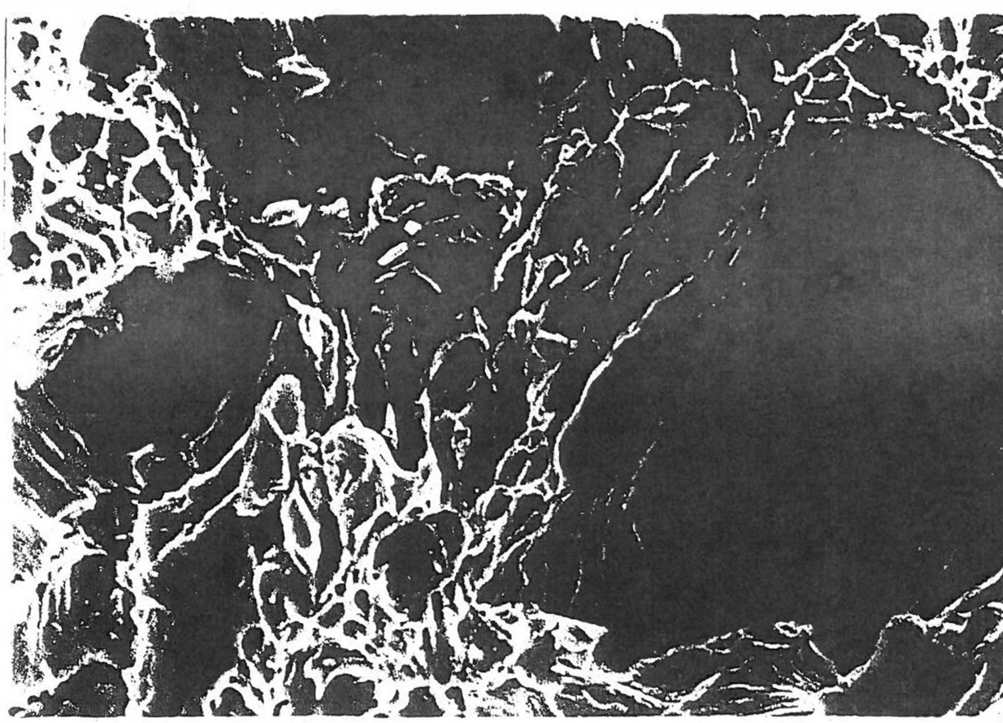
SINTERING & FORGING AT 1000°C

SINTERING TIME : 60 MINS.

FIG. NO. 5.P3 S.E.M. TENSILE FRACTOGRAPHS.



(a)



(b)

(a) x 2000

MATERIAL : ATST-A + 1%C + AHC 100.29

SINTERING & FORGING AT 1200°C

SINTERING TIME : 60 MINS

(b) x 1000

MATERIAL : ATST-A + 1%C

SINTERING & FORGING AT 1200°C

SINTERING TIME : 60 MINS.

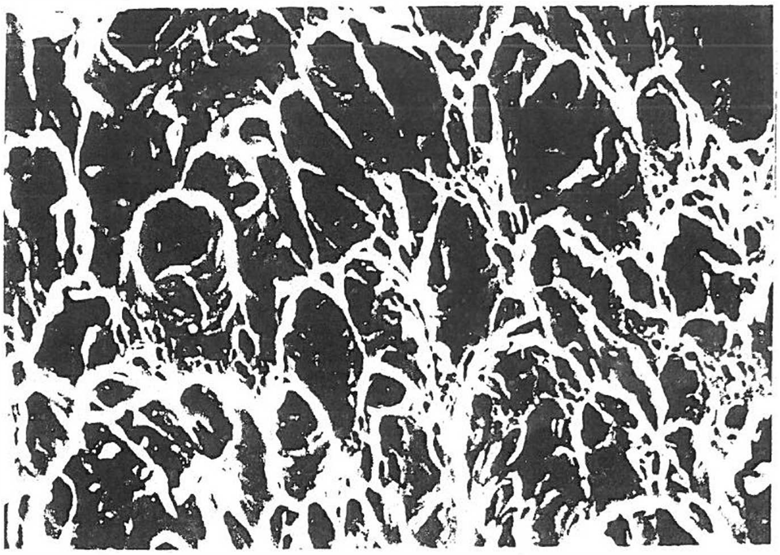
(c) x 2000

MATERIAL : ATST-A + 1%C

SINTERING & FORGING AT 1100°C

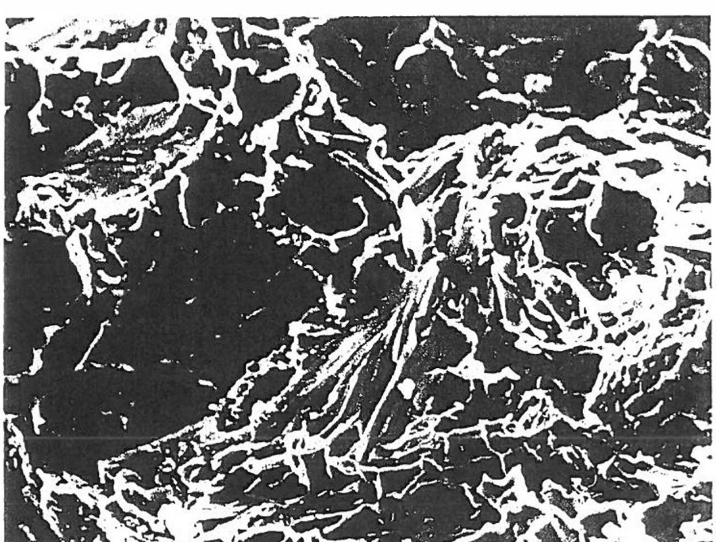
SINTERING TIME : 30 MINS.

FIG. NO. 5.P4 S.E.M. TENSILE FRACTOGRAPHS.



(a)

(b)



(c)

(a) x 160

MATERIAL : ATST-A + 1%C + AHC 100.29

SINTERING & FORGING AT 1150°C

SINTERING TIME : 60 MINS.

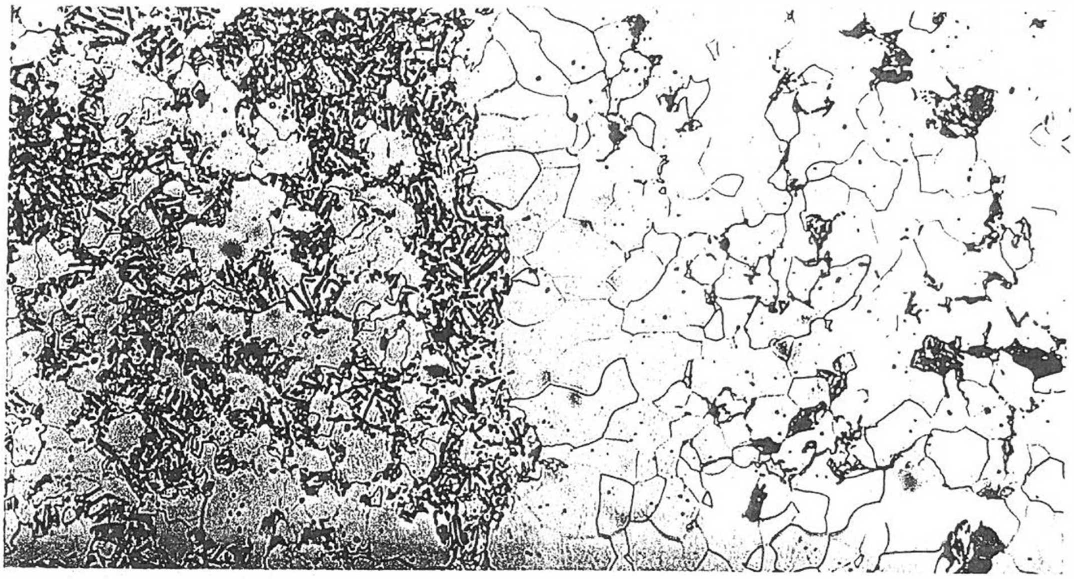
(b) x 160

MATERIAL : ATST-A + 1%C AHC 100.29

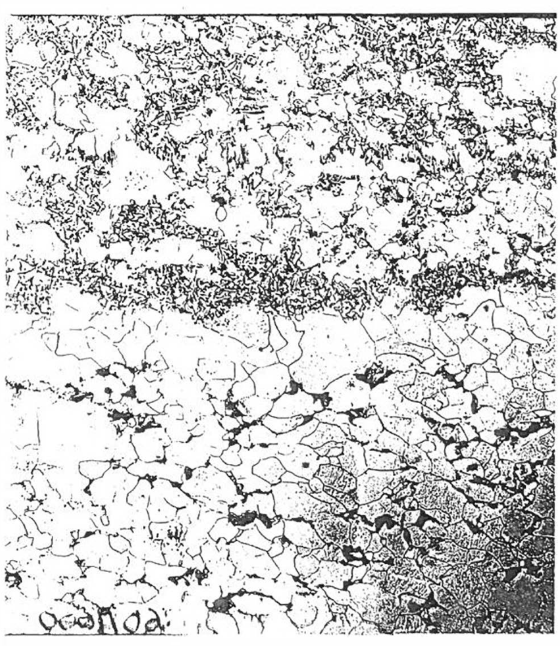
SINTERING & FORGING AT 1000°C

SINTERING TIME : 60 MINS.

FIG. NO. 5.P5 MICROSTRUCTURES OF THE BOUNDARY REGION



(a)



(b)

(a) x 160

MATERIAL : ATST-A + 1%C + AHC 100.29

SINTERING & FORGING AT 1200°C

SINTERING TIME : 30 MINS.

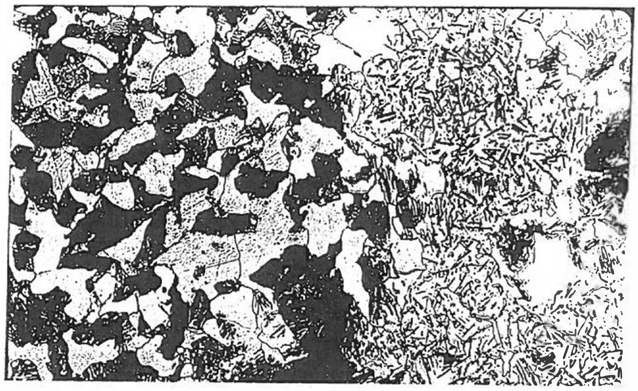
(b) x 80

MATERIAL : ATST-A + 1%C + AHC 100.29

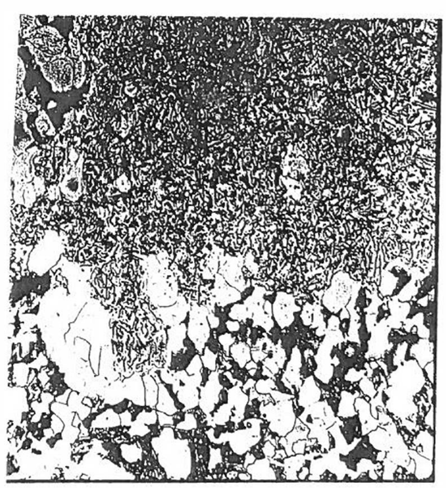
SINTERING & FORGING AT 1000°C

SINTERING TIME : 30 MINS.

FIG. NO. 5.P6 MICROSTRUCTURES OF THE BOUNDARY REGION.



(a)



(b)

CHAPTER SIX

Conclusions and Future Work

6.1 Conclusions

The main conclusions which were drawn from the present work, are given below:

1. A minimum sintering and forging temperature of 950°C is necessary for producing a well bonded two-layer hot P/M forged part.

2. Increasing temperatures above 950°C generally improve the bond strength between plain iron and low alloy steel layers.

3. For effective reduction of metal oxides which could be present in steel powders (in particular magnanase oxide), a minimum sintering temperature of 1200°C or higher is necessary. The reduction of metal oxides leads to a significant improvement of Charpy impact resistance.

4. As far as the mechanical properties are concerned, high sintering temperatures (higher than 1100°C) would be beneficial, except for the UTS of the hot P/M forged parts.

5. Increasing sintering time has a similar effect as the increasing temperature on the strength of the interface.

For a well bonded hot P/M forged part a minimum of 20 minutes sintering time was found to be necessary.

6. Sintering times of longer than 60 minutes are only marginally beneficial as far as the overall mechanical properties are concerned.

7. Sintering times of longer than 40 minutes increase the Charpy impact resistance, compressive shear stress, percentage elongation and percentage reduction of area, but reduce the UTS.

8. In the production of a well bonded two-part P/M hot forged parts, sintering and forging temperature was found to be more effective than sintering time.

9. As was shown from the impact and tensile tests, a strong bond between the two-parts of sinter forged materials can be obtained successfully. None of the two-part P/M hot forged tensile and impact specimens broke at the bond. All failures occurred in the weaker of the two base metals.

6.2 Recommendations for Future Work

6.2.1 Production of Components

The problems involved and the advantages to be gained by the manufacture of two-part components by powder forging needs to be investigated. Applications which show promise are those in which different properties are required in different parts of the components. For example the production of gears in which the teeth, where wear resistance is important could be produced in alloy steel which is bonded to an inner portion of more ductile, cheaper material. Bearings also could be produced with either improved overall properties or reduced cost by this method.

6.2.2 Heat Treatment of Components

The heat treatment of parts which consist of materials which normally require different heat treatments, should be investigated. Particular problems could arise through overheating, decarburisation etc.

6.2.3 Ferrous and Non-Ferrous Materials

Only a limited amount of work has been carried out on the bonding behaviour of ferrous with non-ferrous materials during powder forging. Applications would be for bearings and components in the electrical industry.

6.2.4 Economics

The cost of producing components by two-part powder forging techniques is an important factor in deciding the commercial viability of the process. An investigation should be carried out to compare costs with those of conventional processes.

6.2.5 Production Techniques

Methods of producing two part components by a) pressing the powders together, b) pressing the two parts separately and assembling before sintering and c) pressing the two parts separately and assembling after sintering but before forging need further development and assessment.

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