REX NORMAN BAYLISS.

THE EFFECT OF FORGING REDUCTION ON THE PROPERTIES OF A DIE STEEL.

Submitted for : Master of Science Degree. October, 1967.

SYNOPSIS.

The literature dealing with the effect of forging reduction upon mechanical properties and service performance, with particular reference to hot work steels, has been reviewed. The effect of forging reduction on mechanical properties and the service performance of BSS.224 No. 5. nickel-chromium-molybdenum hot work die steel has been studied by processing a 3½-ton electrically melted ingot, forged in three stages, with forging reductions of 2.92, 9.0 and 18.5/1 respectively. From each bloom six die inserts were manufactured to a hardness range of 363/388 B.H.N. The life of the die inserts was compared with routine production data. The effect of forging reduction on mechanical properties, measured in three mutually perpendicular testing directions, was investigated at room temperature.

The variation of properties is related to the distribution and form of non-metallic inclusions, and not to heterogeneity which, although apparently present on metallographic examination, was not confirmed by chemical and micro-probe analyses. Even so the micro hardness varied in a random manner. Increasing forging reductions did not affect mechanical properties measured in the longitudinal direction. Transverse and depth properties were impaired by heavier amounts of forging, but not to the same extent

that other workers have previously reported. Forging had no influence upon die life except that heavy reduction minimised scatter. Room temperature mechanical properties do not appear to have a bearing on die life.

The published conversion tables of hardness and tensile strength were found to be inapplicable to the range of hardness studied.

THE EFFECT OF FORGING REDUCTION ON THE PROPERTIES OF A DIE STEEL.

1. Factors affecting die life.

The performance of die tools and its economic implications in the drop-forging industry is important. The Drop-forger likes to be able to predict an average number of forgings that can be made from his die tools so that the necessary estimated cost per forging will not be exceeded. Premature failure can involve financial loss by the need for :

(a) Repairing the die tool whilst still in the hammer, i.e. by local reclamation by welding.

(b) Having to remove the die blocks and resink a further impression. This operation is generally caused by local wear or cracking around the impression.

(c) In the event of (b) above, especially if such a change has to take place midway through a working shift, the loss of remunerative production can be considerable.

1.1. The following two examples illustrate the financial loss which was incurred when premature failure of dies was experienced.

(a) A solid die block, 46" x 19" x 17", costing £500., was used to produce light diesel crankshaft drop forgings. The machining costs were at least £500. more. Based upon previous manufacturing statistics, the Dropforger expected to produce a minimum of 5000 drop forged crankshafts. It had been noted that occasionally die block lives were extended and 6000 forgings had sometimes been made. This latter figure was welcomed by the Drop Forging Company, for clearly, production costs are more favourable when the calculated minimum die life is exceeded. However, in the particular case under discussion, premature failure occurred after producing only 1500 forgings. Excluding labour costs etc., and assuming an average life had been obtained, the minimum financial loss, due to die block failure, was £260. This quoted figure of £260. does not take into account re-setting costs involved when new die blocks were fitted, or the even greater loss caused by production discontinuity.

(b) A further typical case of early die block failure, when producing drop forged connecting rods, gave a minimum financial loss of £400. due to not achieving the desired production commitment.

In both cases quoted, severe transverse cracking led to withdrawal of the die blocks from service. These two simple examples demonstrate how important it is for die blocks to have a reasonable and above all, predictable life.

2. Production of Die Blocks.

There are two methods of making wrought die blocks.

1. From a solid block which can be re-machined, perhaps more than once, when the impression becomes worn by use so that forging can continue until the block is too thin to allow the operation of re-sinking to be carried out again.

2. By using inserts in which the die impression is produced in a relatively thin section of die steel which when worn out cannot be re-machined.

The mechanical properties of fully heat treated solid die blocks, and those existing in the smaller insert form, cannot be, in any way, comparable. For example, the properties existing in large solid die blocks having a cross section of 30" x 18" will differ from the properties present in smaller sections of die inserts where dimensions may only be 6" x 3", even assuming that an identical heat treatment was carried out.

This investigation will deal with the properties of hot work die steel in insert form.

2.1. Variations affecting performance.

Many working variables affect die life during the normal course of hot working. These variables fall into two groups.

(a) The inherent properties of the wrought die block which are determined by the steelmaker's procedure and the die block producer's manufacturing technique.

(b) Determined by drop forging practice, such variables include :

1. Preheating of the die blocks.
2. Die design.
3. Lubricants used.
4. Preheating of the stock bar.
5. Mechanical damage.

2.2. It is proposed to evaluate the effect of the die block producer's practice alone on die life variability, so that by determining the contribution of any variations in the properties of the forged stock from which the die is made, a basis may be provided for further studies in the drop forge itself. It is assumed in the trials of the inserts in the forge that the variables are random and of a generally consistent nature.

3. Steelmaking.

British Standard Specification 224, No. 5. die steel is a medium carbon, nickel-chromium-molybdenum steel, of the following composition :-

Carbon	0.50/0.60%
Silicon	0.30% max.
Manganese	0.50/0.80%
Sulphur	0.04% max.
Phosphorus	0.04% max.
Nickel	1.25/1.75%
Chromium	0.50/0.80%
Molybdenum	0.25/0.30%

The use of either acid open hearth or basic electrically melted steel is permitted. Electrically melted steel has a characteristically low sulphur content of approximately 0.015% because of double slagging practice. Whilst it is universally agreed that a low sulphur content is associated with high quality, such low sulphur contents in fully processed die blocks lead to bad machinability. Otherwise no differences have been reported between alloys of the same composition made by the two steelmaking processes.

In the remainder of this section the comments relate to steel which may have been made by either process.

3.1. Deoxidation.

An important part of the process of steelmaking is deoxidation which prevents the formation of blowholes and pores in the solidified ingot. Plockinger (1) has shown that deoxidation must be carried out in such a manner that the liquid steel, before teeming, contains as little dissolved oxygen as possible, otherwise an increase in the content of non-metallic inclusions will occur.

Whittaker (2) emphasized that in controlling exogeneous inclusions, the steelmaker must always ensure that the refractory materials which come into contact with the molten steel must be of high quality and also that care must be exercised in the casting pit. Such precautions are always adhered to in steelmaking.

Nevertheless, it is inevitable that all die steels will contain non-metallic inclusions, whose effect on mechanical properties was studied by Dieter, McCleary & Ransome (3). They showed that in specimens taken from forgings, transverse ductility is influenced primarily by the size, type and distribution of the non-metallic inclusions. Loria (4), and Wells & Mehl (5) agree that inclusions play a vital part in determining the transverse percentage on fully processed forgings, but the latter workers attributed differences between the transverse and the longitudinal percentage of reduction of area in steel forgings largely to the presence of inclusions.

Examination of fractured tensile specimens showed that discontinuous stringers consisting of hard angular oxide inclusions caused consistently lower values of percentage of reduction of area in the transverse direction than did continuous sulphide stringers. Finally, Finneston & Fearnhough (6) indicated that composition, size, shape and positions of the inclusions, relative to the applied stress, are important factors in determining their effect on mechanical properties, e.g. it is probable that silicate inclusions which are drawn out and pointed at both ends are more dangerous than sulphide inclusions which are rounded at their ends.

There is adequate recorded evidence that non-metallic inclusions have an adverse effect on mechanical properties, particularly the ductility: therefore, in considering die block manufacture, it becomes readily apparent that controlling the quantity and character of non-metallic inclusions is important inasmuch as subsequent die block properties are affected in this way. Because the steelmaker cannot eliminate non-metallic inclusions from ingots, and the problem is aggravated with increasing ingot size, then the forging process must ensure that the least damage to the mechanical properties by the effect of inclusions takes place.

4. Ingot Structure.

When No. 5. die steel is made, the maximum ingot weight is normally 40-tons. Ingots may be cast into square or octagonal moulds, and subsequently either rolled and/or forged into billets, depending on ingot size. The smaller ingots up to 5-tons can be rolled to billet shape. The solidification of ingots inevitably gives rise to segregation, which is caused chiefly by the constituents last to solidify - for example, at the centre of the ingot, segregation is always found. Two forms of segregation are recognised, the A and V forms. Segregation is called normal or positive when the surface regions of the ingot are enriched in high melting compounds; conversely, the central regions are enriched with low melting compounds. Negative or inverse segregation occurs when low melting compounds solidify first in the regions of high melting solute. Normally inverse segregation can be found in a cone shape at the lower part of ingots. Aitchinson & Pumphrey (7) considered in detail the work of the Heterogeneity Committee (8) on the formation of A and V segregates that occur during solidification of steel ingots. The zones of A segregatesoccur in the middle regions of the ingot, occasionally near the interface between the columnar and equiaxed crystals, but, more usually, within the zone completely occupied by equiaxed crystals. The V segregates occur in the axial regions of the ingot.

4.1. Segregation.

In the work of the Heterogeneity Committee it was reported that higher concentrations of sulphur, phosphorus and carbon occur more readily near the ingot centre and ingot head. The term "segregation" not only applies to concentrations of impurities but equally to major constituent concentrations, for example, the segregation which occurs between the arms of the dendrites is known as microsegregation, whilst the impurities which occur around the major and secondary ingot piping are termed macrosegregation, (Fig 1). Winegard (9) considers that micro-segregation includes short range differences between cells and grains, whilst macrosegregation refers to the large range variations in composition found between centre and ingot outside. This convention will be used in this thesis.

Spretnak (10) studied segregation in six alloy ingots with weights ranging from 3100 to 6160 lbs. One of these was a 13" round fluted electrically melted, bottom poured, nickel-chromium-molybdenum ingot. The results obtained (Fig 2). show that segregation of carbon, manganese and silicon, and an unusual amount of negative segregation in nickel, were present. When considering the position of sampling and the results of chemical analysis, it is unfortunate that Spretnak (10) did not indicate the precise cast chemical composition, apart from stating the general chemical specification requirements of SAE.4340 which has the following composition (the specification for No. 5. die steel is given for comparison).

	SAE.4340. %	B.S.S.224. NO.5. %
Carbon	0.38/0.43	0.50/0.60
Silicon	0.20/0.43	0.30 max
Sulphur	0.025 max.	0.040 max.
Phosphorus	0.025 max.	0.040 max.
Manganese	0.65/0.85	0.50/0.80
Nickel	1.65/2.0	1.25/1.75
Chromium	0.70/0.90	0.50/0.80
Molybdenum	0.20/0.30	0.25/0.35

The chief compositional differences between the two specifications are that No. 5. has a higher carbon content and possibly lower nickel content. Nevertheless, the two alloys are still comparable.

Although Spretnak's results show that nickel and silicon are virtually constant, they are much lower than specification requirements which are assumed to have been met. It can only be concluded that unusual inverse segregation of nickel is present and, in the absence of any other independent confirmation of this effect, Spretnak's conclusions must be treated with reservations as the number of chemical analyses undertaken were too few to indicate firm conclusions as to the segregation pattern.

After examining the results for six ingots it was concluded that the segregation ratios (maxima/minima %) for the elements in the investigated ingots agreed fairly well with the work of the Heterogeneity Ingot Steel Committee (11), although the magnitude of the ratios varied from ingot to ingot.

The most sophisticated studies on ingot heterogeneity were undertaken by Rogues, Martin, Dubois & Bastien (12) who studied steels of the following compositional range :-

	%
Carbon	0.10/0.55
Silicon	0.15/0.40
Manganese	0.50/1.25
Nickel	Up to 2.7
Chromium	Up to 3.5
Molybdenum	Up to 1.0

The ingot weights varied from 5 tons to 150 tons. The statistical survey of the results on the heterogeneities of these large forging ingots was based upon the examination of about 200 ingots, of which 119 ingots were cast by the open hearth process, the remainder being electrically melted. Of these ingots, 136 were carbon steel, the remaining ingots were manufactured to alloy steel specification.

The investigators determined by statistical analysis the factors influencing the total amplitude of segregation along the ingot axes. The only two factors which proved to affect segregation were (a) ingot weight and (b) to a lesser extent, alloying elements of which molybdenum, chromium, manganese and nickel segregate less than sulphur, carbon and phosphorus. As ingot weight increased from 6 tons to 150 tons, so did positive segregation.

Segregation of all elements other than hydrogen and oxygen was essentially a function of ingot weight. It was concluded that whatever the composition of the steel, or ingot size, the prime factor in the tendency for inclusions to form in the ingot base depended upon the casting conditions. The different steelmaking processes of open hearth and basic arc did not have an influence, except when the presence of gases (oxygen, hydrogen, nitrogen) and impurities (sulphur and phosphorus) were directly affected by choice of steelmaking process.

The Heterogeneity Committee (11) in their work considered chemical segregation in carbon and alloy steel ingots, and some of the results on carbon steels are given, (Table 1).

The method used to calculate the segregation ratios given in Tables 1 & 2. was based upon the weight of each ingot and then, under the heading of each element present, are set out -

(a) the mean composition as determined by the pit sample, or other means.

(b) the maximum content of the element in the body of the ingot, ignoring the segregated region in the head, which is discarded for the purposes of the enquiry.

(c) the minimum content of the element.

(d) the difference between these two values calculated as a percentage of the mean.

Table 1. demonstrates that sulphur, phosphorus and carbon segregate extensively, particularly severe segregation is observed in the axial position.

Although errors in chemical analysis may affect the range percentage index, the results demonstrate most clearly the marked increase in segregation that occurs as the ingot weight and dimensions increase.

Table 2. shows that chromium segregates moderately, whilst nickel shows the least segregation. Unfortunately, there is no information in Table 2. which enables the influence of alloy additions in the segregation of carbon, sulphur and phosphorus to be evaluated.

In the report of the Heterogeneity Ingot Steel Committee (11), it is stated that "in passing it should be noted that the presence of alloy additions has been found not appreciably to alter the segregation of the

TABLE 1. SEGREGATION IN PLAIN CARBON STEELS.

WEIGHT.			CARBON.				SILICON.				MANGANESE.				SULPHUR.				PHOSPHORUS.			
			Mean.	Max.	Min.	Range.	Mean.	Max.	Min.	Range.	Mean.	Max.	Min.	Range.	Mean.	Max.	Min.	Range.	Mean.	Max.	Min.	Range.
T.	Cwts.		%	%	%	%	%	%	%	%	%	%	%	%	%	%	%	%	%	%	%	%
1	0		0.36	0.37	0.32	15	0.21	0.25	0.24	5	0.60	0.57	0.55	5	0.030	0.028	0.023	15	0.033	0.030	0.026	10
1	5		0.52	0.54	0.48	10	0.21	0.22	0.20	10	0.84	0.84	0.79	5	0.037	0.040	0.032	20	0.036	0.038	0.032	15
1	16		0.42	0.48	0.41	15	0.23	0.25	0.24	5	0.68	0.71	0.67	5	0.012	0.010	0.007	25	0.020	0.023	0.013	50
2	10		0.41	0.46	0.40	15	0.52	0.52	0.50	5	1.06	1.04	0.98	5	0.056	0.057	0.044	25	0.052	0.062	0.047	30
2	15		0.34	0.40	0.33	20	0.28	0.23	0.22	5	0.72	0.78	0.73	5	0.040	0.043	0.032	20	0.043	0.052	0.039	30
2	18		0.60	0.64	0.53	20	0.23	0.21	0.20	5	0.77	0.79	0.76	5	0.038	0.040	0.031	25	0.040	0.048	0.037	30
2	18½		0.34	0.40	0.31	30	0.16	0.15	0.14	5	0.70	0.74	0.69	5	0.049	0.050	0.034	35	0.039	0.051	0.034	45
3	5		0.40	0.45	0.35	25	-	-	-	-	0.84	0.85	0.78	10	0.034	0.050	0.027	70	0.042	0.048	0.032	40
5	14		0.13	0.11	0.08	25	0.05	0.07	0.06	20	0.42	0.54	0.50	10	0.029	0.041	0.028	45	0.025	0.020	0.014	25
8	0		0.39	0.45	0.31	35	-	-	-	-	0.96	1.04	0.95	10	0.024	0.29	0.016	55	0.025	0.031	0.023	30
8	5		0.21	0.28	0.18	50	-	-	-	-	-	-	-	-	0.027	0.33	0.017	60	0.059	0.069	0.050	30
10	10		0.30	0.34	0.21	45	0.14	0.14	0.13	5	0.74	0.77	0.68	10	0.017	0.026	0.009	100	0.010	0.011	0.009	20
20	0		0.21	0.25	0.16	40	0.17	0.20	0.19	5	0.67	0.68	0.64	5	0.032	0.035	0.024	35	0.040	0.051	0.037	35
24	17½		0.44	0.49	0.35	30	0.29	0.29	0.27	5	0.78	0.80	0.73	10	0.033	0.048	0.025	70	0.042	0.042	0.030	30
54	0	a	0.36	0.42	0.28	40	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
54	0	b	0.34	0.39	0.24	45	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
64	0		0.40	0.50	0.33	45	0.16	0.17	0.16	5	0.63	0.69	0.63	10	0.035	0.050	0.029	60	0.038	0.050	0.026	65
110	0		0.27	0.37	0.15	80	0.24	0.23	0.20	15	0.71	0.71	0.63	10	0.033	0.059	0.021	115	0.024	0.050	0.024	100
172	0		0.33	0.55	0.19	110	0.13	0.14	0.11	25	0.79	0.92	0.73	10	0.030	0.080	0.018	205	0.033	0.090	0.023	205

TABLE 2. ALLOY STEELS.

T.	Cwts.	NICKEL. Mean.	NICKEL. Max.	NICKEL. Min.	NICKEL. Range.	CHROMIUM. Mean.	CHROMIUM. Max.	CHROMIUM. Min.	CHROMIUM. Range.	TUNGSTEN. Mean.	TUNGSTEN. Max.	TUNGSTEN. Min.	TUNGSTEN. Range.	MOLYBDENUM. Mean.	MOLYBDENUM. Max.	MOLYBDENUM. Min.	MOLYBDENUM. Range.	COPPER. Mean.	COPPER. Max.	COPPER. Min.	COPPER. Range.	ALUMINIUM. Mean.	ALUMINIUM. Max.	ALUMINIUM. Min.	ALUMINIUM. Range.
		%	%	%	%	%	%	%	%	%	%	%	%	%	%	%	%	%	%	%	%	%	%	%	%
	13	8.24	8.33	8.11	2½	17.98	18.09	18.01	½	0.60	0.62	0.59	5	-	-	-	-	-	-	-	-	-	-	-	-
	15	3.11	3.10	3.02	2½	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
	17	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	0.030	0.035	0.027	25
1	16	0.05	0.06	0.04	40	-	-	-	-	-	-	-	-	-	-	-	-	0.06	0.07	0.06	-	0.075	0.078	0.071	10
2	4	4.38	4.49	4.35	2½	1.41	1.35	1.30	2½	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
2	4	4.38	4.44	4.32	2½	1.41	1.36	1.27	5	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
2	17	3.11	3.15	3.05	2½	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
49	-	3.31	3.34	3.14	5	1.67	1.68	1.57	5	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
50	-	0.40	0.40	0.36	10	-	-	-	-	-	-	-	-	-	-	-	-	0.045	0.052	0.037	10	-	-	-	-
119	10	2.48	2.52	2.39	5	0.68	0.67	0.58	15	-	-	-	-	0.63	0.74	0.55	30	-	-	-	-	-	-	-	-

other elements, carbon, sulphur and phosphorus". However, it must be pointed out that no figures were given.

Winegard (9) states "one of the difficulties met in solidification of metals is that they are composed of non-isotropic crystals which can be grouped (into columnar or equiaxed, or a combination of both)in ways to produce a solid that is microscopically anisotropic and thus, unhomogeneous. Because macroscopic segregation gives rise to non-uniform physical properties, only with the control of structure can segregation be minimised". However, there are practical reasons why some degree of segregation is inevitable.

Smith (13) and (14) concluded that alloy segregation was greatest in the equiaxed zones of 50 Kg. ingots of En.24. This specification is similar to that for No. 5. Die steel, except the carbon range is lower. The rate of freezing of the ingot also affects segregation in that the slower the rate of solidification, the higher the segregation ratio of alloying elements. The severity of segregation of the elements studied was in the following order -

Mo. > Cr. > Mn. > Ni.

(i.e. confirming the work of the Heterogeneity Committee). He also suggested that the segregation ratio and ingot macrostructure could be controlled by altering the columnar/equiaxed crystalline ratios so that in subsequent hot working the degree of banding could be minimised. When trying to apply this approach to large ingot casting, on an industrial scale, no immediate hopes for success can be seen because the rate of ingot solidification can never be rapid enough. Therefore, it can be confirmed the larger the ingot the greater the segregation.

With the development of the continuous casting process, greater scope for minimising segregation is possible. Johnson et al (15) states that segregation of phosphorus, sulphur and carbon is much lower with continuous cast carbon and alloy steels than with conventionally cast ingots. The limiting size of 9" square in producing continuous cast die steel, and severe internal cavities, precludes the use for die block manufacture in its present form. However, with continuing development, if the problem of quality and casting size can be improved, then this process could be re-examined for potential die block forging.

The as cast structure presents various crystalline shapes relative to ingot position. For example, the bottom end of an ingot will have larger columnar crystals than will form at the ingot top. The high re-heating temperature which must be employed on the ingot before forging commences, besides rendering the metal plastic in order to allow deformation, has the further advantage of assisting in the diffusion of the segregated elements, thus improving, but regrettably only to a small extent, chemical homogeneity. In this connection, Thorneycroft & Stevens (16) state "the microsegregation intensity in forged bar samples derived from a small En.24 $1\frac{1}{2}$-ton ingot was reduced by heat treating the samples at 1200°C. in Vacuo for 7 days". Evidence of the successful application of homogenisation was obtained on thin slices taken from a 2" square bar. The samples were re-austenitised and quenched, after being allowed to transform to various degrees, after which a hardness survey was made. The specimen hardness was then compared with that of the unhomogenised material. The comparison shows that the transformation characteristics are much more uniform when the high temperature homogenising treatment is applied. A further factor stated in support of the beneficial influence of the diffusion anneal was that the microstructures of the partly transformed specimens exhibited freedom from banding.

This response proved that the homogenising treatment can reduce the intensity of microsegregation, and prolonged diffusion annealing treatment at high temperature would produce almost complete chemical homogeneity. This type of high temperature treatment is not practical because large ingots suffer from more pronounced segregation than that encountered in the small experimental ingots, and the cost of a minimum of 7 days vacuum treatment at 1200°C. would be prohibitive. The literature survey has shown that a considerable amount of segregation data has been obtained from studies of large ingots. The metallurgical condition of forged products from such ingots is distinctly different from that prevailing in forged blooms manufactured from smaller ingots; however, such results can be expected to indicate the general trend of segregation in wrought products derived from smaller ingots, and will be of relevance to the results reported in this thesis which have been obtained from forged blooms made from a relatively small ingot weighing 3-tons 9-cwts.

5. HOT WORKING.

The processing of die blocks includes the following considerations:

a. Reheating.
b. Position in ingot.
c. Direct forging.
d. Upset forging.
e. Heat treatment.

The conversion of the primary ingot structure, consisting of columnar and equiaxed crystals into a grain flow or fibre parallel to the direction of working, together with the shaping of the component, is an important part of the forging process. Forging breaks down the original coarse grain size of the ingot structure, and thereby greatly assists grain refinement and the subsequent achievement of the mechanical properties when the forging is ultimately heat treated.

5.1. Reheating of ingot and/or billet.

Care must be exercised when reheating ingots that a very uneven fast heating rate is not employed: otherwise, severe internal cracking takes place, becauses stresses generated by fast uneven heating can give rise to multi-axial tensile stresses great enough to cause cracking at the ingot centre. In the early stages of heating an ingot above room temperature, thermal stresses must be kept low by limiting the rate of heating. When the temperature has been raised sufficiently, stresses will decrease rapidly due to plastic relaxation. In dealing with billet reheating, because an as cast structure is no longer present, even faster heating rates can be used. Whether ingot or billet heating, the maximum forging temperature is the same, 1240^{o}C. for nickel-chromium-molybdenum steels.

5.2. Hot Working.

When an ingot possessing dendritic crystal shapes is forged, the amount of deformation governs the degree to which these dendrites are transformed into fibres. During forging all the dendrites are affected. Non-metallic inclusions lying alongside the dendrites are deformed also. The resulting fibres (7) consist of elongated metallic crystals plus threads of impurities. This fibre, or grain flow, is important in the production of die blocks.
Two methods apply in die block manufacture -

(a) direct forging, (no upsetting).

(b) upset and cross forging.

The impression is normally machined parallel to the principal grain flow which must be controlled by the die block manufacture. When a deep-seated "V" impression is required in a block, it must be cross forged in order that service stress can be better withstood than would be the case, for example, with a direct forged die block.

5.3. Position in ingot.

Because ingots taper from the head to the ingot well, it follows that a non-uniform forging reduction will be given along the length of the ingot if tools with parallel faces are used. In carrying out this hot working operation, the forging press, using flat anvil tools, converts the tapered ingot into a square or rectangular billet. The prediction of spread and elongation of forging stock in flat tools was considered by Tomlinson & Stringer (17) who produced a Nomogram to enable forging reductions to be calculated.

5.4. Direct forging.

The direct forging from ingot to die block is the simplest type of forging operation. It is possible, however, that uneven forging may produce internal stresses. Cook (18) had two 6" square x 2'-0" long carbon steel ingots cogged down, one in four passes, using narrow bites in order to produce uniform internal deformation, and the other into two wide bites to produce uneven internal deformation. Uniform ductility was found throughout the ingot that was evenly strained. Cook related this effect to compression strain around the centre plane. Loria (4) investigated the mechanical properties of forgings having diameters ranging up to 35" and chemical composition of :

C : .21/.29%; Ni : .25/.30%; Mo : .40/.60%; and V : .06/.12% - and Wells (19) studied the effects of forging and rolling upon steel compositions ranging from a plain carbon steel to alloy steels in which the alloys varied from .25% to 4.5%. Both authors agreed that forging reductions above a 4:1 ratio (ingot area/forged area) are detrimental to ductility as measured by the reduction of area percentage obtained in test bars taken transversely to the principal direction of grain flow. Loria further stated that when forgings with a good transverse ductility are required, the forging technique should be sufficient only to break up the ingot structure to promote subsequent beneficial heat treatment, but little more, as deformation greater than this small amount confers no other benefits. This unconventional opinion of Loria on forging structure is a radical departure from the attitude generally adopted by many Metallurgists who still consider, when viewing metallographic specimens, that the presence of dendritic structures in steel forgings indicates that insufficient attention has been paid to the forging technique and invariably conclude that the physical properties so obtained are inferior.

5.5 Upset Forging.

Cook & Blythe (20) stated that rolled billets from which die blocks can be forged exhibit mechanical anistropy of severe proportions. These workers undertook forging trials on rolled billets produced from a 54-cwt. 19½" square ingot, containing : C .59%., Ni 1.36%., Cr .78% and Mo .30%. The billets were rolled to 4" square and, therefore, received a 20:1 reduction, (original ingot area to rolled area). A series of 8" lengths was sawn from the billets (2:1 ratio of height to width) and upset forged to various heights by :

(a) a single upset from 8" to 6" (25% reduction).

(b) a single upset from 8" to 5" (37½% reduction).

(c) a single upset from 8" to 4" (50% reduction).

then three upsetting operations were given upon new samples of 8" length billets, in which two additional forging sequences of re-forging to near the original length were carried out.

(d) three upsets from 8" to 6"

(e) three upsets from 8" to 5"

(f) three upsets from 8" to 4"

The samples were subsequently heat treated, and compared with the original bloom properties. The results indicated that progressive upsetting impairs the original good rolled bloom properties, measured in a longitudinal direction, with a corresponding improvement in the inferior transverse properties of the original rolled bloom. The reduction of area and notch strength of the forged pieces showed that the arithmetical improvement in the properties in the transverse direction was less than the deterioration of the properties measured in the longitudinal direction. Repeated upsetting, with the consequent drawing back to nearly original bloom length, accentuates the previously mentioned effects. Therefore, equalisation of longitudinal and transverse properties can be obtained by using a smaller upset, providing that upset forging is repeated. It was concluded in the forging trials carried out on the rolled bloom that a reasonably uniform condition can be produced by using three upset forging operations,consisting of 1.6:1 reduction, with the addition of intermediate longitudinal re-forging operations. What must be acknowledged with these tests is that the die blocks had barrelled sides. This shape would not be acceptable in industrial practice. Therefore, to square up the blocks, subsequent forging must take place on each die face. The finishing operation would mask some of the beneficial effects of original upsetting. In comparing the direct forged with the upset forging technique, it becomes apparent that some compensation for mechanical anistropy can be achieved when upset forging is carried out; furthermore, the factor of 4:1 stipulated as a maximum forging reduction on a direct forged component, need not apply when small upset repeated forging is practised.

5.6. Segregation effects.

One effect of forging upon the ingot structure is that banding can be produced from dendritic segregation of alloying elements, and residuals such as tin, phosphorus and arsenic. These dendritic areas, according to Smith (13) give rise to banding.

Quantitative evidence of macrosegregation and its influence on forgings produced from large ingots was investigated by Finniston & Fearnhough (6) on two nickel-chromium-molybdenum steel ingots of 20 and 37 tons weight respectively, having the following chemical analyses :

	20 tons ingot weight. %	37 tons ingot weight. %
Carbon	0.37	0.41
Silicon	0.24	0.19
Manganese	0.70	0.74
Phosphorus	0.035	0.021
Sulphur	0.034	0.031
Nickel	3.61	2.46
Chromium	0.17	0.13
Molybdenum	0.10	0.10
Vanadium	-	0.11

In these experiments a single direct forging operation resulted in a reduction of approximately 2:1. This unusually low deformation is the minimum that would generally be acceptable in a forging. The hardened and tempered samples showed that a hardness gradient across each segregate was higher than that existing in the surrounding matrix. Although the segregates had higher proof and ultimate tensile stress, their ductility and notch toughness was usually lower than the properties of the surrounding matrix. The segregates were found to contain more non-metallic inclusions, which the authors considered might account for the loss of notch toughness. In assessing the dendritic structures it was shown that metal in the interdendritic branches, or arms, had variations in composition and hardness comparable to those seen to be present in segregates. Commenting on the above results, Allsop (21) stated that transverse mechanical properties depend upon the characteristics of the segregate bands inherited from the cast structure. The variation in composition of the segregate bands is very important since they form an integral part of the normal macrostructure of all wrought objects, and these properties may determine the success or failure of parts subjected to transverse or circumferential stress either in testing or in service. Stevens & Thorneycroft (16) state that hot working only slightly reduces the intensity effect of microsegregation.

The contribution of non-metallic inclusions to the inferior mechanical properties associated with only a 2:1 forging reduction, as stated by Finniston & Fearnhough, requires further investigation. For example, the question may be posed whether non-metallic inclusions have a similar effect on properties when greater forging reductions are given.

5.7. Heat treatment.

The refinement of the as-forged structure can only occur when heat treatment is carried out. On nickel-chromium-molybdenum die blocks, grain refining, either by annealing or normalising, followed by oil hardening and tempering, is usually carried out. Banding exerts an effect upon hardenability which may increase or decrease the response of the forgings to heat treatment and subsequent machining.

When the best ingot casting and forging practices have been adopted it is still possible to produce banded structures. However, by forging, the grain size can be completely refined, carbide particles broken up, and chemical heterogeneity reduced simultaneously. All these effects facilitate the production of homogeneous austenite prior to quenching. These factors exert an influence upon the final microstructure in which tempered martensite will predominate, and so help to achieve uniformity of toughness and hardness. Heat treatment is in fact the most closely controlled operation in the whole process of die block manufacture.

Austenitising of No. 5. die blocks is carried out from 850°C., and they are subsequently tempered to yield hardness es within standard ranges which are given in Table No. 3.

Table 3. Effect of Tempering Times on Hardness.

Range.	Hardness B.H.N.	Tempering Temperature °C.
A	429-401	560-570
B	388-363	590-600
C	352-331	620-630
D	321-302	630-640
E	293-269	650-660
F	262-241	660-680

In practice, therefore, hardness results within any range, are closely controlled, as it is unrealistic to attempt to control the temperature distribution of large heat treatment furnaces to more close limits. The actual hardness ranges imposed are extremely narrow for such large products. However, the associated properties of ductility and toughness are possibly more closely related to service properties and will be investigated.

5.8 Forging Reduction.

The general impression obtained from the records of previous investigators (4) and (5) was that direct forging reductions above 4:1 are held to be harmful to transverse mechanical properties - this contention has been confirmed by practical experience.
To examine the variation of chemical heterogeneity with

forging reductions would throw additional light upon which hot working reductions are acceptable and, in particular, to note if dendritic patterns persist and what effect they have on mechanical properties. The width of the banded areas is also diminished with increasing hot working reductions. The effect this has upon the mechanical properties and more so upon the aspect of die life should be examined.

An American steelmaker (22) has shown that when vacuum re-melted steels are used, the degree of ingot segregation and non-metallic inclusions is much reduced and allows greater forging reductions to be given, without adversely affecting transverse mechanical properties. This is further evidence of the importance which non-metallics play in determining mechanical properties.

Keshian (23) in his extensive study on dendritic structures stated that whether or not dendritic steel is inferior to non-dendritic steel depends primarily upon the amount of heat treatment and hot working given. He concluded in comparing a 1% carbon steel, fully dendritic, with one which he terms "a non-dendritic steel" forged and heat treated to the same conditions, that there was no difference noted in mechanical properties. The microstructure, and not the macrostructure, is the dominant factor in the development of physical properties. This may not be applicable to nickel-chromium-molybdenum type steel. So far the discussion has centred around mechanical properties but, generally, the quality of die steel is judged by die life, and no literature is available to relate structure or mechanical properties to this important practical consideration. At the present time, a view prevalently held is that banded structureshave a detrimental effect upon die life, as it is considered that early crack initiation can take place within the segregate bands, thus rendering the die tool to premature failure. Works trials are, therefore, the only means of assessing the influence of these factors upon service performance, and should form the basis of the interaction of strength and structure on die properties.

5.9. <u>To summarise</u> :

It is an established fact that in the heavy forging industry, when examining specimens taken from heat treated forgings, ranging from constructional to high alloy content steels, the macro-structure can always reveal dendritic patterns. This is particularly noticeable in the nickel-chromium-molybdenum steel to BSS.224. No. 5. composition. On a number of occasions Metallurgists investigating die block failure have commented adversely upon the dendritic pattern, and have implied that an inferior forging technique, or that insufficient reduction, had been responsible for producing undeformed dendritic structures. As revealed by the literature survey, a distinct cleavage of opinion exists. It is fair to say that the majority of Metallurgists think that dendritic structures are symptomatic of inferior quality. Many variables can affect die block life, therefore, investigations on the significance which different forging reductions have upon die block properties, together with observing any variation in die block life, would be advantageous.

No published studies of the effect of forging on the mechanical properties of BSS.224 No. 5. die steel have been found in the literature. Furthermore, the relationship of the mechanical properties of die blocks to their service performance does not appear to have been recorded. The aim of this investigation is to remedy this deficiency by comparing structure and mechanical properties to the working life of No. 5. die steel tools produced under controlled Works conditions. The general outline of the work to be reported is as follows -

A 3-ton 9-cwt. No. 5. die steel ingot was forged into 18 insert die tools, using forging reductions of 2.92/1., 9.0/1., and 18.5/1. respectively. The inserts were heat treated to 363/388 B.H.N. The structure and mechanical properties of the inserts were examined, after which their drop-forging service life was determined.

6. EXPERIMENTAL.

A basic electric ingot, cast by The Low Moor Alloy Steelworks, Bradford, weighing 3-tons 9-cwts., and having a girth size of 24" square with an overall length of 67" was used for all the tests. The cast analysis was :-

C. %	Si. %	S. %	P. %	Mn. %	Ni. %	Cr. %	Mo. %
0.53	0.25	0.025	0.028	0.59	1.53	0.74	0.30

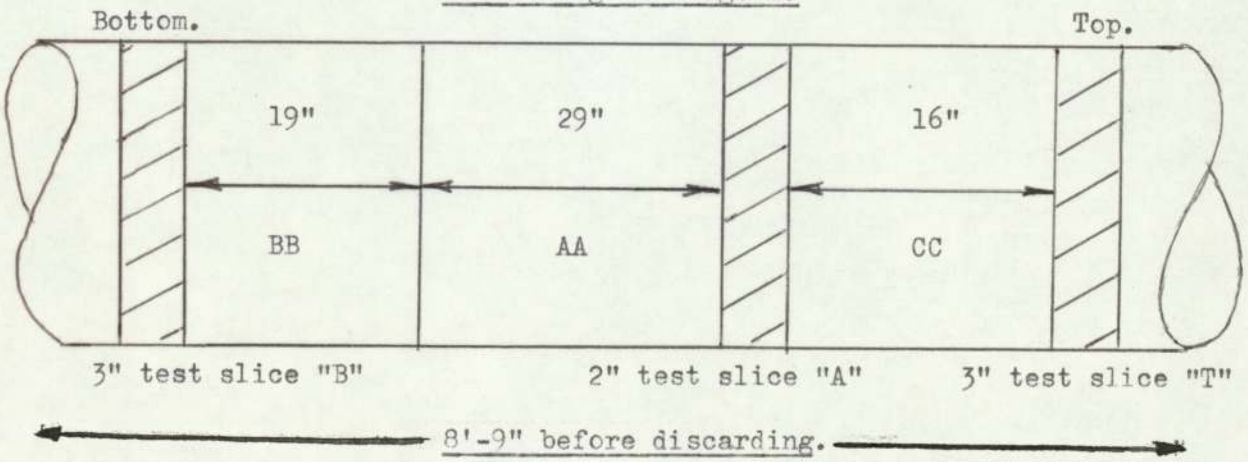
The ingot was charged in a furnace at 700°C. After allowing a 30-minute period for heat conduction by the cold ingot, a fast heating rate to the forging temperature of 1240°C. was used. This operation took 5 hours. A further 6 hours at 1240°C. was allowed to equalise the temperature throughout the ingot before forging. A single direct forging operation, converting the tapered 24" square ingot into a bloom, 14" square, then took place, (see Fig. 3). The bloom was then immediately charged into a heat treatment furnace, and the following thermal cycle was applied :-

Held at 650°C. for 7 hours. Withdrawn from the furnace, and the bloom air cooled to 400/300°C. Re-charged into the furnace, held at 300°C. for 5 hours, followed by heating to 850°C. in 8 hours, allowing 10 hours at 850°C. for temperature equalisation. The bloom was then furnace cooled to 650°C. in 6 hours. The final phase o heat treatment, consisting of maintaining the temperature of the bloom at 650°C. for 5 days, followed by air cooling, completed the treatment. The object of carrying out such a complex heat treatment schedule was to ensure that grain refinement and, more important, hydrogen diffusion, had been accomplished, thus making sure that the die steel would have freedom from hairline cracks.

6.1. Stage 1. Forging.

The ingot was cogged to 14" square. Samples were then taken at three sections of this bloom, namely top, termed "T", mid-chill, termed "A" and bottom, termed "B". It was decided that the blooms should be sectioned thus :

Sectioning of Stage 1.



Forging reduction 2.92/1
Fig -3a-

From the shaded areas shown in Fig 3a. slices "A", "B" and "T" were cut from which specimens were taken. The location of the test pieces in each slice is indicated in Fig 4.

The bloom was then sectioned by cold sawing. From the bloom section "AA" in which the original forging operation had resulted in a 2.92/1 reduction, six die blocks were sawn, numbered 1 to 6.

NUMBERING OF STAGE 1. FORGING.

No. 1.		No. 4.
No. 2.	Impression	No. 5.
No. 3.	faces	No. 6.

6.2. Stage 2. Forging.

Section "BB" of the 14" square bloom was heated to the forging temperature of 1240°C. in a reheating cycle, consisting of heating to 1240°C. in 6 hours, holding at 1240°C. for 5 hours, and then directly forging into a slab, 8" x 8" x 44". Subsequently, the slab was given a softening heat treatment in which full annealing from 850°C. was carried out.

From this 8" x 8" x 44" slab, which had received a 9/1 forging reduction (from the original ingot area), six die inserts were cold sawn, 14" x 8" x 3.5/8", and identified by cold stamping the numbers 7 to 12. The sampling positions for mechanical properties on Stage 2. forging are shown in Fig 5.

NUMBERING OF STAGE 2. FORGING.

No. 7.	No. 9.	No. 11.	Impression faces.
No. 8.	No. 10.	No. 12.	

Forging reduction 9/1.

6.3. Stage 3. forging.

The remaining "CC" section of the original 14" square bloom was reheated to the forging temperature of 1240°C. in 6 hours, held at 1240°C. for 5 hours, and subsequently forged into a bar, 8" x 3.5/8" x 7'-6" long. The heat treatment afforded to this reforged section was identical to that carried out on the reforged "BB" part bloom.

No. 13	No. 14	No. 15	No. 16	No. 17	No. 18

Forging reduction 18.5/1.

NUMBERING OF STAGE 3. FORGING.

The resultant bar, of cross section 8" x 3.5/8", had received an 18.5/1 forging reduction, relative to the original ingot. Six die blocks were sawn from the bar to a size 14" x 8" x 3.5/8" each, and identified 13 to 18. Fig 6 . indicates the position from which the test pieces were prepared.

6.4. Sample Die Tools.

All the 18 die insert blocks were finally heat treated simultaneously by oil quenching from 850/870°C., followed by tempering at 590/600°C. for 6 hours. Brinell hardness testing indicated that the desired "B" range of 363/388 B.H.N. had been achieved. Ultrasonic examination indicated that the forgings were free from abnormal defects.

Smethwick Drop Forging Company, Kidderminster, kindly agreed to evaluate the die life of the 18 experimental insert blocks. It is fortunate that at these Works there exists statistics on die block lives. A swivel pin drop forging was selected as the component which would be produced on the 18 experimental die blocks. (Fig 7). From the considerable data which Smethwick Drop Forging Company have upon the die life incurred in producing swivel pin drop forgings, the average number of drop forgings previously produced per die set was 5430.

METALLURGICAL TESTS.

7. HARDNESS TESTING.

After each hot working stage had been completed, the forgings were allowed to cool to ambient temperature. They were subsequently fully annealed by heating to 850°C., furnace cooled to 300°C. and re-heated to 850/870°C. Finally, oil quenching and tempering at 590/600°C. for 6 hours took place.

Brinell hardness testing, using a Jackman radial arm machine with 3000 Kg. load, and 10 mm. ball was employed to confirm that the stipulated hardness range of 363-388 B.H.N. had been achieved. Hardness testing took place at two sites on the longitudinal face of each forging.

Table 4. Brinell Hardness Results.

Forging Stage.	Forging Reduction.	Hardness.
Stage 1.	2.92/1	388/375 B.H.N.
Stage 2.	9.0/1	375/363 B.H.N.
Stage 3.	18.5/1	388/373 B.H.N.

The results in Table 4. show that the specified hardness had been achieved. From the fully heat treated die blanks, test slices of 2" and 4" thickness were cold sawn.

8. MACRO-ETCHING AND SULPHUR PRINTING.

The transverse faces of the test slices were ground and polished prior to etching. As the three specimens of Stages 1. and 2. forging were large, the edges were built up with Plasticine, thus allowing the etchant to be contained on the surface. The test slices taken from Stage 3. forging were smaller in area than those from Stages 1. and 2. and were, therefore, etched in dishes.

Humphrey's reagent, consisting of copper ammonium chloride 120 grms, hydrochloric acid 50 mls. and water 1000 mls., was used for macro-etching. Because copper precipitated from the solution very strongly, it was found advisable to omit the acid at the early stages of etching until a fairly thick but loose copper layer had been built up, when additions of hydrochloric acid were made. After 24 hours etching, the test slices were rinsed with water and dried. Contrast of the etched surfaces was obtained by lightly rubbing with 000 emery paper. The macro-structures of each forging stage shown in Figs. 8-10., were developed on the faces opposite to those which had been sulphur printed previously. The corresponding sulphur prints are shown in Figs. 11 & 12.

The developed structures show that increased hot working has caused refinement in structure, as revealed by the macro-etching, and further confirmatory evidence of this is seen in the sulphur prints. Remnant dendritic patterns are still visible even when an 18.5/1 reduction has produced considerable localised refinement associated with those areas which came into contact with the press tools when forging took place. Evidence of this refinement is shown in the sulphur print, Fig. 12. The sulphur print shows a normal sulphide distribution associated with satisfactory electrically melted killed steel.

9. SAMPLING PROCEDURE FOR MECHANICAL TESTING.

(a) Stage 1.

Hot working ratio : 2.92/1 (total reduction).

After the first forging reduction, three test slices were taken, as follows :

1. From the top end of the ingot - Position "T"
2. From the mid-chill or middle length of the ingot. - Position "A"
3. From the bottom end of the ingot - Position "B"

The exact sampling positions are shown in Fig 4.

Tensile, Izod, Charpy and Hounsfield test pieces were prepared. Longitudinal samples were coded L1, L2 and L3. Transverse samples were coded T1. Depth samples taken from positions at right-angles to the principal grain flow were coded D1.

From the sampling positions, T2. (transverse, and D2. (depth), Tensile, Izod and Charpy test pieces were machined. The Hounsfield tensile sample was omitted, because insufficient material remained to produce trepanned spark-eroded specimens. The lack of these two Hounsfield tensile test bars was not important, as fifteen other specimens were available to compare the properties of large and small test piece sections. The Stage 1. forging inserts were manufactured from Section "AA" of the 14" square bloom, as shown in Fig 3.

(b) Stage 2.

Hot working ratio : 9.0/1 (total reduction).

Fig 3. shows the sampling position of three test slices, after a section of the original 14" square bloom, termed "BB" had been re-forged to give a 9.0/1 reduction. The samples "A" and "B" shown in Fig. 3b. were taken from the ends of the 8" x 8" x 14" billet, whilst the slice "M" was taken from a mid position.

Sampling Positions.

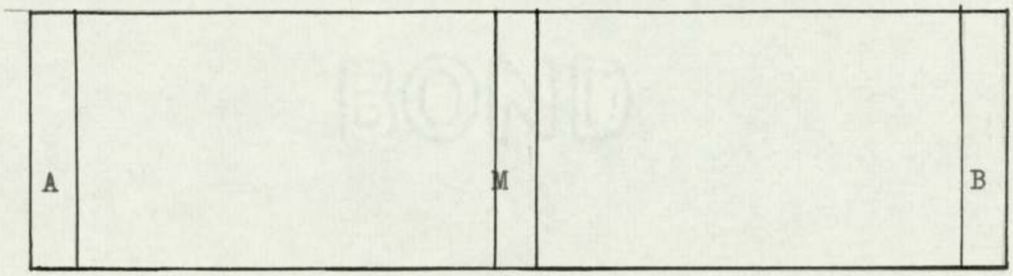


Fig - 3b -

An error was made in the calculation of the required billet length to provide Stage 2. forging, which accounts for the availability of the additional test slice (M). The inserts were stamped with reference numbers 7-12, after being cold sawn from the billet termed "BB".

(c) Stage 3.

Hot working ratio : 18.5/1. (total reduction).

Tensile and Izod test pieces were taken, again in three directions, Fig 6. Six die inserts, stamped with the numbers 13-18, were sectioned from this re-forged Stage 3. billet. This billet was forged from the original 14" square bloom termed "CC", Fig 3.

TEST PIECE DIMENSIONS.

Tensile test pieces were prepared in accordance with Appendix D. page 217 - British Standard 970 - 1955. The diameters of the test bars were either 0.564", 0.559" or 0.282" according to the availability of material.

When testing samples derived from the longitudinal direction, the smallest diameter test pieces had to be used, because they were also the shortest, as the test slices of Stages 1. and 2. were only 2" thick. Round tensile test pieces were tested on a 50-ton Denison Testing machine, by loading up to 0.2% proof stress level at a constant rate of 2-tons per minute. Thereafter the loading was increased to 6-tons per minute. Tensometer test piece No. 13, sample having a gauge diameter of 0.178/0.179", was prepared. Testing was carried out on the Hounsfield tensile testing machine, fitted with a 2-ton beam. Charpy and Izod notched bar impact test pieces were prepared, in accordance with Appendix D. of BSS.970. The Charpy test piece was 10 mm. square with a 2 mm. vee notch depth. The Izod test piece had a 0.12" notch depth with a 1.005" radius. Two or three notches were placed in each bar, depending upon the length of bar available. All impact tests were carried out at room temperature in a combined Avery Izod/Charpy testing machine.

TABLE NO: 5 MECHANICAL PROPERTIES OF STAGE 1. FORGING 2.92/1.

Test No.	Test piece. Diameter inches.	Test piece. Area sq.in.	Yield point tons/in^2.	.1% Proof stress tons/in^2.	.2% Proof stress tons/in^2.	Tensile stress tons/in^2.	Elongation %	Reduction of area %	Izod ft. lbs.	Average.	Charpy ft. lbs.	Average.
							1"					
AL. 1	.282	.0625	74.5	-	-	81.5	15	44.0	23.23	(23)	27.26	(26.5)
AL. 2	.282	.0625	72	-	-	79.5	15	39.2	22.20	(21)	23.19	(21)
AL. 3	.282	.0625	76.5	-	-	82	16	47.2	20.20	(20)	26.26	(26)
							2"					
AT. 1	.564	.25	-	68	69.5	78.5	8	15.2	10.13.13	(12)	15.14.13	(14)
AT. 2	.564	.25	-	66.5	69	78.5	9	20.0	11.12.13	(12)	13.17.16	(15.3)
AD. 1	.564	.25	-	69.5	70.5	80	11	24.8	13.11.13	(12.3)	17.17.14	(16)
AD. 2	.564	.25	-	69	69.5	79	9	18.4	12.13.14	(13)	14.14.18	(15.3)
							1"					
TL. 1	.282	.0625	76	-	-	82.5	14	44.8	15.16.13	(14.6)	19.18	(18.5)
TL. 2	.282	.0625	74	-	-	80.5	14	44.8	14.14.15	(14.3)	18.19	(18.5)
TL. 3	.282	.0625	76.5	-	-	82	14	50.4	10.12	(11)	20.20	(20)
							2"					
TT. 1	.564	.25	-	65.5	66.5	78.5	9	18.4	8.7.7.	(7.3)	11.10.12	(11)
TT. 2	.564	.25	-	67.5	69	79	9	18.4	8.8.11	(9)	17.13.12	(14)
TD. 1	.564	.25	-	70	71	81	7	15.2	8.9.9	(8.6)	13.12.10	(11.6)
TD. 2	.564	.25	-	68.5	71	80	9	12.0	11.8	(9.5)	16.11.10	(12.3)
							1"					
BL. 1	.282	.0625	73	-	-	79.5	14	47.2	23.24.23	(23.3)	24.24.	(24)
BL. 2	.282	.0625	71	-	-	77	14	39.2	13.12.15	(13.3)	14.14	(14)
BL.3	.282	.0625	70.5	-	-	78	14	39.2	10.10	(10)	18.19	(18.5)
							2"					
BT. 1	.564	.25	-	65.5	66.5	77	9	21.6	10.9.14	(11)	17.12.14	(14.3)
BT. 2	.564	.25	-	64.5	65.5	76	9	21.6	8.10.13	(10.3)	11.15.12	(12.6)
BD. 1	.564	.25	-	67.5	68	78	9	21.6	8.8.10	(8.6)	14.10.12	(12)
BD. 2	.564	.25	-	62	63	75	9	18.4	6.9.8	(7.6)	12.12.13	(12.3)

The gauge lengths of 1" and 2" have the same $4\sqrt{A}$ factor.

10. THE MECHANICAL PROPERTIES OF STAGE 1.

The mechanical properties of the three test slices, "T", "A" and "B" are given in Table 5. These sampling positions relate to the top, middle and bottom of the ingot respectively. The yield point quoted in Table 5., and in subsequent tables, is defined as the stress at which a distinctly visible increase occurs in the distance between gauge points on the test piece observed by using dividers, or the stress at which the load increases at a moderately fast rate. At such stress there is a distinct drop of the gauge pointer on the testing machine. The yield and proof stresses show minor fluctuations, whilst the tensile strength is generally uniform apart from specimen numbers BT2. and BD2.

The ductility and impact strength show variations which relate to the specimen testing directions of longitudinal, transverse and depth respectively. The longitudinal samples give the highest values. Three specimens TT1, TT2 and TD1 show a reduced ductility and notch toughness compared with the properties of the specimens originating from identical sampling positions of the other two test slices.

The results of test slice "A" show overall superiority for each direction of testing. However, this best combination of strength and toughness of test slice "A" shows that the transverse and depth samples are inferior to the longitudinal specimens, thus confirming that even when a small hot working reduction of 2.92/1 is applied, mechanical anisotropy can be encountered.

Table 6. MECHANICAL PROPERTIES OF STAGE 1. FORGING OBTAINED FROM HOUNSFIELD TEST PIECES.

SPECIMEN.	ULTIMATE TENSILE STRESS, TONS/IN^2.	ELONGATION %	REDUCTION OF AREA %
AL.1	81	15	35
AL.2	80	18	30
AL.3	80	18	25
AT.1	79	6	12
AD.1	79	10	15
TL.1	80	15	35
TL.2	81	13	30
TL.3	80	16	30
TT.1	79	9	15
TD.1	80	11	10
BL.1	81	14	36
BL.2	79	14	30
BL.3	80	18	30
BT.1	79	6	10
BD.1	82	5	15

10.1 THE EFFECT OF SIZE OF TENSILE TEST PIECE (STAGE 1).

Comparison of the results of the Standard and Tensometer tests show that no linear relationship exists between the Hounsfield reduction in area, and the large test piece reduction in area. It will be recalled that the large test pieces are of two different gauge diameters and are much larger in section than the Tensometer test pieces. As shown in Fig. 13. two groupings of reduction in area results occur - (a) above 35% reduction in area and (b) at the lower range of 15-25%. Only on six test piece tensile samples taken out of thirty is any approach to equality of reduction in area noted between the two test piece sizes.

TABLE 7. REDUCTION OF AREA %.

Sample.	Hounsfield.	Large round tensile test sample.
AT.1	12	15
TD.1	10	15
TT.1	15	18

The Ultimate Tensile Stress does not show a wide fluctuation, (Fig. 14), but there is, nevertheless, a small difference of the order of 5-tons/sq.in. It is considered that the difference in test piece size, together with the actual sampling site of the individual specimens, have contributed to these variations in mechanical properties, through the effect exerted by non-metallic inclusions especially, grain size, banding and general structure to a lesser extent. The variation of ductility reflects these differences. This effect will be considered later in more detail.

The difference between the results in corresponding tensile tests is unlikely to be due to inconsistencies in either test machines, because both testing units had been calibrated by the manufacturers immediately before testing commenced.

NOTCH TOUGHNESS OF STAGE 1. SPECIMENS.

When the results of the Izod and Charpy tests respectively are compared, only one sample AL.2. is seen to give an identical figure in both tests. The remaining 20 Charpy results are slightly higher than those of the corresponding Izod samples. The corresponding results of the Izod and Charpy are plotted in Fig. 15. No simple relationship between these results is evident. Fig. 16. gives some indication of the spread of r esults.

TABLE NO: 8. MECHANICAL PROPERTIES OF STAGE 2. FORGING 9/1.

Test No.	Test piece. Diameter inches.	Area sq. inch.	Yield point tons/in^2.	.1% Proof stress tons/in^2.	.2% Proof stress tons/in^2.	Tensile stress tons/in^2.	Elongation %	Reduction of area %	Izod ft. lbs.	(Average).
AL.1	.282	.0625	69	-	-	76	17	53.2	21.26.31	(26)
AT.1	.564	.25	-	68	68.5	77.5	7	12.0	12.9.10	(10.3)
AT.2	.564	.25	-	67	68	77	6	12.0	12.11.10	(11)
AD.1	.564	.25	-	67.5	68	77.5	7	15.2	12.10.8	(10)
BL.1	.282	.0625	73.5	-	-	80	15	42.0	23.24	(23.5)
BT.1	.564	.25	-	69	70	78	7	15.2	12.10.9	(10.3)
BT.2	.564	.25	-	69.5	70.5	78.5	8	13.6	11.11.10	(10.6)
BD.1	.564	.25	-	68.5	69.5	77.5	6	14.0	10.11.9	(10)
ML.1	.282	.0625	71.5	-	-	78.5	15	48.2	26.26	(26)
MT.1	.564	.25	-	69	69.5	77.5	6	12.0	10.10.12	(10.6)
MT.2	.564	.25	-	70	70.5	79	9	16.8	11.10.9	(10)
MD.1	.564	.25	-	68.5	69	78	9	16.8	10.11.9	(10)

.282 and .564 of 1" and 2" gauge lengths have the same $4\sqrt{A}$ factor.

10.2 THE MECHANICAL PROPERTIES OF STAGE 2. FORGING.

These properties are given in Table 8. on the previous page. The yield and proof stresses, and ultimate tensile stress, are uniform. Specimen AL1 had the lowest tensile stress of the twelve specimens. The ductility and impact strength is highest for the longitudinal specimens. The nine transverse and depth samples show remarkable uniformity of impact strength. This group of Izod results is the most consistent of the entire mechanical testing programme.

TABLE NO: 9. MECHANICAL PROPERTIES OF STAGE 3. FORGING 18.5/1.

Test No.	Test piece. Diameter inches.	Test piece. Area sq. inch.	Yield point tons/in^2.	.1% Proof stress tons/in^2.	.2% Proof stress tons/in^2.	Tensile stress tons/in^2.	Elongation %	Reduction of area %	Izod ft.lbs.	Average.
L.1	.564	.25	-	69	70.5	81	15	43.2	14.10	(12)
L.2	.564	.25	-	69	70	80	15	46.8	13.14	(13.5)
T.1	.564	.25	-	70.5	71	81	11	24.8	8.10.13	(10.3)
T.2	.564	.25	-	69.5	70.5	79	8	21.6	7.7.6	(6.6)
T.3	.564	.25	-	70	71	79.5	7	13.6	8.7.8	(7.6)
T.4	.564	.25	-	69.5	70.5	80	11	24.8	8.8.11	(9)
D.1	.282	.0625	73	-	-	81	6	11.2	7.8	(7.5)
D.2	.282	.0625	71	-	-	75	4	15.5	6.6	(6)

.282 and .564 of 1" and 2" gauge lengths have the value $4\sqrt{A}$.

10.3 <u>MECHANICAL PROPERTIES OF STAGE 3. FORGING.</u>

The properties are listed in Table 9. on the previous page. Whilst the proof and tensile stresses retain uniformity, apart from Specimen D2, a deterioration in the ductility of the transverse and depth specimens has occurred. The longitudinal ductility remains the highest. The lowest recorded tensile strength of 75 tons/sq.in. on sample D2. coincided with the rather high figure of 15.5% reduction in area. Notch strengths of the transverse and depth samples are lower than that of the longitudinal specimens.

TABLE -10-

STAGE 2. COMPARISON OF MECHANICAL PROPERTIES OF STAGE 2. FORGING WITH TEST SLICE "B"OF STAGE 1. FORGING.

STAGE 1.

Sample.	Tons/in^2 Tensile.	Reduction of area %	Izod. ft.lb.	Sample.	Tons/in^2 Tensile.	Reduction of area %	Izod. ft. lb.	Sample.	Tons/in^2 Tensile	Reduction of area %	Izod ft.lb.
				BL.1	79.5	47.2	23.3				
				BL.2	77	39.2	13.3				
				BL.3	78	39.2	10				
				BT.1	77	21.6	11				
				BT.2	76	21.6	10.3				
				BD.1	78	21.6	8.6				
				BD.2	75	18.4	7.6				
STAGE 2.											
AL.1	76	53.2	26	ML.1	78.5	48.2	26	BL.1	80	42	23.5
AT.1	77.5	12	10.3	MT.1	77.5	12	10.6	BT.1	78	15.2	10.3
AT.2	77	12	11	MT.2	79	16.8	10	BT.2	78.2	13.6	10.6
AD.1	77.5	15.2	10	MD.1	78	16.8	10	BD.1	77.5	14	10

10.4 COMPARISON OF MECHANICAL PROPERTIES OF STAGES 1. AND 2. FORGING. (TABLE 10).

Stage 1. test slice "B" was located next to the material of Stage 2. forging, so the properties are directly comparable. The strength and toughness of the longitudinal samples are the most superior of Stages 1. and 2. properties. In the transverse direction, the ductility is lower, otherwise the properties are the same. The Izod values in the depth direction are slightly greater, but the reduction in area is lower in Stage 2.

Generally the properties are uniform between these two stages, but the ductility is slightly lower in Stage 2.

TABLE -11-

STAGE 3. COMPARISON OF MECHANICAL PROPERTIES OF STAGE 3. FORGING WITH THOSE EXISTING ON TEST SLICE "T" OF STAGE 1. FORGING.

STAGE 1.				STAGE 3.							
Sample.	Tons/in^2 Tensile.	Reduction of area %	Izod ft.lb.	Sample.	Tons/in^2 Tensile	Reduction of area %	Izod ft. lb.	Sample.	Tons/in^2 Tensile	Reduction of area %	Izod ft. lb.
TL.1	82.5	44.8	14.6	L.1	81	43.2	12				
TL.2	80.5	44.8	14.3	L.2	80	46.8	13.5				
TL.3	82	50.4	11								
TT.1	78.5	18.4	7.3	T.1	81	24.8	10.3	T.3	79.5	13.6	7.6
TT.2	79	18.4	9	T.2	79	21.6	6.6	T.4	80	24.8	9
TD.1	81	15.2	8.6	D.1	81	11.2	7.5				
TD.2	80	12	9.5	D.2	75	15.5	6				

10.5 COMPARISON OF STAGE 3. MECHANICAL PROPERTIES WITH STAGE 1. (TABLE 11).

Stage 1. test slice "T" was adjacent to the material of Stage 3. forging, so the properties are directly comparable. The longitudinal samples of Stage 3. forging show a marginal decrease with the Izod values, compared with Stage 1., the remaining properties being constant. Stage 3. transverse and depth ductility are far less uniform.

TABLE -12- MECHANICAL PROPERTIES.

Sample.	Tons/in^2 Tensile.	Elongation %	Reduction of area %	Izod ft. lb.	Forging Reduction.
LONGITUDINAL.					
Stage 1. Test Slice "A"	81.5	15	44	23	2.92/1
	79.5	15	39.2	21	
	82	16	47.2	20	
Stage 2.	76	17	53	26	9.0/1
	80	15	42	23.5	
	78.5	15	48.2	26	
Stage 3.	81	15	43.2	12	18.5/1
	80	15	46.8	13.5	
TRANSVERSE.					
Stage 1. Test Slice "A"	78.5	8	15.2	12	2.92/1
	78.5	9	20	12	
Stage 2.	77.5	7	12	10.3	9.0/1
	77	6	12	11	
	78	7	15.2	10.3	
	78.5	8	13.6	10.6	
	77.5	6	12	10.6	
	79	9	16.8	10	
Stage 3.	81	11	24.8	10.3	18.5/1
	79	8	21.6	6.6	
	79.5	7	13.6	7.6	
	80	11	24.8	9	
DEPTH.					
Stage 1. Test Slice "A"	80	11	24.8	12.3	2.92/1
	79	9	18.4	13	
Stage 2.	77.5	7	15.2	10	9.0/1
	77.5	6	14	10	
	78	9	16.8	10	
Stage 3.	81	6	11.2	7.5	18.5/1
	75	4	15.5	6	

COMPARISON OF MECHANICAL PROPERTIES OF STAGES 2. AND 3. WITH STAGE 1. (TABLE 12).

Test slice "A" of Stage 1. forging represented the best overall properties of the initially cogged ingot. The properties of Stages 2. and 3. are compared with those of test slice "A" in Table 12. Longitudinal impact strength of Stage 3. is reduced to nearly 50% of that of Stages 1. and 2., but at the same time the other properties remain nearly uniform.

Stage 3. transverse ductility and impact strength show minor variations within the group of results.

The depth impact strength and ductility of Stage 3. show a decrease compared to Stages 1. and 2. The high reduction of area of 15.5% recorded in the Stage 3. depth sample is associated with a low tensile stress.

11. THE EFFECT OF FORGING REDUCTION.

When the mechanical properties attained at each forging stage are studied individually, the results show that the presence of anisotropy is confirmed. For example, the ductility and Izod values in the transverse and depth testing positions are lower than those existing in the longitudinal plane. This feature confirms the view previously expressed by Wells (19) who stated that there is an optimum forging reduction of approximately 3:1 when hot working in a parallel direction to the ingot axis. He found that with a forging reduction greater than 4:1 longitudinal ductility is little affected, but a decrease in transverse ductility occurs. Dieter, McCleary & Ransome (3) agreed with this view, but considered that the tensile ductility in the depth direction is lower than that present when testing in a transverse plane after forging has been hot worked by a straight-line reducing or slabbing operation.

The mechanical properties of Stage 1. indicate that after a reduction of 2.92/1 anistropy occurred, particularly with ductility and notch strength, on the transverse and depth samples. Variable results of the transverse and depth properties were brought about by forging deformations of 9.0/1 and 18.5/1 respectively. The deterioration in properties is not directly proportional to the increase in forging reduction, but the best properties, in general, are obtained with the smallest forging reduction used in Stage 1. The longitudinal properties of all three forging stages, excluding the impact strength of Stage 3. were remarkably uniform, and were of a high standard.

12. THE VARIATION OF BRINELL HARDNESS WITH TENSILE STRENGTH.

For these inserts the specification requirement for Brinell hardness is 363/388 B.H.N., which is equivalent to an approximate tensile strength of 81/87 tons/sq.in. according to published tables (25). This tensile strength has not been achieved by the majority of specimens tested from the three forging stages. For example, the values of Ultimate Tensile stress shown in Table 5. (excluding Hounsfield samples) have an average of 77.9 tons sq. in., which is significantly lower than the tensile stress requirements converted from the Brinell hardness of 363/388 B.H.N. As the original test slices from which tensile test pieces were prepared had previously been Brinell hardness tested, in which the specification requirement of 363/388 B.H.N. had been achieved, it appears that either the published conversion tables are wrong or that variations exist between the surface hardness of the test slices and the tensile strength of the samples selected beneath the area of Brinell testing. So that the relationship between Brinell hardness, Vickers hardness and tensile strength could be further assessed, fractured tensile specimens were Brinell or Vickers hardness tested. The results, when plotted, are shown in Fig. 17. and the following comments are relevant :

1. Only 8 tensile specimens achieved the specification range of 81/87 tons/sq.in.

2. Of the fractured tensile specimens, the specification range of 363/388 B.H.N. was achieved in every case.

3. Of the specimens, Vickers hardness tested only, 4 specimens tested failed to achieve the lower range of the specification, 384/411 V.P.N.

Nevertheless, when comparing the individual tensile strengths of specimens with Brinell or Vickers hardness, it is noted that variations occur which are not compatible with the conversions given in published tables (25), particularly on specimens :

<u>Stage 1.</u>	AT2.	TT1.	BL1.	BL3.	BD2.
<u>Stage 2.</u>	AL1.	AT2.	BT2.	MT1.	
<u>Stage 3.</u>	D2.				

Further evidence that whilst the Brinell hardness is within specification requirements, the tensile strength converted from the Brinell hardness gives varying values not consistent with data tables (25) is given by the following example :-

Of the 56 tensile samples selected from the three forging stages, (see tables 5-9), only 12 specimens conformed to the tensile strength converted from Brinell hardness. This inconsistent difference that has been found between measured values of Ultimate Tensile Strength, and those obtained from hardness conversion tables, indicates that in the higher strength ranges met in die block production, the same confidence cannot be placed on values of tensile strength converted from Brinell hardness in the 200/300 B.H.N. range. In practice, die blocks are supplied to a Brinell hardness range, and not to a specific tensile strength, so there is no danger of misrepresentation in properties. However, the likely errors in conversion have been well demonstrated by these results. Finally, it should be emphasized that the accuracy of the Brinell hardness machine was verified against a standard test block before these hardness tests were carried out. Since the Tensile testing machines had also been proved, the noted disagreement with the published conversion tables cannot be ascribed to inaccuracies in testing.

13. <u>CHEMICAL COMPOSITION.</u>

Chemical analysis was carried out by standard methods, (24) to verify the cast chemical analysis which was stated to be :

Carbon	.57%
Silicon	.25%
Sulphur	.025%
Phosphorus	.028%
Manganese	.58%
Nickel	1.53%
Chromium	.74%
Molybdenum	.30%

Samples for analysis were taken from the three transverse slices - "A", "B" and "T" of Stage 1. forging, (ref. Fig. 4), and the following designations were given :

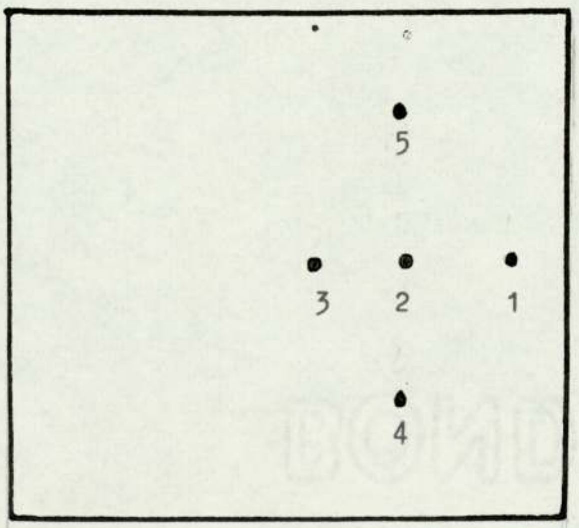


TABLE 13. RESULTS OF CHEMICAL ANALYSIS.

Sample No.		C.	Si.	S.	P.	Mn.	Ni.	Cr.	Mo.%
1		0.54	0.22	0.024	0.032	0.60	1.56	0.68	0.30
2	Test	0.54	0.21	0.025	0.029	0.60	1.55	0.71	0.30
3	slice	0.55	0.21	0.025	0.030	0.60	1.53	0.71	0.28
4	"A"	0.54	0.21	0.024	0.032	0.61	1.57	0.70	0.28
5		0.55	0.21	0.025	0.032	0.61	1.56	0.73	0.29
1		0.54	0.20	0.024	0.029	0.61	1.64	0.69	0.27
2	Test	0.53	0.21	0.025	0.031	0.60	1.59	0.71	0.29
3	slice	0.53	0.21	0.022	0.032	0.59	1.57	0.69	0.28
4	"B"	0.53	0.18	0.022	0.032	0.61	1.57	0.75	0.26
5		0.53	0.18	0.023	0.028	0.60	1.58	0.75	0.26
1		0.54	0.21	0.024	0.029	0.59	1.56	0.69	0.26
2	Test	0.54	0.21	0.024	0.030	0.60	1.58	0.71	0.26
3	slice	0.54	0.20	0.022	0.030	0.59	1.58	0.69	0.26
4	"T"	0.54	0.21	0.024	0.032	0.61	1.62	0.71	0.25
5		0.53	0.21	0.024	0.030	0.63	1.58	0.71	0.26

STATISTICAL ANALYSIS OF THE CHEMICAL COMPOSITIONS GIVEN IN TABLE 13.

1. Cast analysis.

C.	Si.	S.	P.	Mn.	Ni.	Cr.	Mo.	
0.57	0.25	0.025	0.028	0.60	1.53	0.74	0.30	%

2. Mean values.

0.54	0.20	0.024	0.030	0.60	1.57	0.70	0.27	%

3. Standard deviation.

0.0034	0.012	0.011	0.0014	0.01	0.026	0.022	0.016	%

It is concluded that the mean values of the chemical analyses are representative of the ingot, whereas the steelmakers cast analyses are taken from spoon samples and are not, therefore, truly representative of the chemistry of the ingots. Nevertheless, there is not a great variation.

Molybdenum is an element which is known to segregate markedly in steel ingots (13) and whilst the analysis of the 15 samples, (Table 13) of Stage 1. indicates slight variation in the percentage of molybdenum, compared with the cast analysis, it was decided to investigate the molybdenum content of specimens taken from Stage 2. and 3. test slices, as variation, if confirmed, of this element would give a positive indication of a general segregation pattern. Drillings for this analysis were obtained from fractured tensile test pieces, and analysis indicated :

	Sample.	Molybdenum %	Statistical analysis.
Stage 2.	MT.1	0.30	Mean : 0.28 %
	BT.2	0.29	
	AD.1	0.29	
	BL.3	0.27	
	AT.2	0.27	
Stage 3.	T.2	0.27	Standard deviation : 0.014 %
	T.4	0.30	
	D.1	0.30	

The molybdenum contents and statistical analysis of the 8 samples, compared with the statistical analysis (given on the previous page) of Stage 1. chemical composition, clearly indicates that only a slight variation has occurred. Because no significant variation with molybdenum is present with forgings manufactured from Stages 1. and 2. it can be assumed that this trend of little segregation for the other elements, known to segregate, will occur.

14. METALLOGRAPHY.

Preparation of samples.

From the material not used for mechanical testing, twenty-seven speimens were taken for micro-examination, covering all aspects of treatment. The locations of these sampling sites are shown by the holes trepanned by the spark eroding technique using the Agietron BQ. spark machining unit for Stage 1. (see Fig. 4). The chalk-marked spots in Figs. 5&6, indicate the sampling positions used in Stages 2. and 3. It was considered that the structure variations between columnar and equiaxed regions, and any difference between surface and internal transformation products of the quenched and tempered samples, would be fully covered by the choice of sample sites.

The specimens were prepared by wet surface grinding on successively finer grades of emery paper, finishing at the 000 grade, final polishing being accomplished with diamond paste of 1/5 micron particle size.

Oberhoffers reagent consisting of :

Cupric chloride	1 grm.
Stannous chloride	0.5 grms.
Ferric chloride	30 grms.
Hydrochloric acid	30 mls.
Water	500 mls.
Ethyl alcohol	500 mls.

was used to study the macrostructure of the 27 specimens.

This etchant gives good definition of dendritic structures in low alloy steels. Obberhoffer's reagent acts a little slowly when etching first commences. In etching, the precipitation of copper is not important, because the composition of the etchant was so chosen that deposition of copper on pure ferrite is just prevented. The areas which have electrochemically less noble properties, e.g. ferrite containing phosphorus, acquires an extremely thin film of copper over the surface, which protects it from further attack. Therefore, the surface stands in relief because the purest unprotected areas are dissolved by acid and ferric chloride. Further contrast is obtained by repeated etching and light polishing with a Selvyt cloth.

The best definition of the dendritic patterns was produced by first immersing the specimens in 5% nitric acid, in alcohol, followed by immersing the specimens in Obberhoffer's reagent. The structures present, after successive forging reductions i.e. Stages 1 - 3, are shown in Figs. 18-24. The persistence of "fir tree" dendritic structures, even with a forging reduction of 9.0/1 took place, is clearly discernible. The excessive hot working of 18.5/1 has aligned the dendrites to the direction of forging.

Microstructure.

Examination for non-metallic inclusions of the 27 specimens, which had been previously etched with Obberhoffers reagent, was undertaken at a magnification of x 100, after sample polishing had taken place. The microstructure was viewed at magnifications of x 100 and x 650 respectively, after etching with 5% Nital. Seven samples were selected, (Table 14), for detailed comment, as they were representative of the forging stages.

TABLE 14. SELECTION OF SAMPLES.

Sample.	Forging stage.	Deformation.
T4. Test slice "T"	1	2.92/1
A4. " " "A"	1	2.92/1
A1. " " "A"	2	9.0/1
C2. " " "C" (M)	2	9.0/1
C1. T1. and T2.	3	18.5/1

After sample preparation the seven specimens were re-examined in the unetched condition at a magnification of x 100. The inclusions present are shown in Figs. 25 & 26.

Manganese sulphide predominates over oxides and silicates. There is an apparent intergranular form of the inclusions in specimens selected from Stages 1 and 2. The high forging reduction of 18.5/1 of Stage 3. has resulted in the flattening and elongating of the inclusions in the direction of forging. After examination of the inclusion patterns, the 7 specimens were etched in 5% Nital.

ETCHED STRUCTURES.

Stage 1.

Sample T4. shows the appearance of segregated areas and the suggestion of banding is taking place, (Fig. 27), with the light and dark etching patterns. The structure is of tempered martensite and bainite, which is the lighter etching transformation product, as seen in Fig. 33 . A similar structure is seen in sample A4. (Fig. 28). The more acicular pattern of the tempered martensite and bainite in the darker etching band is seen in Fig. 34 .

Stage 2.

In specimens A1. and C2. a more pronounced form of banding than in Stage 1. has emerged. The concentration of non-metallic inclusions, lying in the light etching tempered martensite areas, is visible in Figs. (29) and (30). Such a field is indicated, Fig. (35), which shows elongated MnS. inclusions.

Stage 3.

In sample T2. the extreme forging reduction of 18.5/1 has produced severe banding, Figs. (32) and (39). The structure of tempered martensite and bainite is no more acicular than specimens examined from Stages 1. and 2. forging.

Previous investigators, Wells (19), Dieter, McCleary & Ransome (3), and Finniston & Fearnhough (6), state that inclusions predominate more in banded zones - this statement was confirmed. In order to examine this view in greater detail, four halves of broken tensile specimens were sectioned longitudinally, in order that after preparation the gauge length of the four test pieces, and the area where fracture took place, could be examined. Four samples of Stage 3. (L2; D2; T1; T4), were selected because the high forging reduction of 18.5/1 would have accentuated any banding and associated elongation of non-metallic inclusions. Additionally, these sampling positions apply to the three testing directions of longitudinal, transverse and depth.

The unetched areas, adjacent to the tensile fracture lip of the four specimens, are shown in Figs. (40-41). The number of non-metallic inclusions of the transverse and depth specimens are significantly greater than the number of inclusions present in the longitudinal sample near the fracture lip.
The longitudinal tensile test pieces gave the best ductility compared with the mechanical properties of the transverse and depth tensile test bars, the effect of fracture shape on the lip of the tensile test pieces depends upon the associated number of inclusions.

So that the dendritic structure associated with the fracture tensile lips could be viewed, the four samples were etched, (Figs. 42. and 43), with Obberhoffer's reagent. The structures revealed the difference between longitudinal and transverse grain fibre. The site of fracture, and associated ductility, clearly indicate that the measured mechanical properties and their variance are closely connected with the concentration and form of the non-metallic inclusions.

To assess further the level of steel cleanliness, specimens were taken from the heads of the longitudinal tensile specimens, from each forging stage.

Sampling positions.

Stage 1. forging. Tensile bar Nos. AL1, TL1 and BL1.
Stage 2. forging. Tensile bar Nos. AL1, ML1 and BL1.
Stage 3. forging. Tensile bar No. L1.

These samples were chosen so that the effect of forging reduction and position in the ingot could be assessed on the distribution of non-metallic inclusions. The Jernkontoret method of inclusion counting, as outlined in the American Society for Testing Materials Specification E.45 (26) was used. Seven samples, approximately ½" square, were sawn from the heads of the tensile test piece samples from each forging stage. The specimens were prepared in the normal manner for microscopic samples. Diamond paste of 1/5 micron particle size being used for polishing. Care was taken that the samples were not "pulled". The Jernkontoret method used in this inclusion assessment involves the examination of the whole of the unetched sample at a magnification of x 100. Each field view is compared with the standard reference chart, and the worst inclusion areas of the samples were recorded.

The Inclusion Standard Chart is divided into four sections :

A	for sulphides.
B	for alumina.
C	for silicates.
D	for oxides.

Each of these four grades is further sub-divided to distinguish between thin or thick inclusion type. The left-hand side of the Reference Chart is numbered 1-5, denoting inclusion density. The results of the Jerkontoret count are shown in Table 15. It is not possible to draw any firm conclusion on these results. It appears probable that the Stage 2. alumina and silicate inclusions have combined to give a total content of 3 alumina thin, because of the consolidation effect of the increased forging deformation given in Stage 3.

TABLE 15. JERNKONTORET INCLUSION COUNTS.

	Sulphides.		Alumina.		Silicates.		Oxides.	
	Thin.	Thick.	Thin.	Thick.	Thin.	Thick.	Thin.	Thick.
Stage 1.								
BL.1	1½	-	-	-	1	1½	2½	-
TL.1	1½	-	-	-	1	1½	2	1
AL.1	2	-	-	-	1	2	1½	1
Stage 2.								
BL.1	2	-	1½	-	-	-	1½	-
ML.1	2	1	1	1	1½	-	1½	1
AL.1	3	-	1½	-	-	-	1½	-
Stage 3.								
L.1	2½	-	3	-	-	-	1½	-

No manual method of inclusion counting has been devised which allows for operator reproducibility.

Hardy & Allsop (28) indicated that the three popular methods of inclusion counting, Fox, SAE and Jernkontoret, proved to give very poor reproducibility when Microscopists employed in six different laboratories examined the same specimens for inclusion counting. Care was taken that experimental conditions were completely standardised throughout the conducted survey. At a later stage of the study, the University acquired a Quantitative Television Microscope. (Q.T.M). The principle of this instrument is that an image is projected from a normal metallurgical microscope into a television camera, then to display the image on to a television monitor. The image characteristics can then be assessed and counted. Careful sample preparation is necessary, such as takes place with micro-specimens. It is possible with the Q.T.M. to area count oxides and sulphides individually or collectively, because these inclusions have different optical reflectivities from the matrix ; therefore, if an examination for MnS. is required, the selected sulphide area is produced on the television monitor, the instrument is adjusted so that only the oxides are viewed. A meter indicates the percentage of these oxide inclusions, and the instrument is then adjusted so that the field indicates the total inclusions present, which again is indicated by a meter reading. The difference between these two meter readings indicates the volume of MnS. inclusions. Because of the important part that inclusions play in determining mechanical properties, a more detailed examination of their distribution was undertaken on the seven specimens, and the Jernkontoret results are given in Table 14. These samples were examined with the Q.T.M. in order to assess total inclusion area density. A minimum of 100 area field counts were carried out on each specimen so that a statistical total percentage inclusion could be calculated. No differentiation between inclusion type, in the total area counting, was carried out. The shape factor of the inclusions for the six specimens was also determined for each forging stage, the shape factor being calculated thus :

$$SF = \frac{\sum Px}{\sum Py}$$

where Px = width of inclusion.
where Py = length of inclusion.

TABLE 16. Q.T.M. INCLUSION ASSESSMENT.

Specimens.	Total Area Count.	Standard Deviation.	Shape Factor.
Stage 1.	%		
BL.1	.135	Stage 1 = 0.089 %	.638
TL.1	.142		.690
AL.1	.141		.670
Stage 2.			
BL.1	.157	Stage 2 = 0.102 %	.625
ML.1	.139		.617
AL.1	.140		.620
Stage 3.			
L.1	.100	Stage 3 = 0.066 %	.546

The results given in the table show that -

1. The total area inclusion count is reasonably constant on specimens of Stage 1. and 2. However, sample L.1 of Stage 3. indicates a much reduced concentration of inclusions compared with the other six samples. It is considered that the sample site of L.1, which corresponds to the original ingot head area, may account for this reduced inclusion result.

2. The sample L.1 of Stage 3. forging, when Jernkontoret counting was carried out (Table 15), gave differing inclusion concentrations compared with the other specimens.

3. The lowest shape factor in Table 16. again concerns the Stage 3. sample - L.1. With the high forging deformation of 18.5/1, it is logical to expect a severe elongation of the inclusions in the direction of hot working.

By comparing the results given in Table 15. with the results of the total area inclusion count, expressed in Table 16, it appears that no satisfactory method of correlation between the Quantitive Television Microscope and the Jernkontoret methods can be made. It is considered that the total area inclusion counting by the Q.T.M. method is superior to other recognised inclusion assessment methods, as operator reproducibility can be expected, and many more fields can be conveniently and speedily viewed.

15. AUSTENITIC GRAIN SIZE MEASUREMENT.

Fifteen samples, approximately $\frac{1}{2}$" square, were cold sawn from the heads of tensile test pieces, covering the three forging shapes on which differing ductility results had been obtained in order to determine whether grain size varied in these specimens. The samples were prepared in accordance with specification E.19 (27)., published in the A.S.T.M. Standards Book, Part 111. 1958.

TABLE 17. GRAIN SIZE DETERMINATIONS.

Forging Stage.	Specimen No.	Tons/in^2 Tensile.	Reduction of area %	Izod ft. lb.	A.S.T.M. Grain size.
Stage 1.	AL.3	82	47.2	23	5-6
Stage 1.	AT.1	78.5	15.2	12	3-5
Stage 1.	AT.2	78.5	20.0	12	4-6
Stage 1.	AD.1	80	24.8	12.3	5-6
Stage 1.	AD.2	79	18.4	12.6	3-5
Stage 1.	BL.1	79.5	47.2	23.5	5-6
Stage 1.	BT.2	76.0	21.6	10.3	4-5
Stage 1.	BD.2	75.0	18.4	7.6	4-6
Stage 1.	TL.3	82	50.4	11	4-6
Stage 1.	TL.2	80.5	44.8	14.5	4-5
Stage 1.	TT.1	78.5	18.4	7.3	4-6
Stage 1.	TT.2	79	18.4	9	4-5
Stage 2.	BT.1	78	15.2	10.3	4-6
Stage 3.	D.2	75	15.5	6	5-6
Stage 3.	T.3	79.5	13.6	7.6	4-5

Variable grain size is present in all fifteen specimens.

Samples AL3, BL1 and TL3 - the estimated grain size is A.S.T.M. 4-6. Variations ranging from grain size 3 to A.S.T.M. 6 occur in specimens from tensile samples, tested in the transverse and depth directions. There is no definite relationship established between grain size, tensile strength and toughness, as demonstrated in Table 17. Variable grain sizes of specimens AT1, AT2, AD1, AD2 and TL3, are shown in Fig.44.

The majority of grain size determinations were carried out on Stage 1. samples. The results on Stages 2. and 3. specimens approximate to the Stage 1. results.

If a deliberate addition of grain refiners, such as aluminium or vanadium, is used in conjunction with the final deoxidation practice in steelmaking, so that a fine grain size can be achieved, the steelmaker always insists on a specification range of at least 4 grain sizes, for example A.S.T.M. 5-8. The 3-ton 9-cwt. ingot did not have any grain size requirements specified, even so the results given in Table 17. indicate that some grain control was exercised.

16. VICKERS HARDNESS TESTING OF MICRO-SAMPLES.

Because of the discrepancy between the series of Vickers hardness tests and the Brinell hardness, after appropriate conversion as discussed in Section 12, this effect was further investigated by testing the 27 specimens previously used for metallographic examination and reported in Section 14.

Re-polishing of the etched specimens was undertaken before hardness testing. The results are given in Table 18. and give rise to the following comments :-

According to published conversion tables (25) the Vickers hardness range equivalent to 363/388 B.H.N. is 384/411 V.P.N. The final column of Table 18. records that only 7 samples conformed to the specified hardness limits. Fig 45. indicates the variation of hardness results, in which 18 specimens gave hardness values below the specified minimum of 384 V.P.N. The sampling sites of the 27 specimens were different from those of t ensile samples referred to previously in Section 12. To investigate more closely the variation in Vickers hardness of the 27 specimens, micro-hardness testing on selected samples was undertaken so that the possibility that localised variations in hardness were responsible for the discrepancy between the "macroscopic" brinell test and the smaller impression Vickers hardness results could be investigated.

TABLE 18. VICKERS HARDNESS.

Sample No.						Average Hardness	Specn "B"
	Stage 1."A".						
A1	374	370	368	366	370	369	O
A2	378	367	370	373	373	372	O
A3	374	363	370	376	376	371	O
A4	376	370	370	367	376	371	O
A5	374	368	378	385	384	377	O
	Stage 1."B".						
B1	372	373	375	372	372	372	O
B2	378	378	370	380	379	377	O
B3	384	397	400	397	388	392	I
B4	391	376	372	385	387	382	O
B5	375	378	374	375	374	375	O
	Stage 1."T".						
T1	368	368	374	369	370	369	O
T2	362	371	371	367	367	367	O
T3	385	373	386	379	383	381	O
T4	370	360	362	369	369	366	O
T5	374	365	362	363	368	366	O
	Stage 2."A".						
A1	398	402	396	396	387	395	I
A2	381	385	393	393	393	389	I
A3	389	398	393	393	398	394	I
	Stage 2."B".						
B1	402	411	411	415	406	409	I
B2	376	387	387	391	396	387	I
B3	396	409	404	413	409	406	I
	Stage 2."C".						
C1 (M.1)	373	370	368	371	378	372	O
C2 (M.2)	374	377	375	375	370	374	O
C3 (M.3)	377	377	381	376	381	376	O
	Stage 3.						
T1	374	379	373	376	379	376	O
T2	368	371	371	372	374	371	O
1C	372	372	372	369	370	371	O

363/388 B.H.N. ("B" hardness) = 384/411 V.P.N.

O = Outside specification.
I = Inside specification.

17. MICRO HARDNESS TESTING.

Four samples were chosen for micro-hardness testing. These samples differed widely in sampling position hardness and different forging deformation :

Stage 1. Samples A3. and T3.
Stage 2. Sample B3.
Stage 3. Sample 1C.

These four samples had been previously used in the microscopic examination of the general structure. So that the dendrites could be more readily viewed, when micro-hardness testing, the specimens were re-polished and etched with Obberhoffer's reagent. A Leitz micro-hardness testing instrument was employed to hardness examine individual dendrite splines, using a load of 200 grms.

Stage 1. T3.	Stage 1. A3	Stage 2. B3	Stage 3. 1C.
430	386	412	386
427	362	386	410
453	412	358	412
458	362	473	374
441	362	412	348
405	376	457	399
441	399	374	412
449	399	374	410
419	399	348	441
428	386	426	
441	426	386	
418	386	362	
443		399	
412		337	
443			
432			
438			
473			

Because of the large variations in micro hardness, a dendritic arm on sample T3. was systematically tested on a raster, .029" x .012", as shown below :-

418	362	370
367	370	362
406	386	426
362	390	374
	386	

.029"

.012"

As the results obtained in both normal Vickers and micro-hardness testing fluctuated widely, a statistical examination, involving the "Students" T. Distribution, and the variance ratio or Snedecors F. t est, was carried out on the micro hardness values. These calculations are summarised :

Sample Position.		Mean.	∅	Variance S^2	Sum of Squares.
T3	Stage 1.	436.2	17	284.4	4834.5
T3	Raster.	386.1	12	442.1	5304.8
A3	Stage 1.	387.9	11	415.2	4566.9
B3	Stage 2.	393.1	13	1580.8	20509.7
1C	Stage 3.	399.1	8	717.4	5738.9
			61		40954.8

Therefore, if all these variances are estimates of one single variance, the best variance would be :

$$\frac{40954.8}{61} = 671.4$$

applying the x^2 factor -

(γ^2) $$x^2 = \frac{\text{sum of the squares found}}{\text{variance expected}}$$

on n-1 degrees of freedom.

The probability (P) that the sum of squares actually found would exceed 671.4 is given :

Sample.	Sum of Squares.	x^2	∅	P	95% Confidence Limits.	Mean.	95% Limits.
T3. Stage 1.	4834.5	7.05	17	0.984	$\pm$ 8.4	436.2	427.8 - 444.6
T3. Raster.	5304.8	7.90	12	0.78	$\pm$12.7	386.1	373.4 - 398.8
A3. Stage 1.	4566.9	6.80	11	0.82	$\pm$12.9	387.9	374.0 - 400.8
B3. Stage 2.	20509.7	30.6	13	0.004	$\pm$22.2	393.1	370.9 - 415.3
1C. Stage 3.	5738.9	8.55	8	0.4	$\pm$20.6	399.1	378.5 - 419.7

What can be deduced statistically from this survey is that in the case of sample B3. Stage 2., an unlikely situation has arisen in that the variance is different. 95% Confidence Limits were calculated for the four specimens, and it was concluded that :

1. The results are more variable on some samples than others.
2. The sample T3, Stage 1, has higher general hardness numbers than the other samples.
3. Excluding T3, of Stage 1., there is some degree of uniformity with the three remaining results.
4. The hardness variations of the specimens A3, B3. and 1C. follow a similar erratic pattern, being non-systematic from point to point.

In the 53 micro-hardness readings taken on the four specimens, the values of 386 and 399 V.P.N. occur eight and five times respectively; this is surprising in view of the wide spread of results.

5. By comparing the Vickers hardness with the micro hardness -

Specimen.	Average Micro hardness.	Average Vickers hardness	Difference.
Raster T3	386	381	
T3	436	381	-55
A3	387	371	-16
B3	393	406	+13
1C	399	371	-28

It can be seen that Sample T3 Raster average hardness can be equated to the average Vickers hardness, however, the overall spread of micro-hardness within the Raster ranged from 362-426 VPN.

As specimen T3. gave the highest variation in average micro-hardness, compared with the average Vickers hardness, verification of the accuracy of the Leitz micro-hardness testing instrument was undertaken, together with re-testing of the sample. Similar hardness values were observed.

In three out of the four samples, the highest average hardness was obtained by micro-hardness testing, signifying that this latter hardness test must be more sensitive to hardness variations than the normal Vickers hardness tests. No report of this effect was found in the literature.

18. MICRO-PROBE ANALYSIS.

The results of the chemical analyses discussed in Section 13. indicated that severe macrosegregation was not present. However, the widely erratic hardness variations, when testing individual dendrite splines reported in Section 17., suggested that segregation persisted, if only on a limited scale.

To investigate more deeply this hardness variation, Specimen No. T3. was selected for micro-probe analysis. In choosing this sample for examination, it was considered that with the least forging reduction of 2.92/1, the dendrites would not be substantially deformed, thereby facilitating a positive area determination of the identification of the dendrites during probe analysis. Any effect, for which a lack of forging deformation could be responsible, would be at its greatest in this specimen.

The electron probe x-ray micro analysis was carried out using a Cambridge Microscan instrument.

The area of specimen T3. on which probe analysis was carried out, was marked by a scratch indentation for reference purposes. X-ray photographs of the elements, nickel, chromium, manganese and molybdenum, and their concentrations, are given in Fig. 46. Sections of the individual element trace scanning charts, obtained during analysis, are shown in Fig.47 . The results confirmed that microsegregation is present, to the following extent :

Nickel. > Chromium. > Manganese. > Molybdenum.

This surprising result can only suggest that the variations in micro-hardness noted with specimen T3. are not, in any way, associated with chemical segregation, and that there is very little segregation at all, even on a micro scale.

To attempt to explain why this hardness variation exists, in the absence of segregation, is difficult, and emphasis might be directed to the structural conditions. Speculation upon whether there is any difference in hardness of martensitic or bainitic needles may be a significant feature to be considered, for the positive discrimination between these acicular transformation products is by no means as simple as some observers suggest.

Thorneycroft & Steven (16) showed that when specimens from an annealed En.24. ingot were hardness tested, variations between columnar and equiaxed grain structures occurred. After applying a diffusion heat treatment these variations in hardness became less marked. The specimen hardness ranged from 600 V.P.N. in the columnar crystals, to 300 V.P.N. in the equiaxed zone, with the annealed samples and, after diffusion annealing the columnar and equiaxed crystals gave hardness variations in the order of 500-600 V.P.N.

Cairns & Charles (29), in a survey of ferrite formed by diffusion controlled growth, noted in specimens which had received different heat treatments of either cooling in air or quenching from the same temperature, after etching gave structural differences in which variations in hardness were observed. These hardness variations were attributed to the changing solute conditions during diffusion growth. The extent of this growth in the columnar grains being observed by a dark line etch effect at the tip of the columnar crystal. However, when the ferrite was transformed to quenched martensite this dark line etch effect disappeared, whilst the hardness variation remained.

These hardness variations sometimes accompany the solute changes, sometimes not, so that the hardness variations occur without visible explanation.

Only limited diffusion takes place with the re-heating and forging of No. 5. die steel, based upon industrial processing practice, and in no way compares with the homogenizing diffusion annealing suggested by Thorneycroft & Stevens (16). It appears that a micro-structural phenomena reported by Cairns & Charles (29) may also be present with No. 5. die steel.

19. WORKS TRIALS.

Statistical information was supplied by Smethwick Drop Forging Company on the insert die life experienced prior to the undertaken trials, the number of drop forgings produced from 31 actual die runs was 168,283 forgings, the standard deviation being 1021.

The six inserts produced from each of the three forging stages gave die lives -

TABLE 19.

Stamping No.	Drop-forgings made.	Average.	Standard Deviation.	
208	4125			
209	6006	5710	1192	Stage 1.
210	7000			
211	6830			
212	6138	5614	1262	Stage 2.
213	3876			
216	6834			
217	5820	5786	853	Stage 3.
218	4705			

By applying a 95% confidence limit calculation, the following is pertinent (Table 19):

1. On the existing die life prior to the trials in which the standard deviation was 1021, the resultant life could vary between 3141 and 8057 drop forgings.

2. On the results achieved in the trials -
   Stage 1. die lives could vary - 3347 to 8073 forgings.
   Stage 2. " " " " - 3141 to 8057 forgings.
   Stage 3. " " " " - 4116 to 7456 forgings.

Therefore Stage 3. represents the least spread in the number of forgings produced.

CONCLUSIONS.

1. Mechanical Properties.

The mechanical properties developed by forging in increasing amounts show that :-

(a) Increasing forging deformation from 2.92 to 18.5/1 does not impair the longitudinal strength and toughness.

(b) Non-metallic inclusions significantly contribute to the deterioration of properties selected from the transverse and depth testing directions, compared with the properties of the longitudinal plane of sampling.

(c) The relationship between increasing forging reduction and impact strength was not as marked as previous investigators (5) have found. This work has demonstrated that the small 3-ton 9-cwt. electrically melted No. 5. die steel ingot was not badly segregated. It is inevitable that differing conclusions will be reached between investigations concerning impact strength on forgings processed from varying sized ingots in which differences in steelmaking practice may have occurred.

(d) No uniform relationship between the properties measured with the large size tensile test pieces and the strength and toughness obtained with the Hounsfield tensile samples exists, although both specimens were machined from contiguous sites.

(e) It is equally not possible to compare directly the results of the Izod test with those existing with the Charpy Vee test specimen.

(f) No accurate relationship between tensile strength and Vickers hardness, converted from Brinell hardness, exists when the comparison is made at ultimate tensile strengths greater than 80 tons per sq. in. The conversions in published hardness tables (25) associated with the range 363/388 BHN on tensile strength and Vickers hardness have not been confirmed. This investigation has consistently shown that inaccuracies of conversion data presented by standard hardness conversion tables, within the Brinell hardness range of 363/388 occur.

2. Non-metallic inclusions.

It has been confirmed that non-metallic inclusions play a prominent part in reducing the level of ductility, particularly when testing in the transverse and depth directions compared with the properties existing in the longitudinal plane. It has also been demonstrated that the recognised methods of inclusion assessment suffered from a lack of operator reproducibility. The latest Q.T.M. examination offers the most consistent method of determining inclusion contents.

3. Chemical Segregation.

Segregation of carbon, manganese, nickel, chromium and molybdenum, in forgings produced from the 3-ton 9-cwts. ingot, was small. In fact it is considered that this ingot, admittedly of small size and weight, showed remarkable freedom from segregation.

The variations in micro hardness found to be present when testing dendrite splines are still considered by many workers to be the result of segregation by alloying elements : micro-probe analysis has not confirmed this view.

4. Hardness.

It has been considered by Metallurgists generally that widely fluctuating hardness, within a small area of measurement, is attributable to structure or segregation. The influence of segregation upon this hardness variation has already been dealt with. The possible effect that structure might exert upon the variability of hardness has already been discussed in Section 17. Nevertheless, it is not possible to state what is the true origin of these hardness variations.

5. The effect of the treatment on die performance.

It was noted that before the evaluation of the trial insert a large spread in the number of drop forgings actually produced with the identical drop stamping practice was recorded by the drop forge. Statistical analysis of these prior die lives suggested that a range of drop forgings from 3141 to 8057 per die set could be expected. The die lives achieved by 18 trial inserts are given in Section 19. It can be seen that the increased forging reduction, and its influence upon mechanical properties, has not drastically affected die insert performance, inasmuch as the forging reduction given to the Stage 3. inserts of 18.5/1 has inferior mechanical properties to the forged inserts of Stage 1. which received a 2.92/1 reduction.

Statistical analysis of the die insert lives during trials indicates that Stage 3. represents the lowest spread in results, thus :

Stage.	Reduction.	Die Life 95% Confidence Limits.
1.	2.92/1	3347-8073 drop forgings.
2.	9.0/1	3141-8057 " "
3.	18.5/1	4116-7456 " "

A possible explanation for the least spread in results of the more heavily forged steel of Stage 3. die inserts may be found in results of Reynolds (30) which were concerned with studies of the hot workability of Mild Steel. Briefly his experiments consisted of measuring the number of twists which heated torsion specimens could withstand before fracture : he found that specimens derived from steel that had been heavily reduced by hot working sustained far more twists before failure than those made from lightly worked steel. The different reaction of the two types of material was ascribed by Reynolds to elongation of the inclusions in the former specimens. In drop-forging, the impression surface of a die is undoubtedly deformed during the production of drop forgings and, since failure is always preceded by cracking of the surface, to a greater or lesser degree, materials which are inherently more resistant to crack initiation and propagation under hot working conditions will last longer.

It is interesting to speculate that if the hypothesis of Reynolds applying to Mild Steel specimens could also be extended to include No. 5. die steel, in which the elongation of inclusions is beneficial, then large amounts of forging reductions would be indicated with die block material, even though the mechanical properties, measured at room temperature, may be inferior.

It must be concluded that the measurement of mechanical properties is not necessarily indicative of die insert performance during drop forging. The life of a die insert depends upon more subtle parameters than these simple tests are capable of providing.

6. Possible Future Developments.

1. This study was conducted by producing relatively small forgings of insert size, manufactured from a 3-ton 9-cwt. ingot, costing £198. In addition, however, costs were accrued in processing and testing. In suggesting that future work in evaluating the properties and die life characteristics on medium size die blocks of 40" x 20" x 18" should be undertaken, would be the next logical step - the greatest barrier to its implementation would be the cost. To increase the ingot size would mean spending an additional sum of £1158. on the ingot price alone, excluding all the attendant increased processing costs. It would, however, be interesting to

learn if the conclusions reached, relating to the non-segregated material processed from the 3-tons 9-cwt. ingot, apply to material processed from larger ingots in the order of 24 tons weight.

2. An examination of the effect of reducing the content of non-metallic inclusions by using material processed from a consumable vacuum electrode No. 5. die steel would confirm certain of the important conclusions reached in the present investigation.

3. To determine which properties are required in a die block, in this connection, it is suggested that one field of research which is worthy of study would be to amass fracture mechanics data on No. 5. die steel and attempt to equate the results with localised cracking found in some areas of all die impressions. For example, by using a Charpy specimen, and testing at various hardness levels at elevated temperatures, would result in a better understanding of crack growth in die block impressions observed during the course of drop forging.

4. Hardness. The perplexing factor of not being able to accurately define the origin of the hardness variations in the absence of chemical segregation, together with any differing structural conditions, requires further work. In order to elucidate this phenomena, it may well be that by subjecting

specimens of No. 5. die steel to a diffusion controlled growth, and noting the hardness levels, after various heat treatments have been carried out, may assist in the better understanding of this hardness variation.

<u>REFERENCES.</u>

1.	Plockinger E.	Iron & Steel Inst. Special Report No. 77. 1963, 51-56.
2.	Whittaker R.	Iron & Steel Inst. Special Report No. 77. 1963, 36-39.
3.	Dieter G.E., McCleary D.L. & Ransome J.T.	Metallurgical Society of A.I.M.E. 1958. <u>3</u> p. 101-140. Discussion p. 141-142.
4.	Loria E.A.	Blast Furnace & Steel Plant, 1936, <u>44</u>, p. 315-321.
5.	Wells C. & Mehl R.F.	Transactions of the American Society for Metals, 1949, <u>XL1</u>, p. 715-805.
6.	Finniston H.M. & Fearnhough T.D.	Journal Iron & Steel Institute 1951. <u>169</u>. p. 5-12.
7.	Aitchinson L. & Pumphrey W.I.	"Engineering Steels". McDonald & Evans, 1953. p. 44 & 57.
8.		Iron & Steel Institute. Special Report No. 16. 1937. p.6.
9.	Winegard W.C.	Institute of Metals. Monograph & Report Series, 1964. p. 1, 2 & 35.
10.	Spretnak J.W.	Transactions of the American Society for Metals 1948, <u>XL</u>. p.897-916. Discussion p. 917-921.
11.		Iron & Steel Institute. Special Report No. 12. 1936. p. 4-6.
12.	Rogues C. Martin P. Dubois C. & Bastien P.	Revue de Metallurgie, 1960, <u>57</u>, p. 691-1105.
13.	Smith T.B.	Iron & Steel, 1964 Nov. p.536-541.

14. Smith T.B., Thomas J.S. Goodall R. — Journal Iron & Steel Institute, 1963, 201, p. 602-609.

15. Johnson R., Middleton J.W., & Ford D. — Iron & Steel Special Report No. 89. 1965. p. 34-51.

16. Stevens W. & Thorneycroft. Journal Iron & Steel Institute, 1957. 187. p. 15-32.

17. Tomlinson A. & Stringer J.D. — Journal Iron & Steel Institute, 1959. 193. p. 157-162.

18. Cook P. — Journal Iron & Steel Institute, 1955. p. 250-252.

19. Wells C. — Metallurgical Society of A.I.M.E. Interscience Publishers, 1962. p. 209-232.

20. Cook P. & Blythe A.J. — B.I.S.R.A. MW/F/12/55. No. 70. 1955.

21. Allsop H. — Journal Iron & Steel Institute, 1952. p. 221-225.

22. Private communication from The Heppenstall Company, 1964.

23. Keshian H.G. — American Society for Steel Treating, 1936, 17/3. p. 321-371. Discussion p. 371-382.

24. "Standard Methods of Analysis". The United Steel Companies - 1951. p. 13-130.

25. American Society for Metals Handbook, 8th edition, 1961. 1. p. 1234-1236.

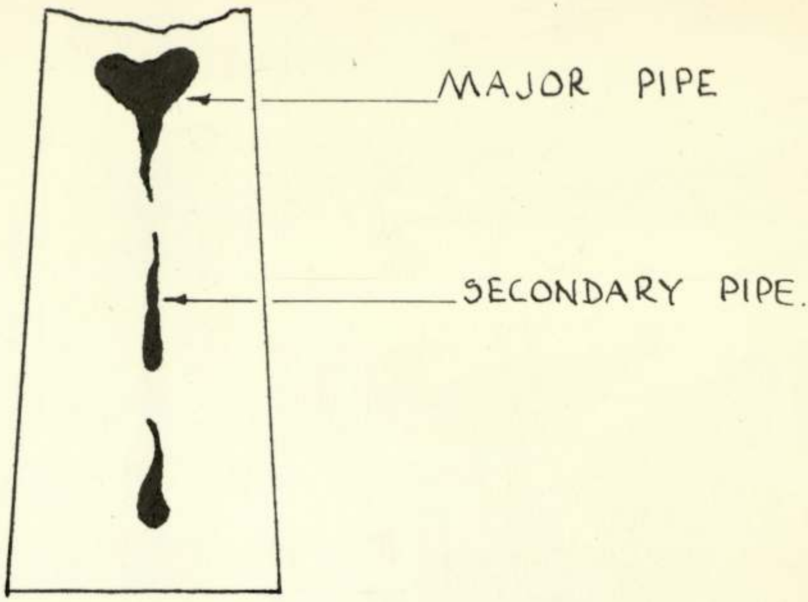
26. American Society for Testing Metals, (Materials). Specification E.45. 1958. 3. p. 529-540.

27. American Society for Testing (Materials). 1958. 3. p. 488-494.

28. Allsop R.T. & Hardy J.V. Journal Iron & Steel Institute, 1960. 3. p. 302-306.

29. Cairns R.L. & Charles J.A. Journal Iron & Steel Institute, 1965. November. p. 558-562.

30. Reynolds K. Private communication. University of Aston in Birmingham.



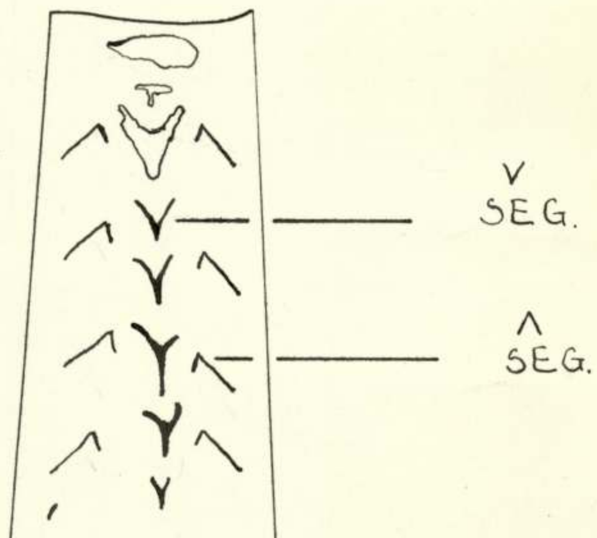
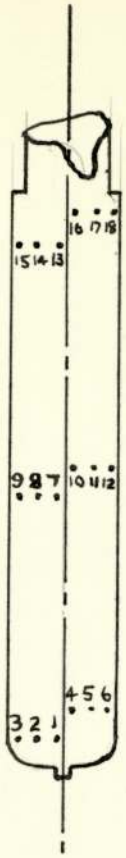


Fig -1-



DETAILED ANALYSIS OF INGOT.

Position No.	C.	Mn.	P.	S.	Si.	Cr.	Ni.	Mo. %
1	0.33	0.76	0.027	0.019	0.17	0.59	1.56	0.25
2	0.35	0.76	0.027	0.016	0.18	0.61	1.57	0.23
3	0.33	0.75	0.029	0.018	0.17	0.59	1.57	0.23
4	Not analysed.							
5	0.35	0.74	0.031	0.016	0.19	0.58	1.58	0.24
6	0.34	0.73	0.031	0.018	0.19	0.60	1.58	0.25
7	0.36	0.75	0.029	0.017	0.18	0.60	1.54	0.23
8	0.35	0.78	0.030	0.018	0.18	0.61	1.55	0.23
9	0.35	0.77	0.031	0.019	0.19	0.60	1.56	0.24
10	Not analysed.							
11	0.35	0.77	0.029	0.018	0.18	0.59	1.57	0.23
12	0.34	0.77	0.028	0.016	0.18	0.61	1.58	0.24
13	0.36	0.78	0.030	0.018	0.18	0.59	1.57	0.23
14	0.35	0.80	0.031	0.019	0.18	0.61	1.55	0.23
15	0.34	0.79	0.030	0.020	0.17	0.59	1.58	0.24
16	Not analysed.							
17	0.35	0.75	0.029	0.017	0.19	0.61	1.58	0.24
18	0.35	0.76	0.027	0.017	0.19	0.59	1.59	0.23

Fig -2-

First forging stage 2.92/1. General positions.
Fig No. 3.

Approx. x 1/7

Location of test sampling positions on three test slices - "A", "T" and "B" of Stage 1. forging 2.92/1.

Fig No. 4.

LOCATIONS OF TEST SAMPLING POSITIONS.

Approx. x 1/5

Stage 2. forging - 9.0/1.
Fig No. 5.

Approx x 1/2.

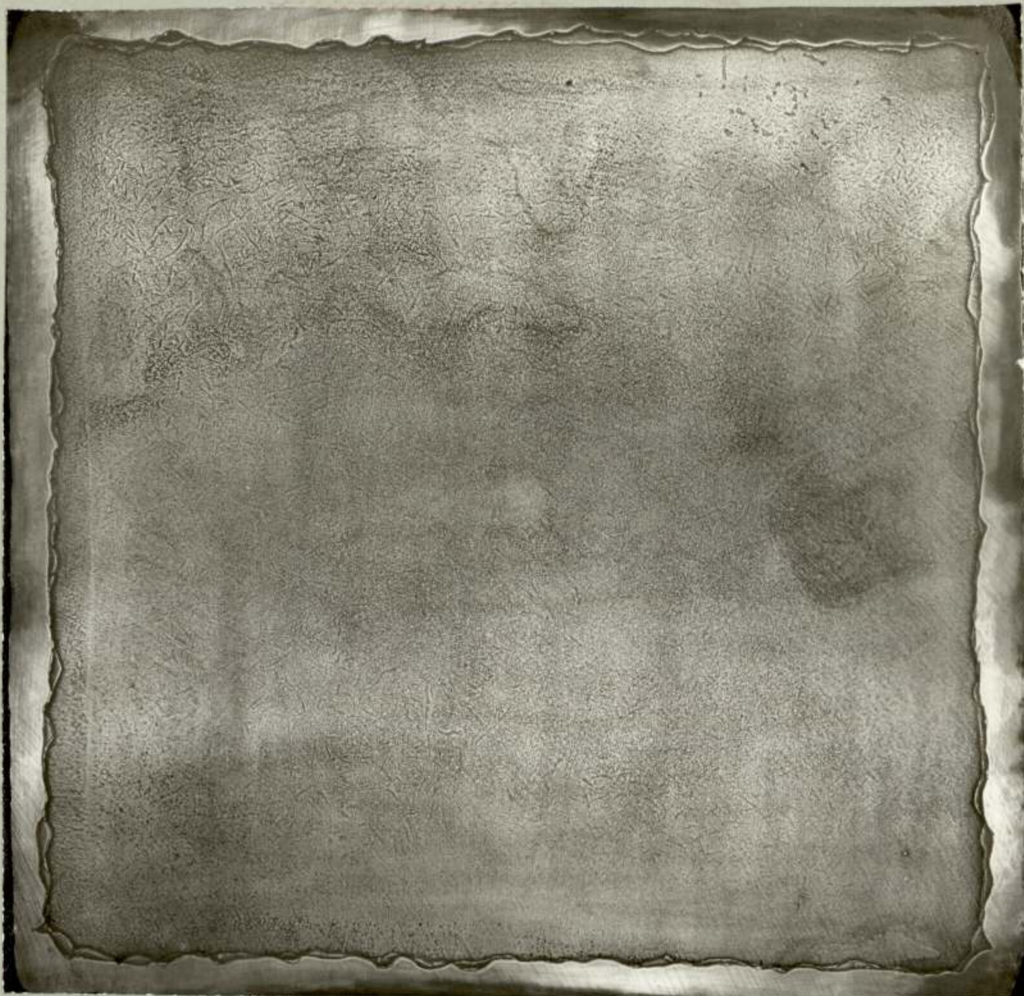
Stage 3. forging - 18.5/1.
Fig No. 6.

DIE INSERTS WITH USED SUNKEN IMPRESSIONS.

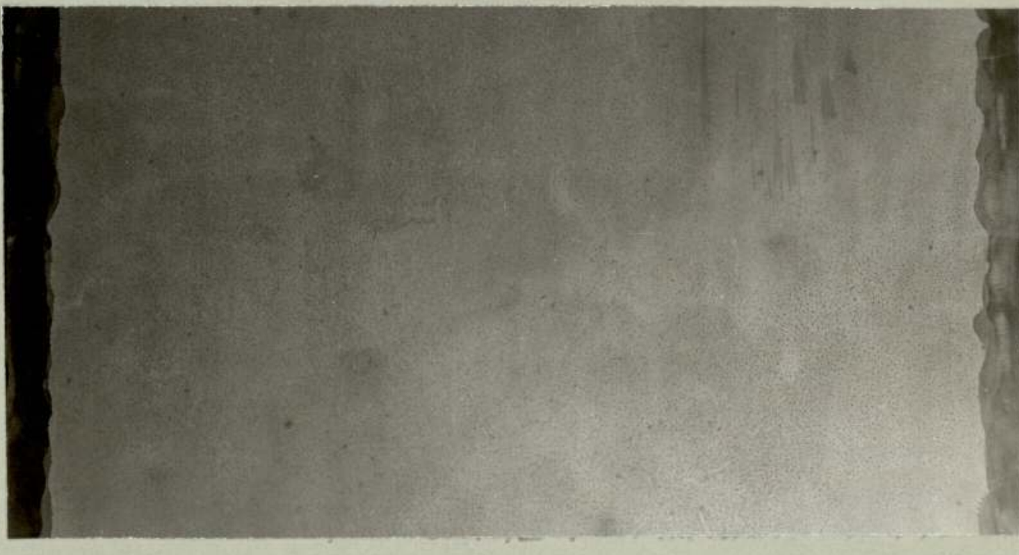
Approx. x 1/3

Fig -7-

MACROSTRUCTURES.

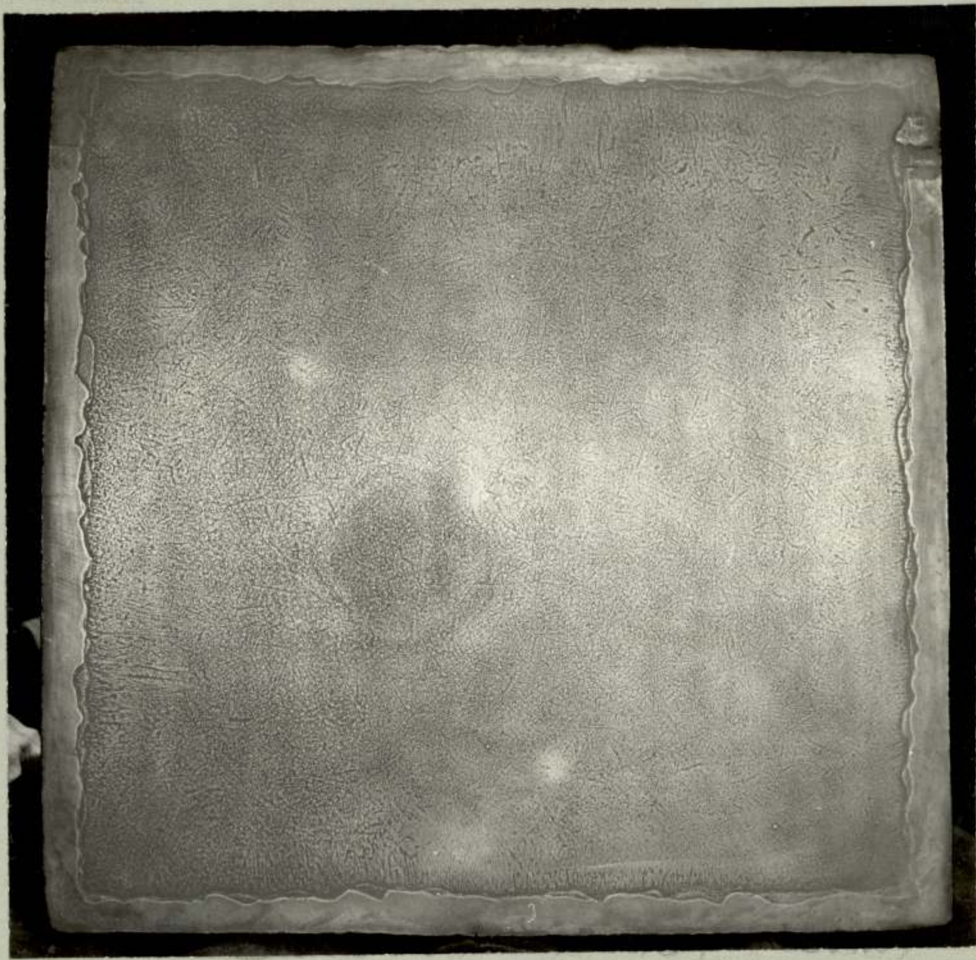


Slice "T". Stage 1. Forging 2.92/1. Approx. x 1/3



Slice "B". Stage 1. Forging 2.92/1.

MACROSTRUCTURE.

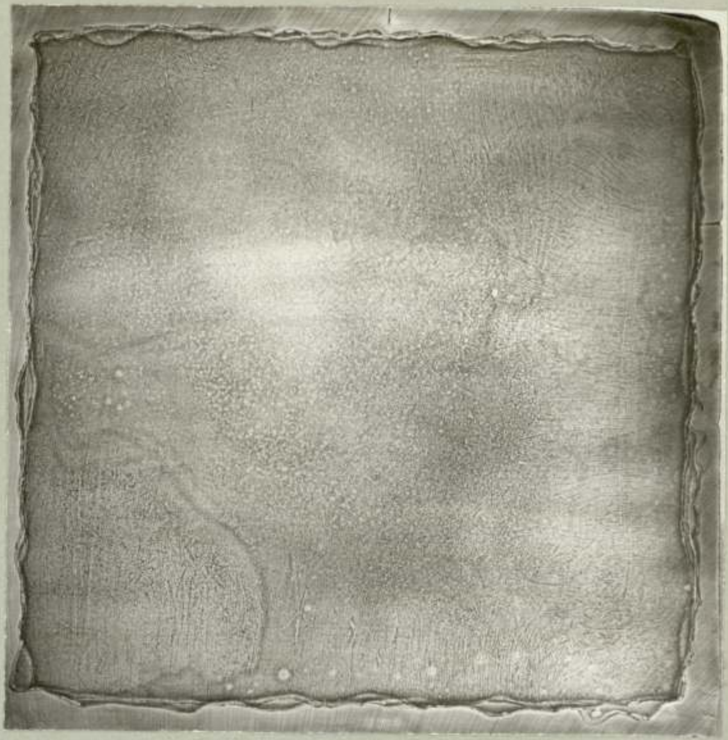


Approx. x 1/3

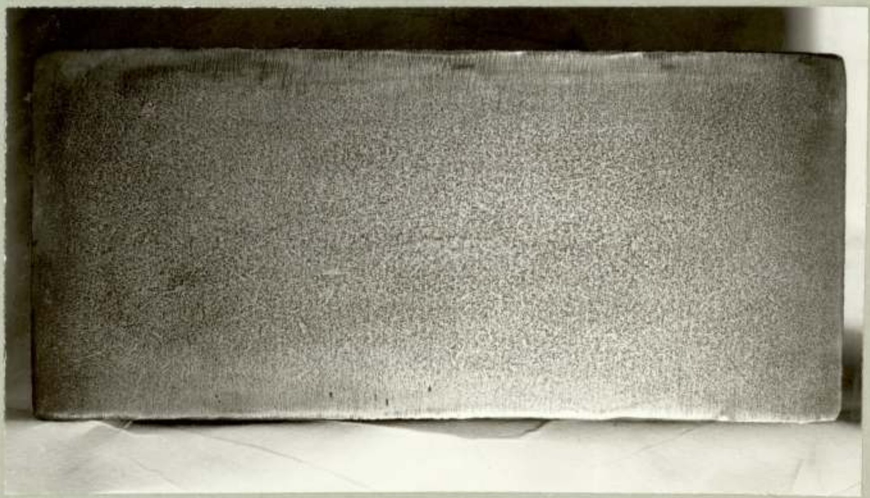
Slice "A" Stage 1. Forging 2.92/1.

Fig -9-

MACROSTRUCTURE.

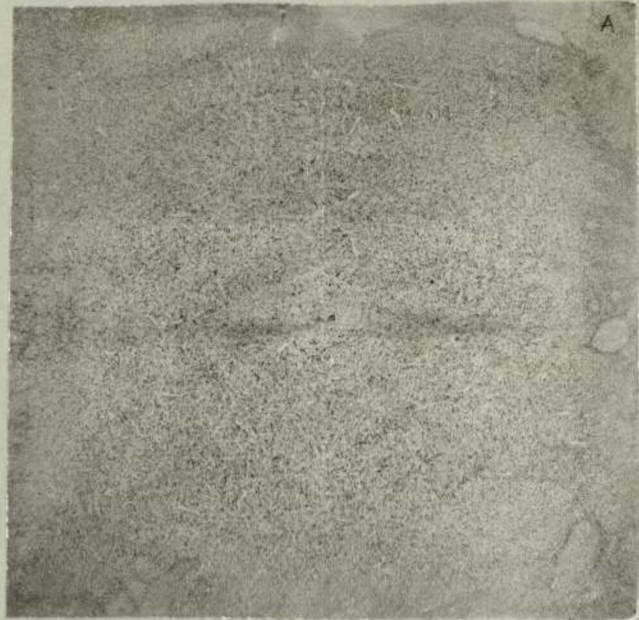


Test slice "M"   Stage 2.   Forging 9.0/1.
Approx. x 1/2.

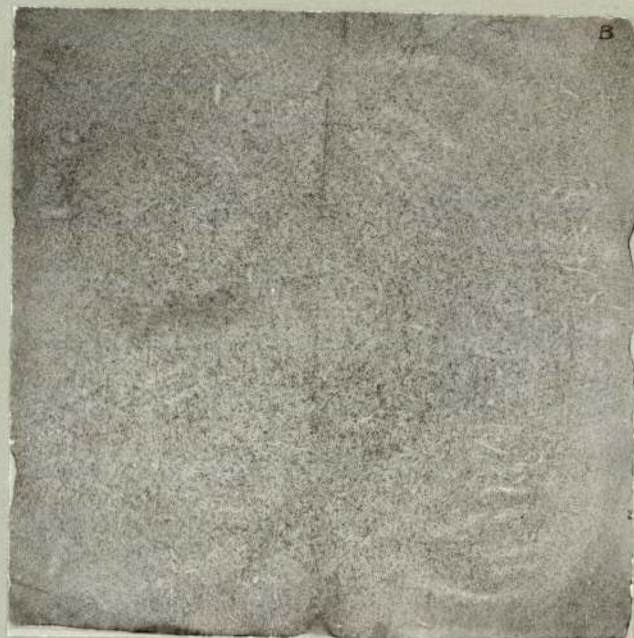


Test slice.   Stage 3.   Forging 18.5/1.
Approx. x 1/2.

Fig -10-

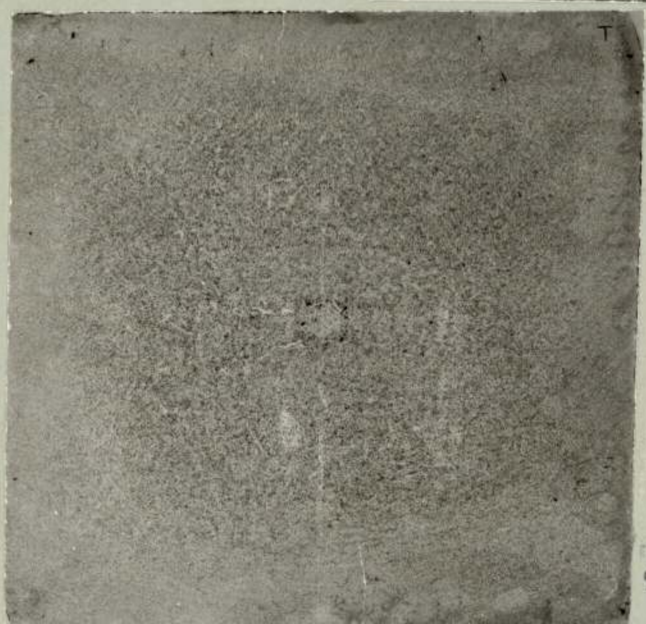


Test slice "A"



SULPHUR PRINTS.

Stage 1. forging
2.92/1.



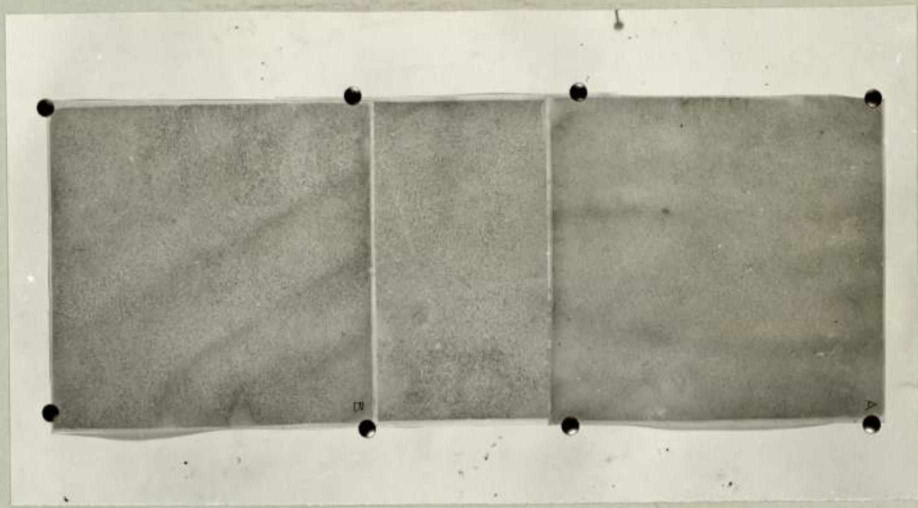
Test slice "B"

Fig -11-

Approx x 1/4.

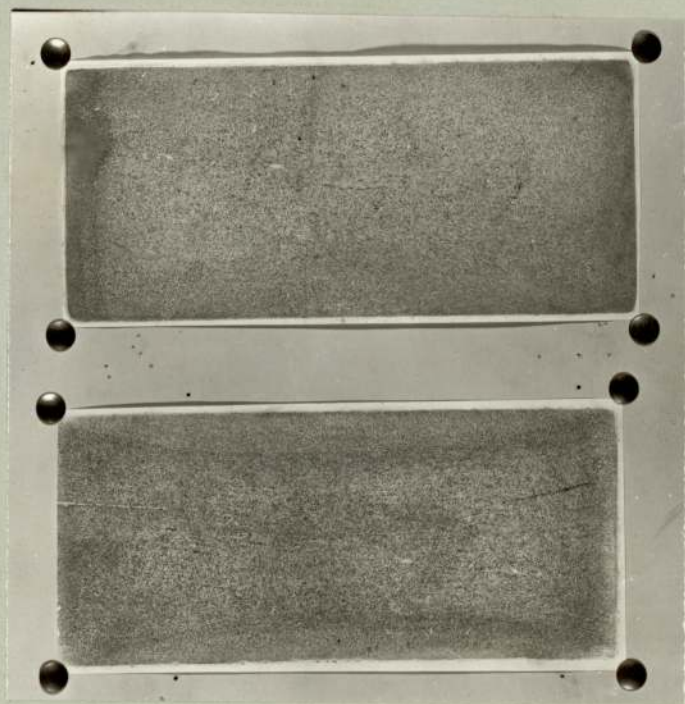
Test slice "T"

SULPHUR PRINTS.



Approx. x 1/4.

Stage 2. forging 9.0/1.
Test slices "B", "M" and "A".



Stage 3. forging. 18.5/1. Approx. x 1/3.

Fig -12-

Fig -13- DUCTILITY - COMPARISON ON TEST PIECE SIZE.

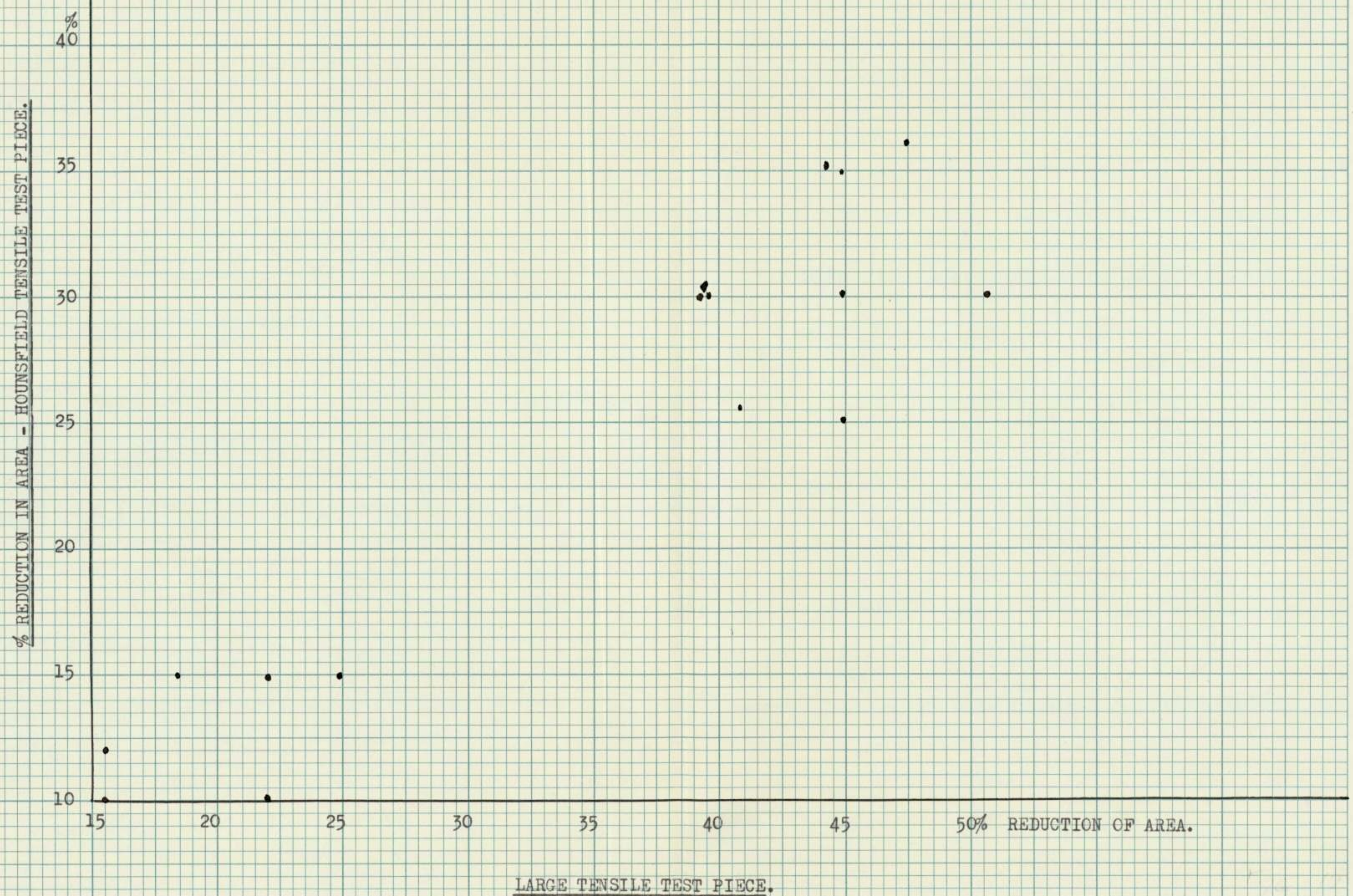


FIG -13-

Fig -14- COMPARISON OF MECHANICAL PROPERTIES.

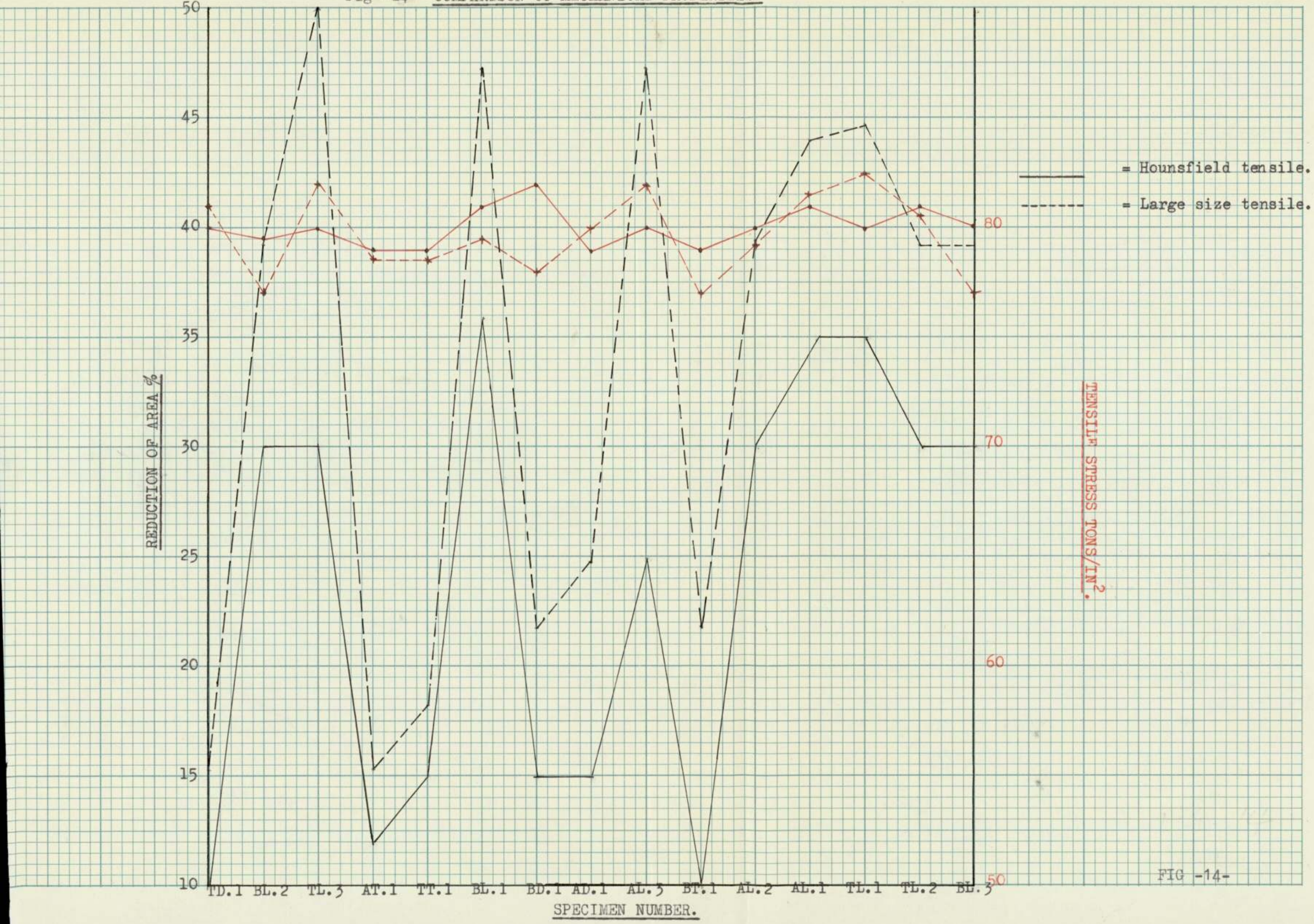


FIG -14-

Fig -15- CHARPY & IZOD NOTCH TOUGHNESS RESULTS.

Fig -16- NOTCH TOUGHNESS.

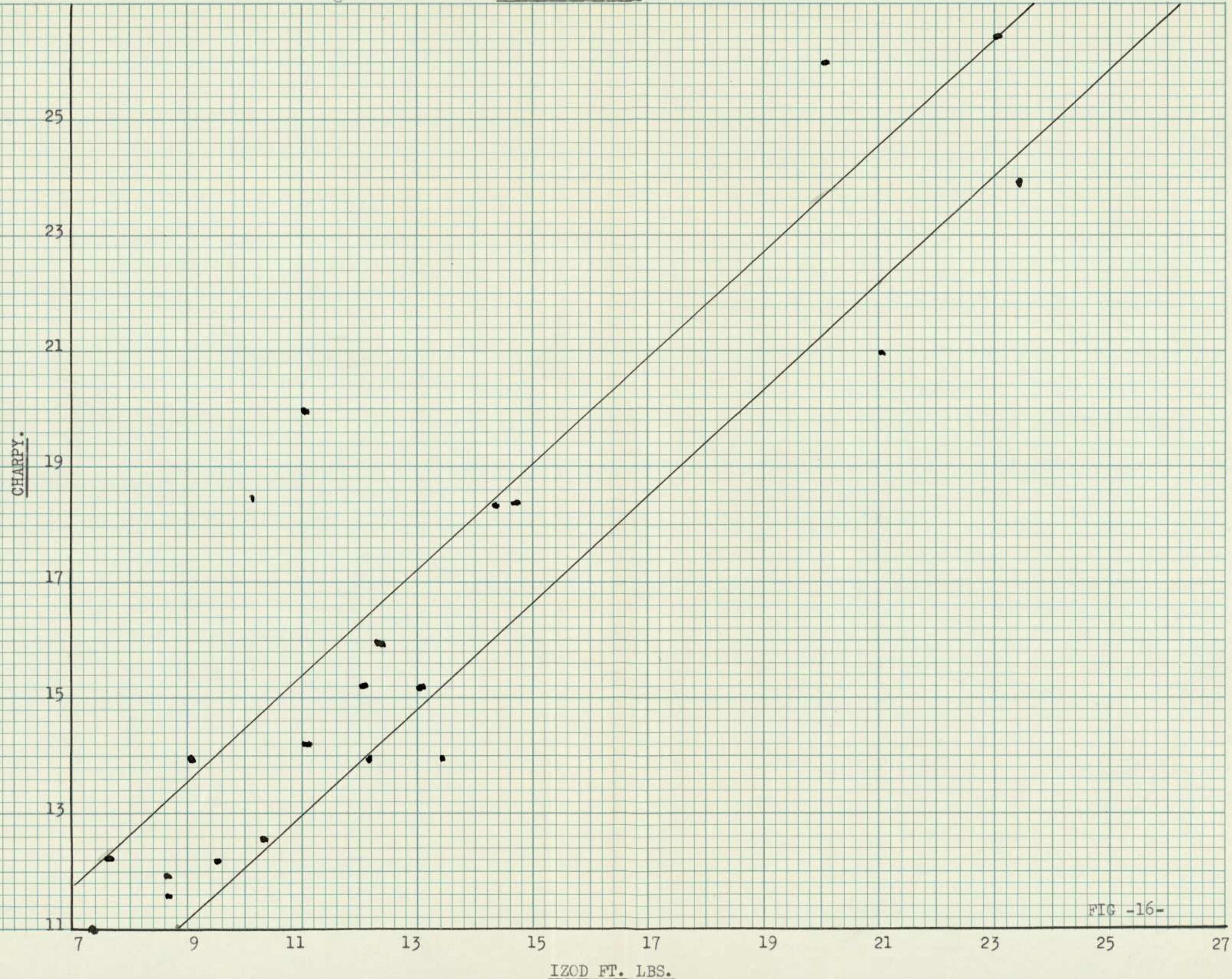


FIG -16-

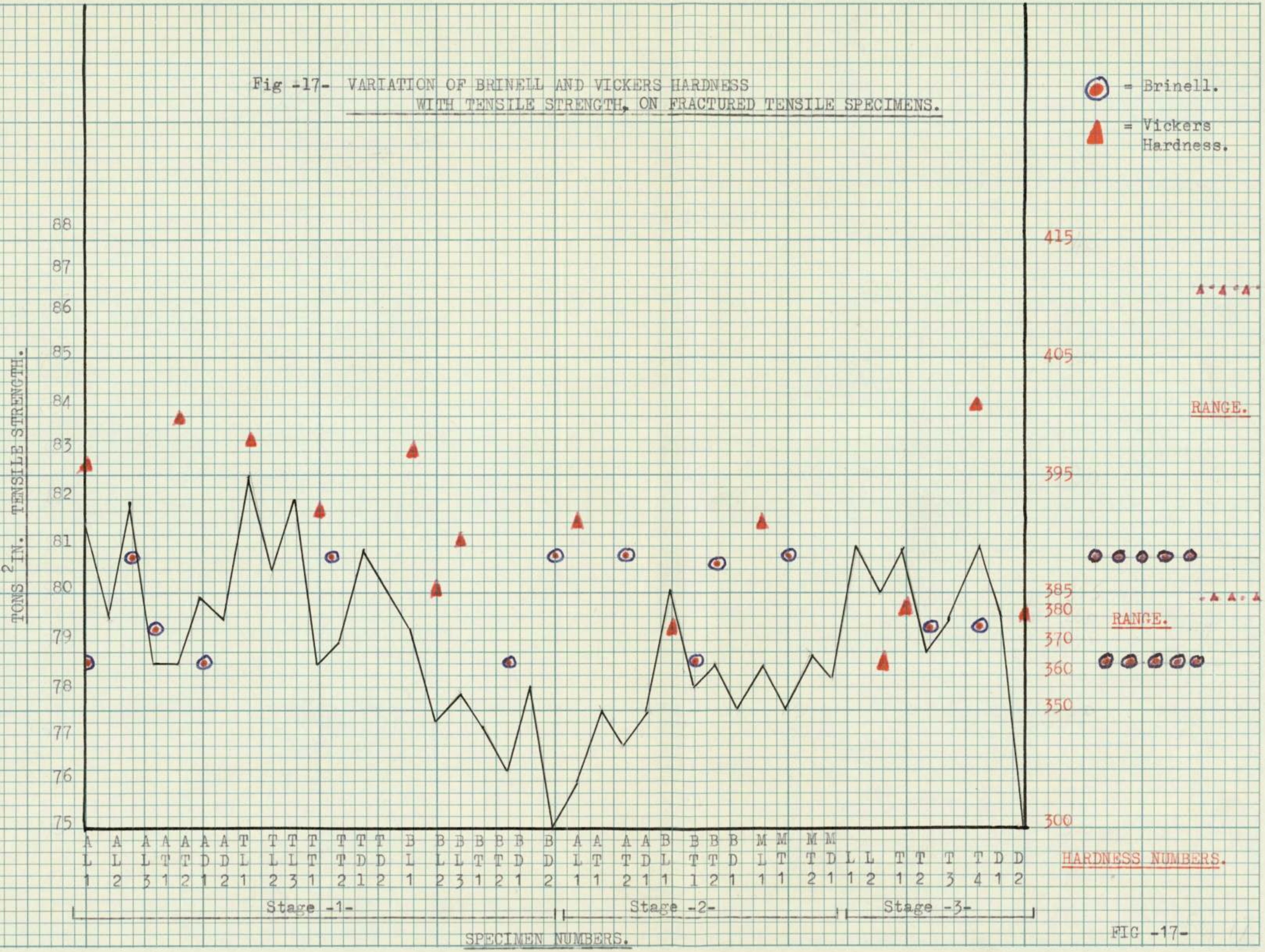
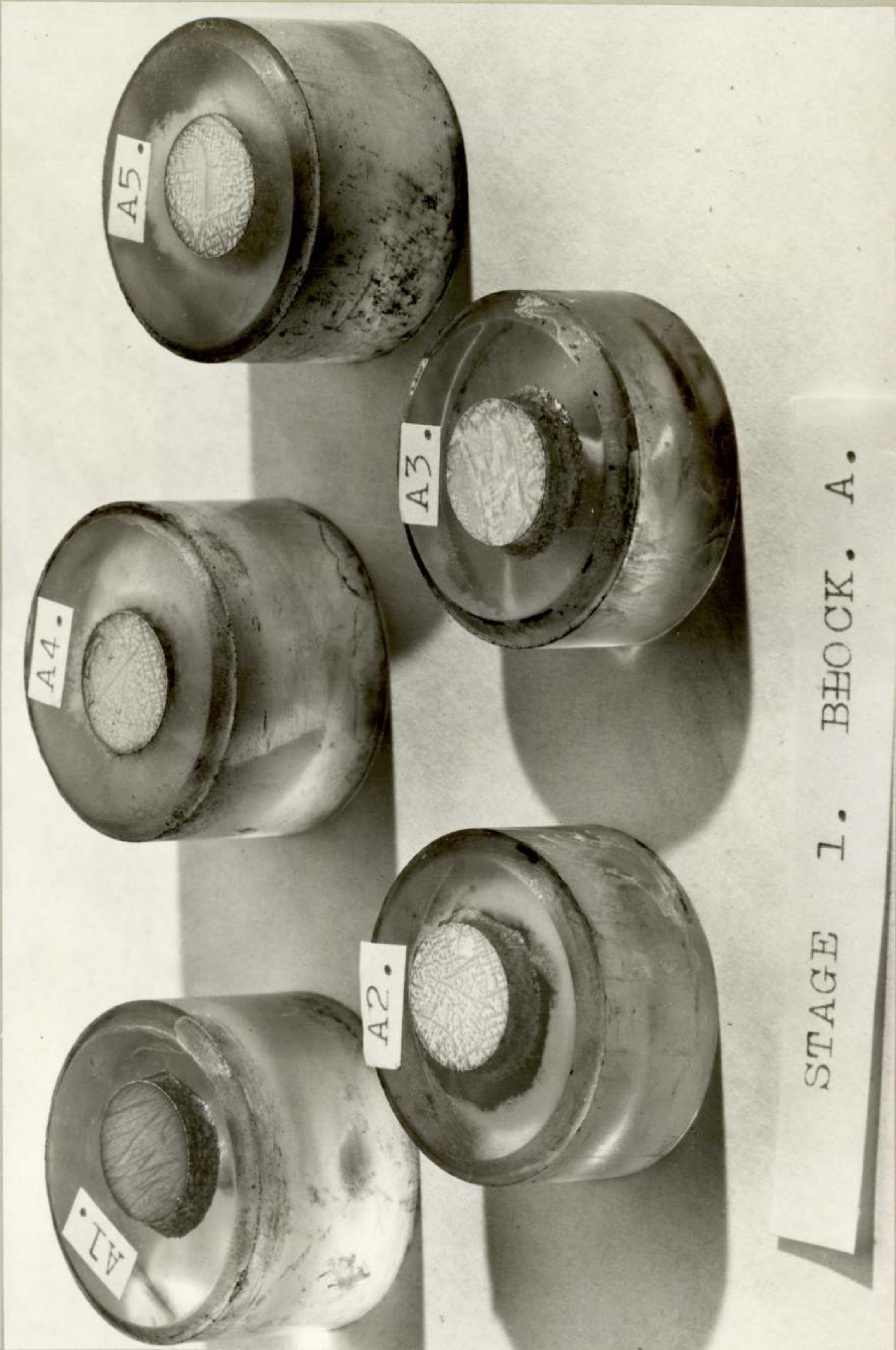


Fig -17- VARIATION OF BRINELL AND VICKERS HARDNESS WITH TENSILE STRENGTH, ON FRACTURED TENSILE SPECIMENS.
= Brinell.
= Vickers Hardness.
TONS 2 IN. TENSILE STRENGTH.
88
87
86
85
84
83
82
81
80
79
78
77
76
75
415
405
395
385
380
370
360
350
300
RANGE.
RANGE.
HARDNESS NUMBERS.
Stage -1-
Stage -2-
Stage -3-
SPECIMEN NUMBERS.

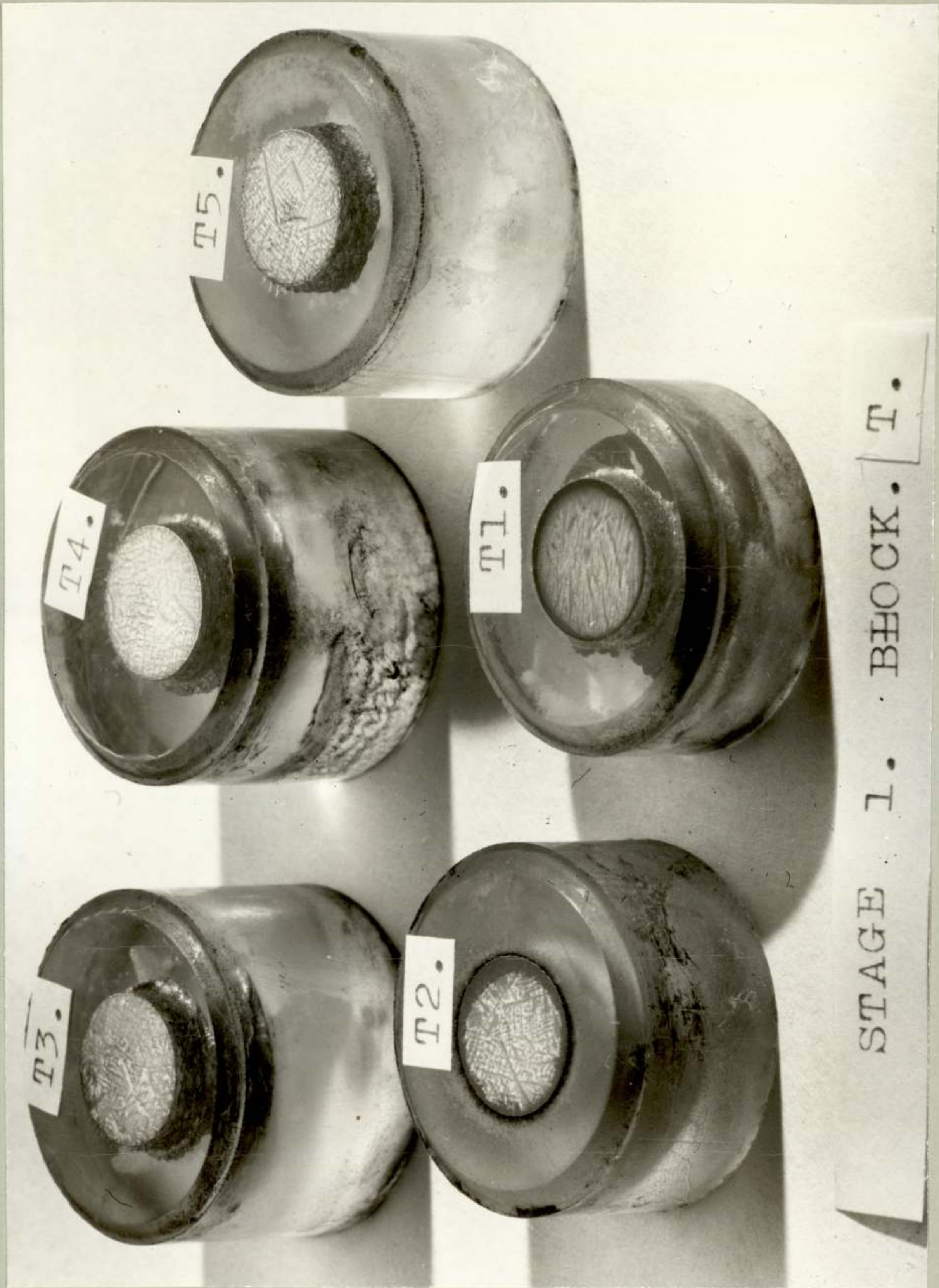
FIG -17-

Stage 1. forging 2.92/1.
Fig No.

x 2

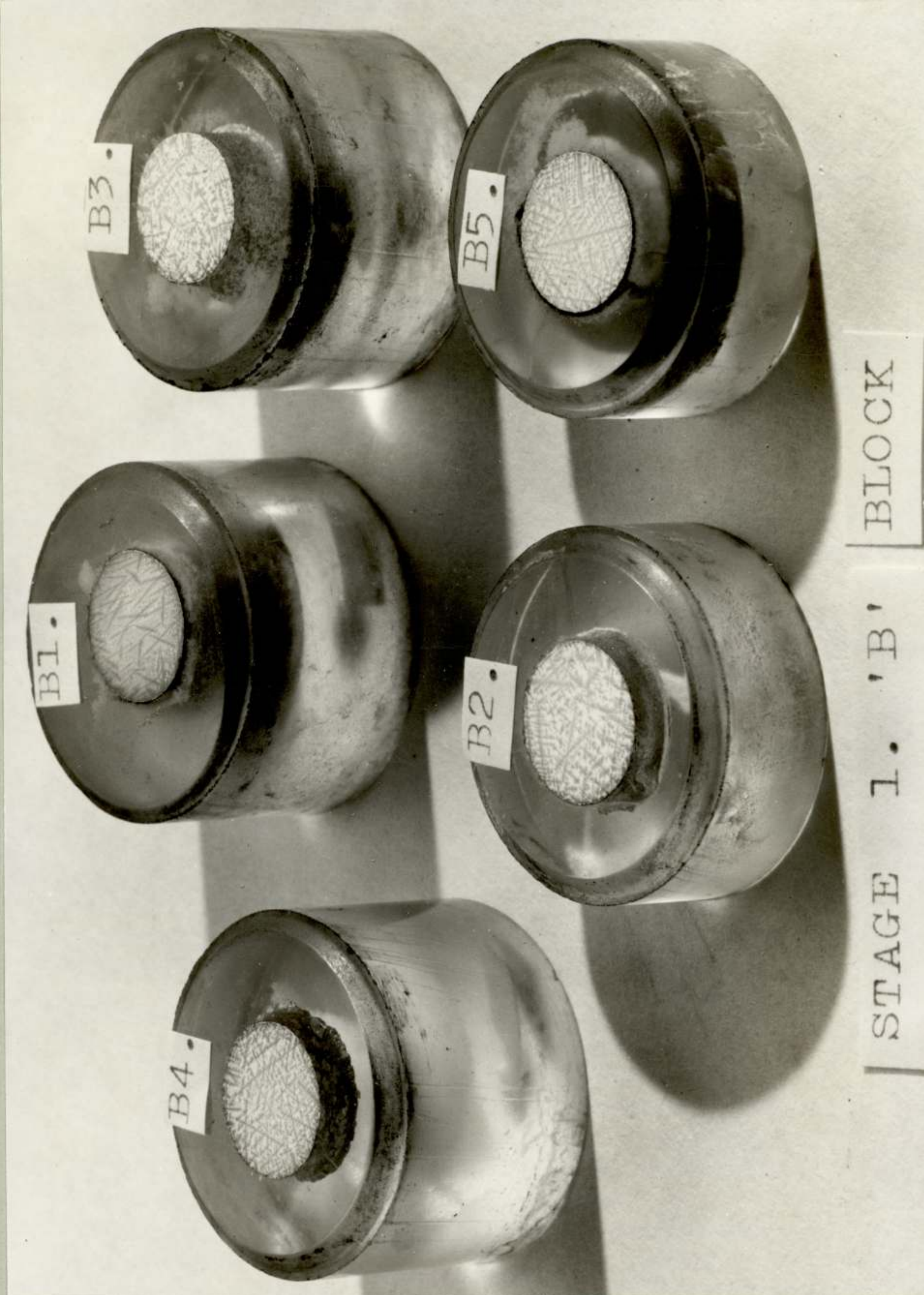


Stage 1. forging 2.92/1.
Fig -18-



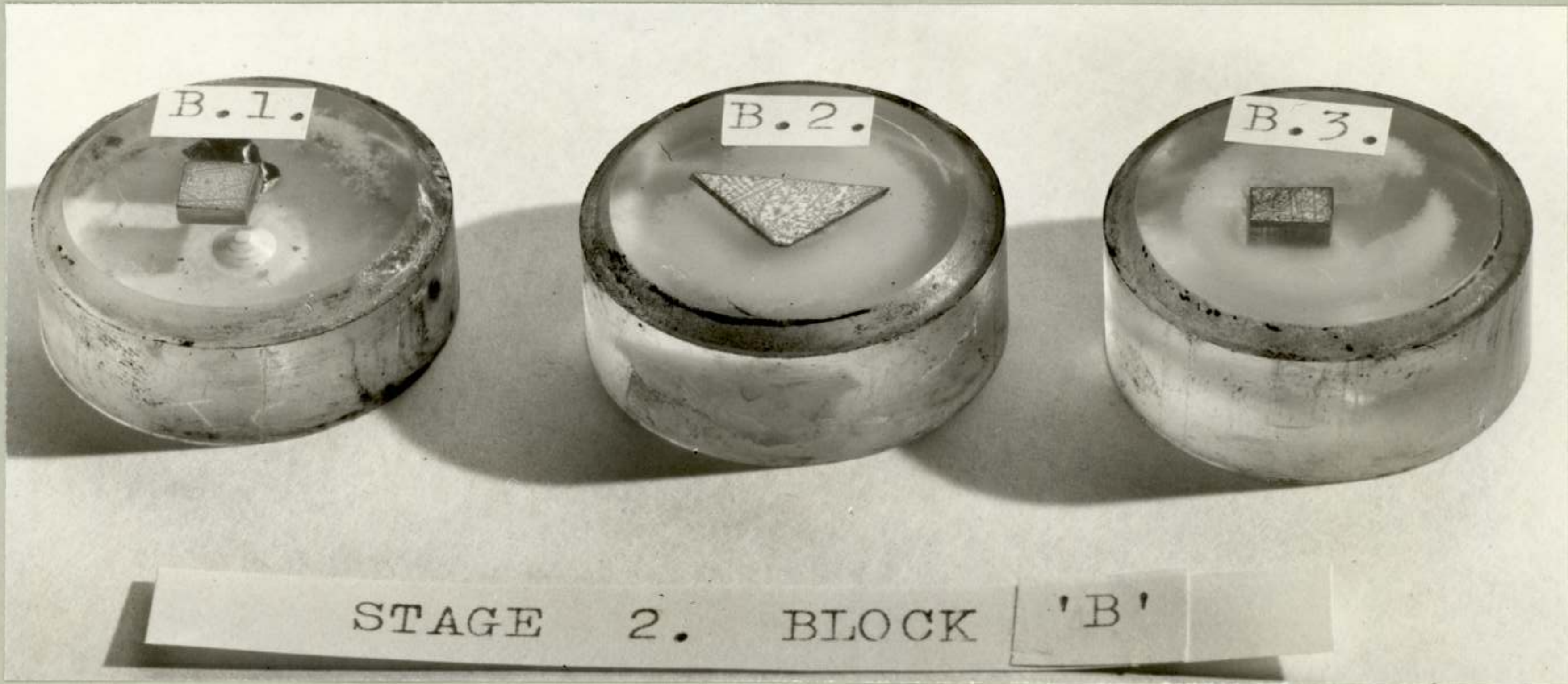
x 2.

Stage 1. forging 2.92/1.
Fig No. 19.



x 2.

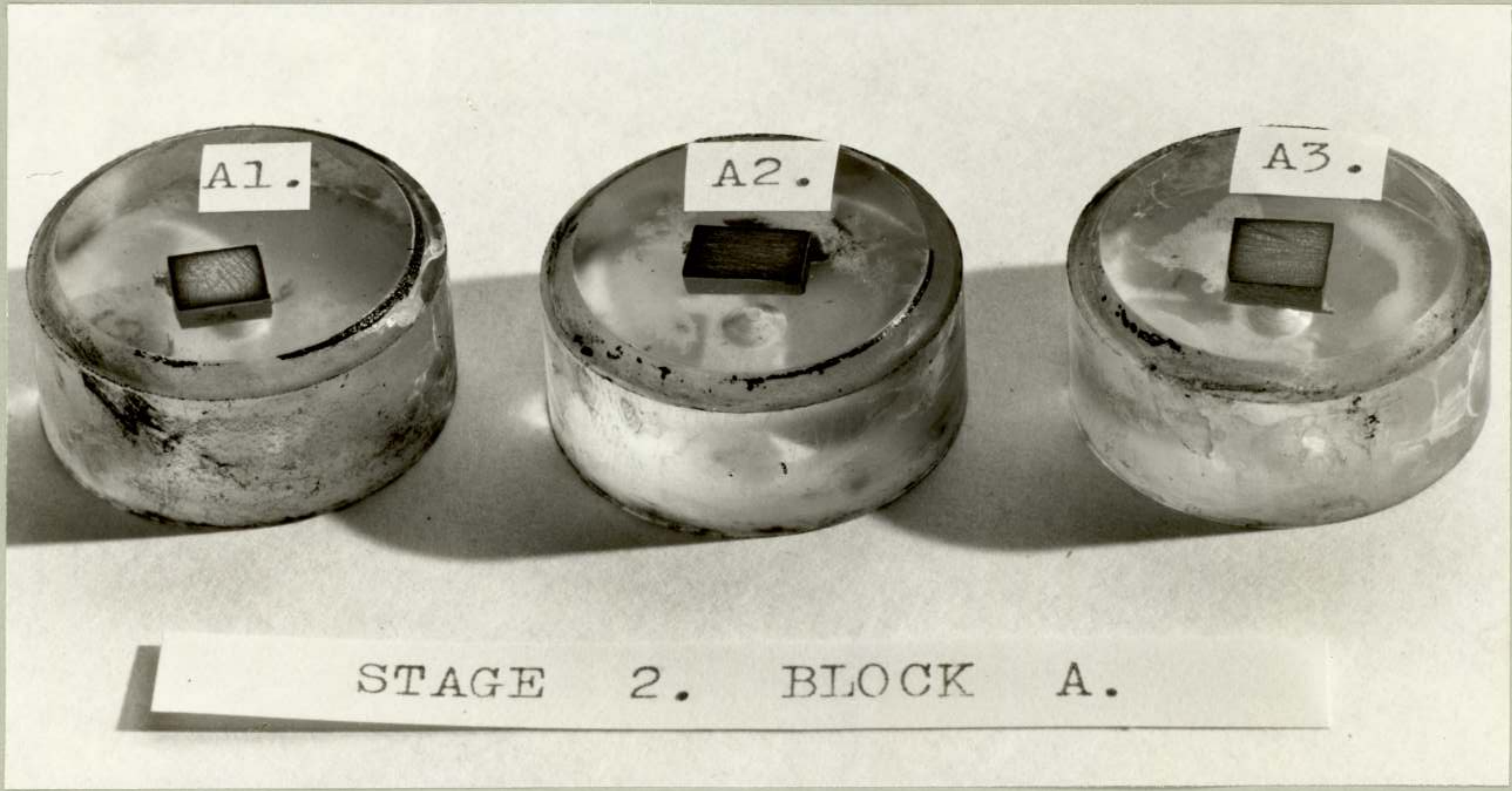
Stage 1. forging 2.92/1.
Fig No. 20.



x 2.

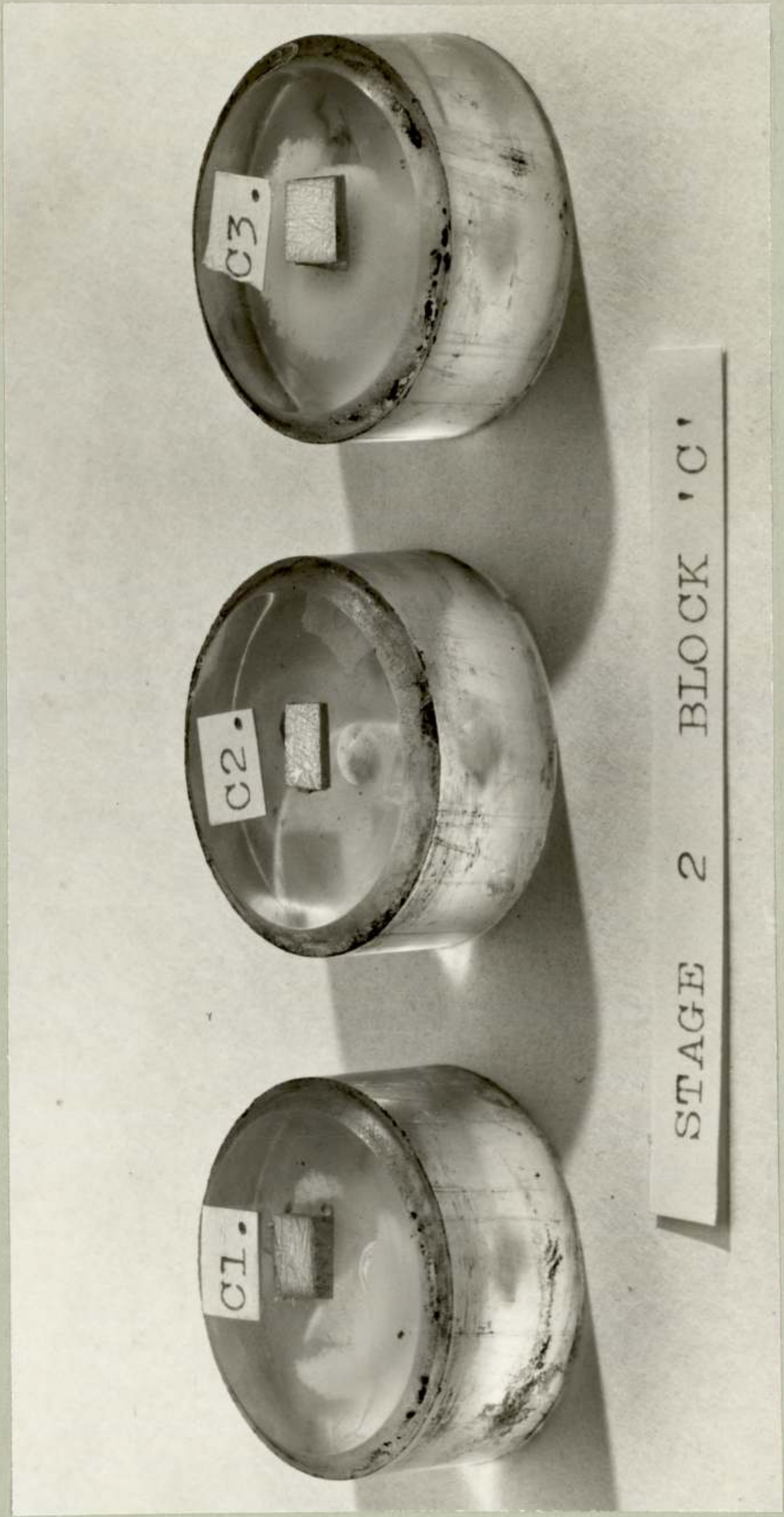
Stage 2. forging 9.0/1.

<u>Fig No. -21-</u>



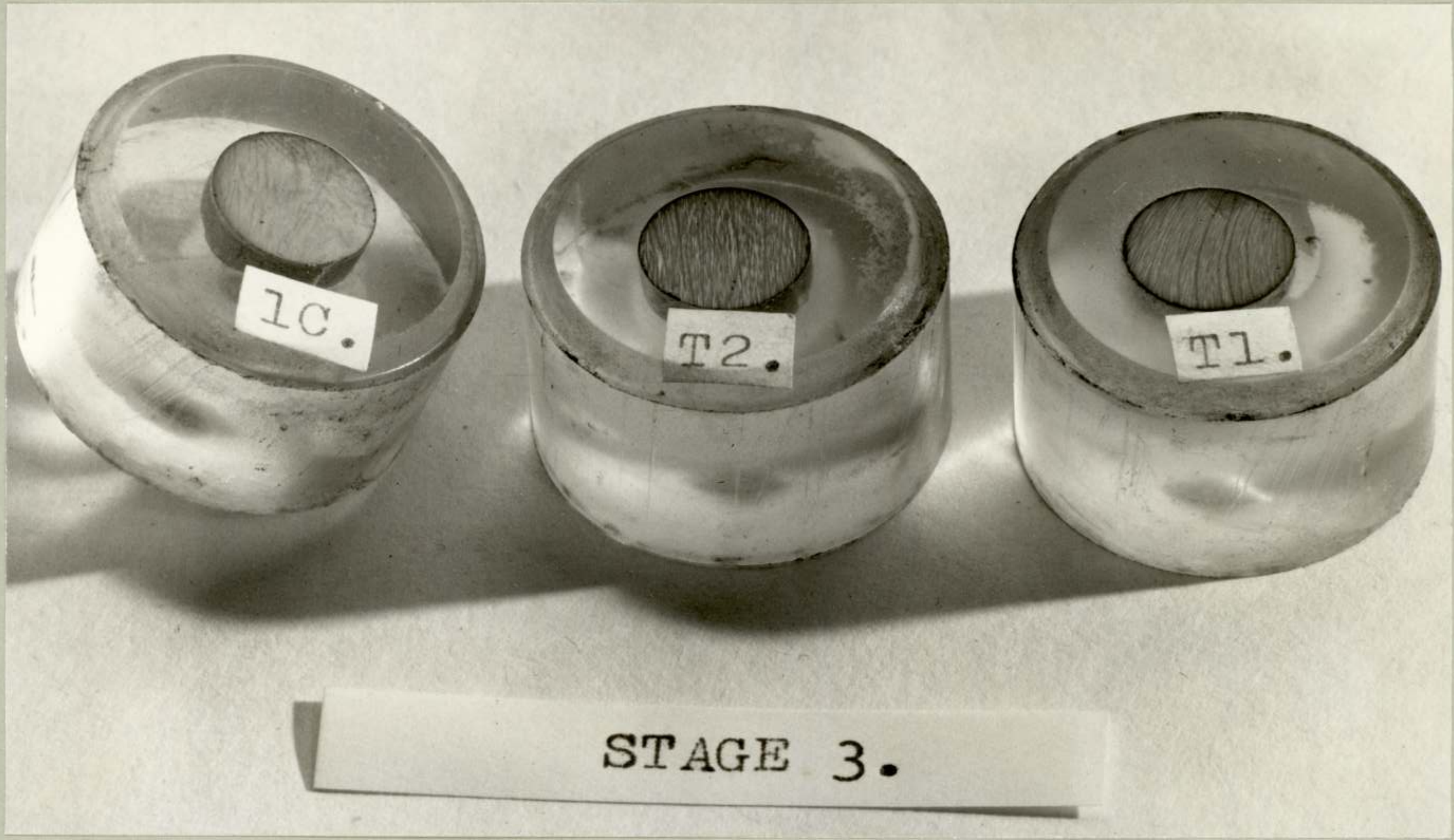
x 2.

Stage 2. forging 9.0/1.
Fig No. -22-



Stage 2
forging 9.0/1.
Fig No. -23-

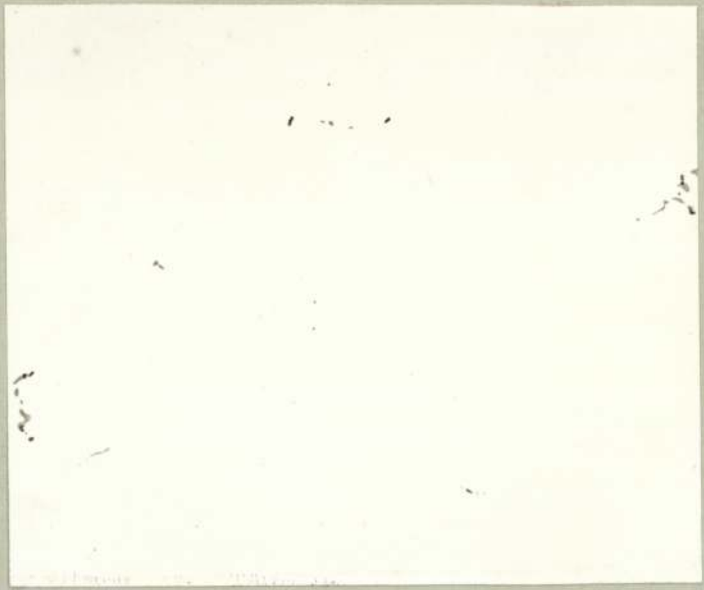
x 2.



x 2.

Stage 3 forging 18.5/1.
Fig No. -24-

Non-metallic inclusions, unetched. x 100.



Specimen A4. Stage 1.

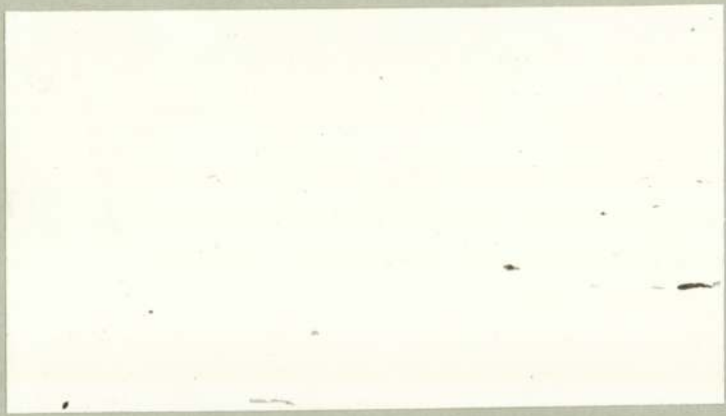
Specimen T4. Stage 1.

Specimen Al. Stage 2.

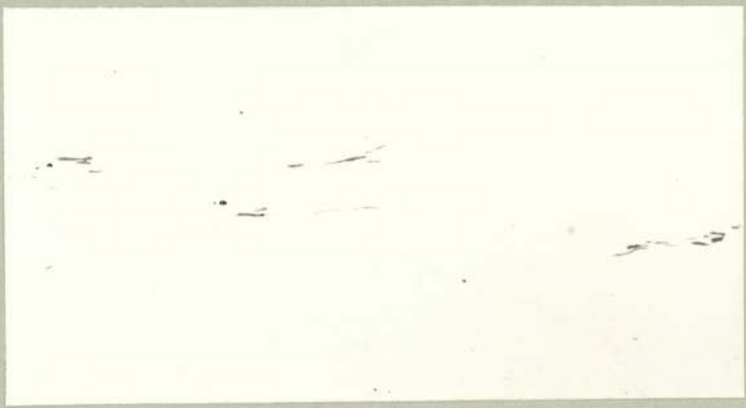
Fig -25-

Non-metallic inclusions, unetched. x 100.

Specimen C2.
Stage 2.



Specimen Cl.
Stage 3.



Specimen T2.
Stage 3.

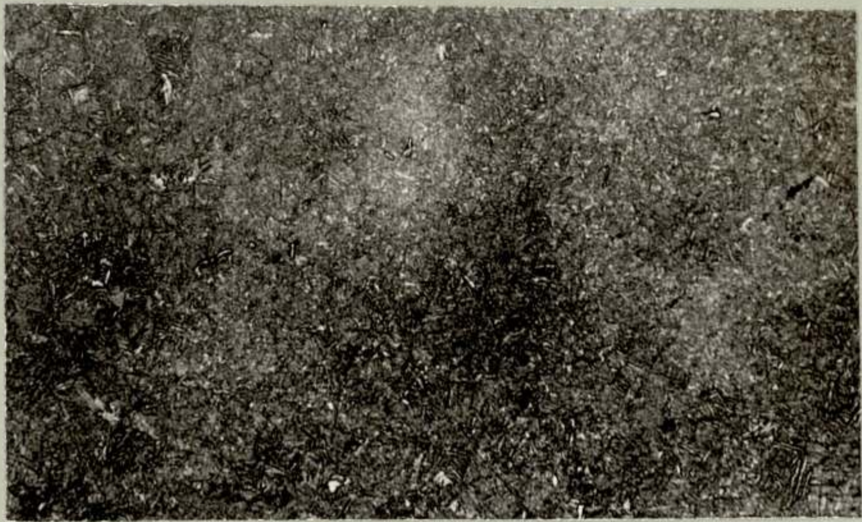
Specimen Tl.
Stage 3.

Fig -26-

General structure. x 100.

Specimen T4
Stage 1.

Fig -27-



Specimen A4
Stage 1.

Fig -28-

Specimen A1
Stage 2.

Fig -29-

General structure. x 100.

Specimen C.2.
Stage 2.

Fig -30-

Specimen T.1.
Stage 3.

Fig -31-

Specimen T.2.
Stage 3.

Fig -32-

General structure.
x 650

Specimen T.4.
Stage 1.

Fig -33-

Specimen A.4.
Stage 1.

Fig -34-

Specimen A.1.
Stage 2.

Fig -35-

General structure
x 650

Fig -36-

Specimen C2.
Stage 2.

Fig -37-

Specimen Cl.
Stage 3.

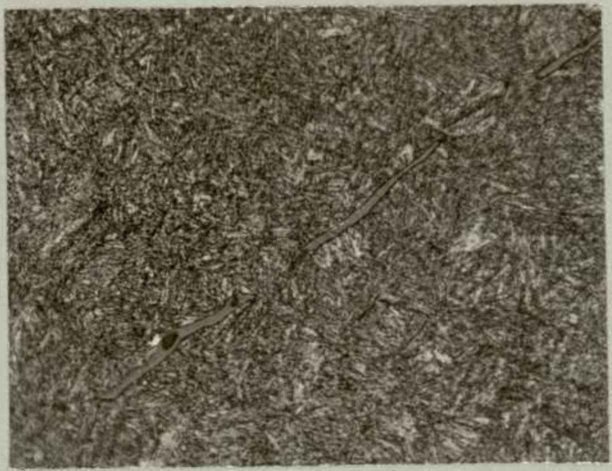


Fig -38-

Specimen Tl.
Stage 3.

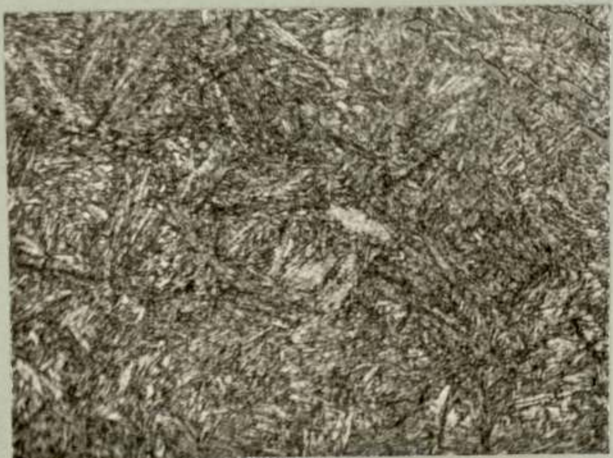


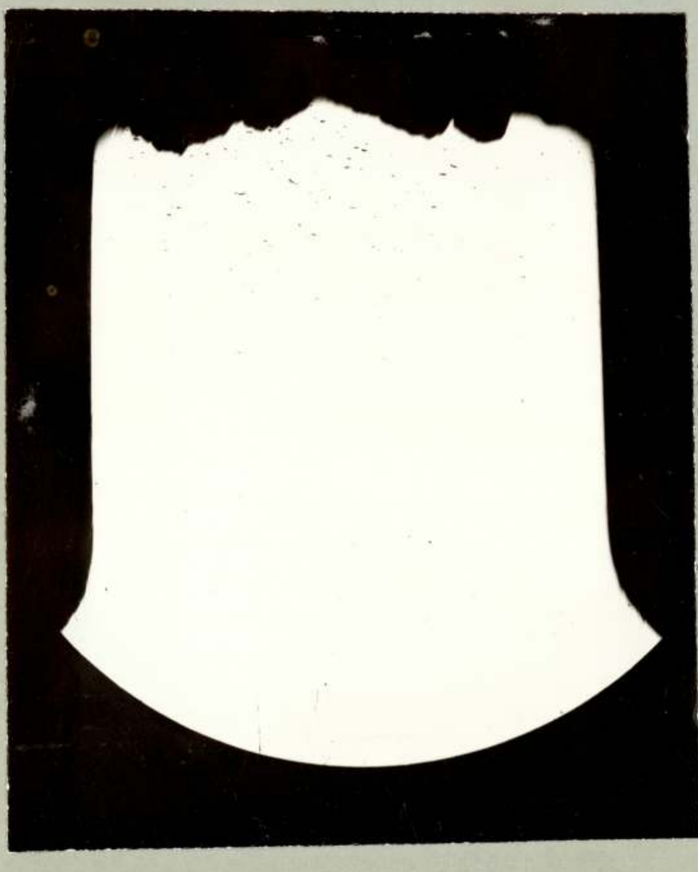
Fig -39-

Specimen T2.
Stage 3.

Unetched half tensile samples x 10

Longitudinal testing direction.

Stage 3 forging 18.5/1



Depth testing direction.

Unetched half tensile samples x 10

Transverse testing direction.

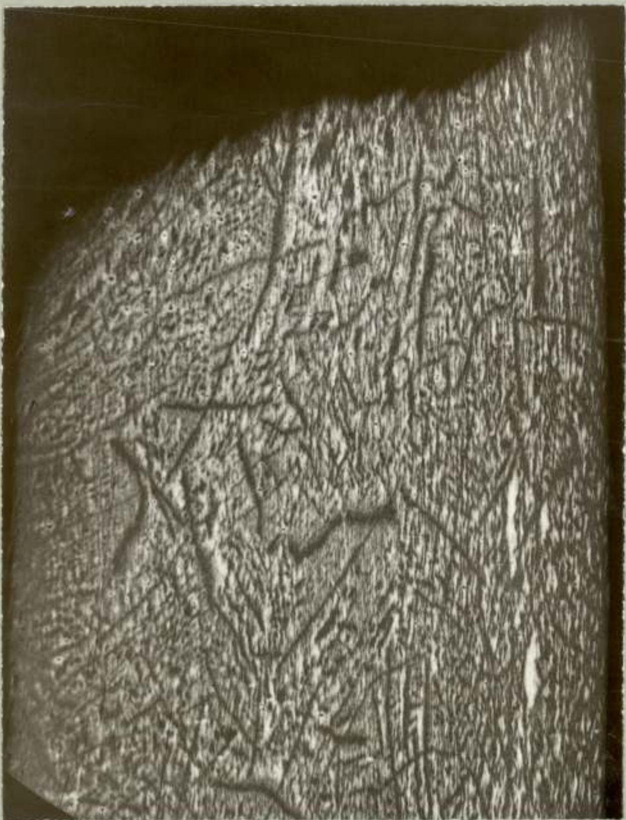
Stage 3 forging 18.5/1.

Transverse testing direction.

Etched half tensile samples x 10.

Transverse testing direction.

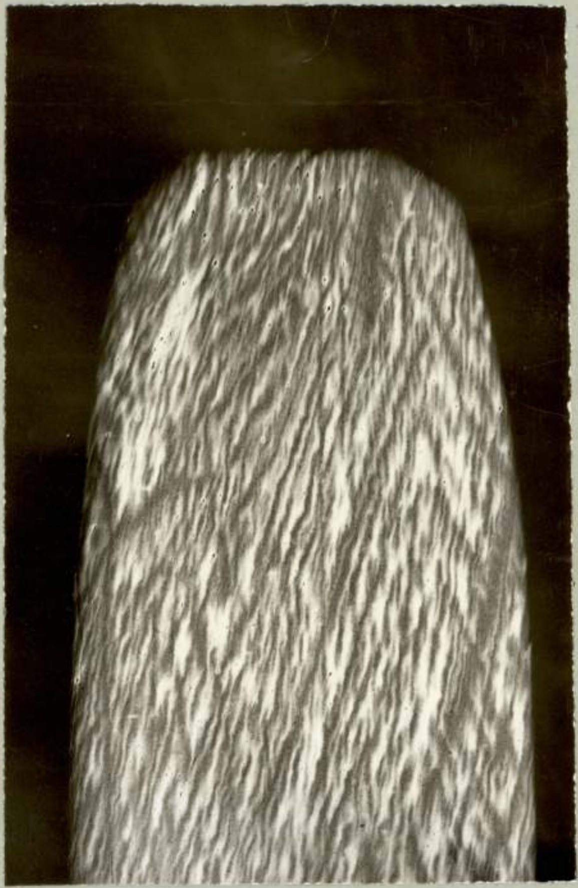
Stage 3. forging 18.5/1.



Transverse testing direction.

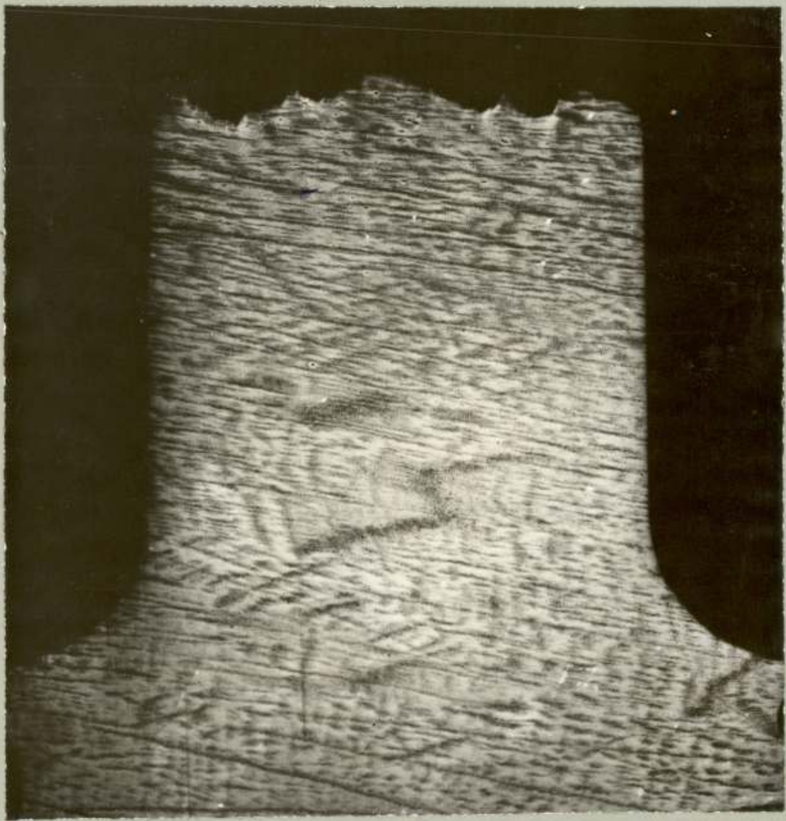
Fig -42-

Etched half tensile samples x 10



Longitudinal testing direction.

Stage 3 forging 18.5/1.



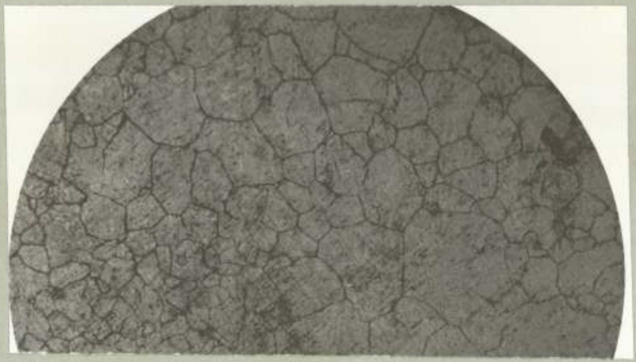
Depth testing direction.

Fig -43-

Grain size
determinations
x 100.

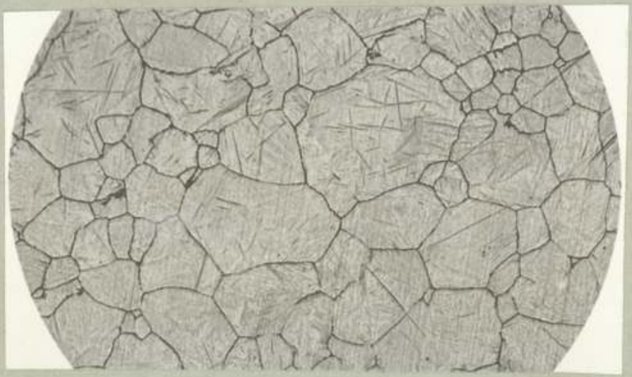
TL3.

AD.1



AD2.

AT.2.



AT1.

Fig -44-

Fig -45- VICKERS HARDNESS TESTING.

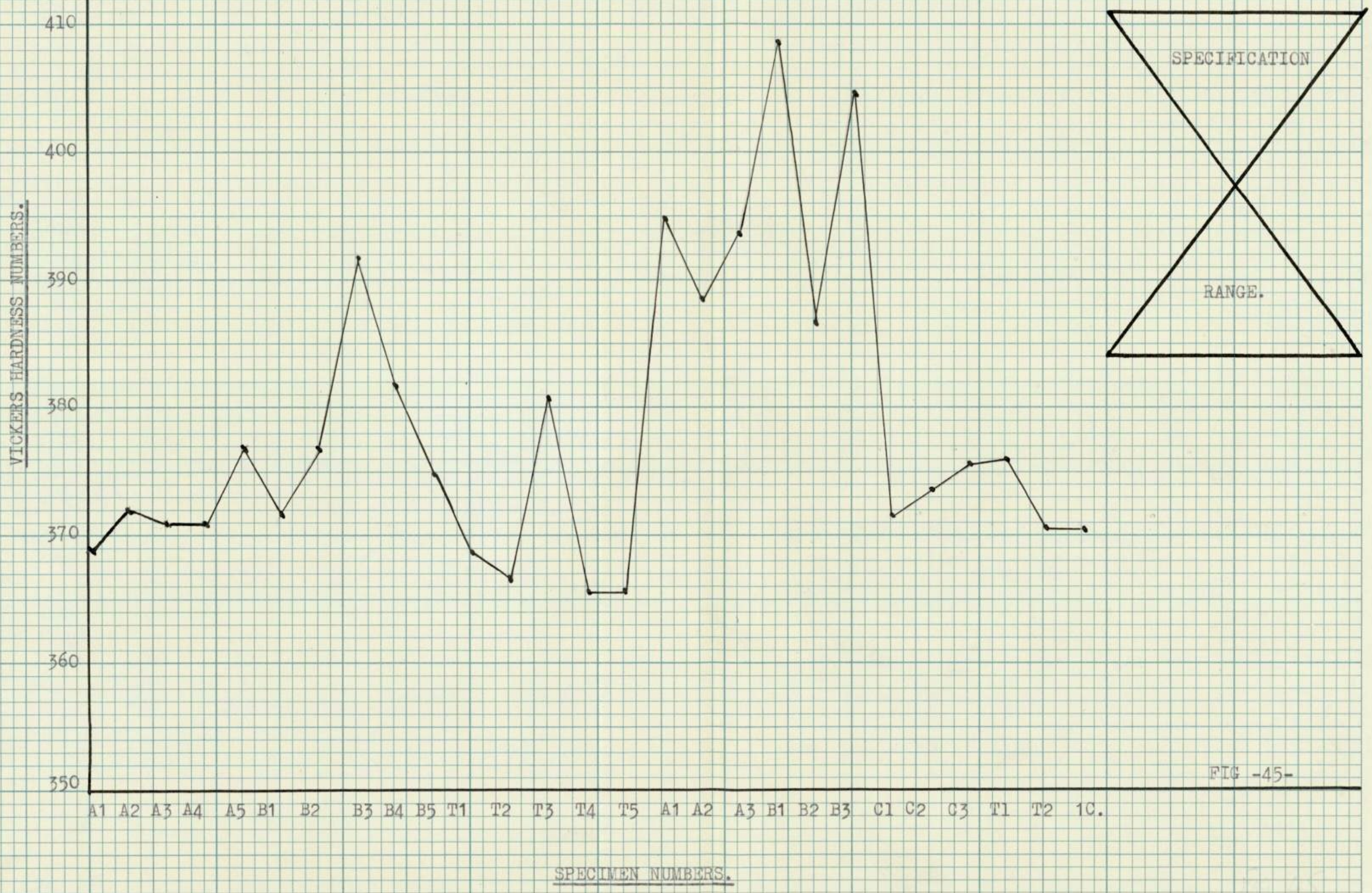


FIG -45-

SHOWING X-RAY PICTURES AT
MAG. x 250 ACROSS DENDRITE
AS SEEN UNDER OPTICAL
MICROSCOPE INDICATED BY SCRATCH.

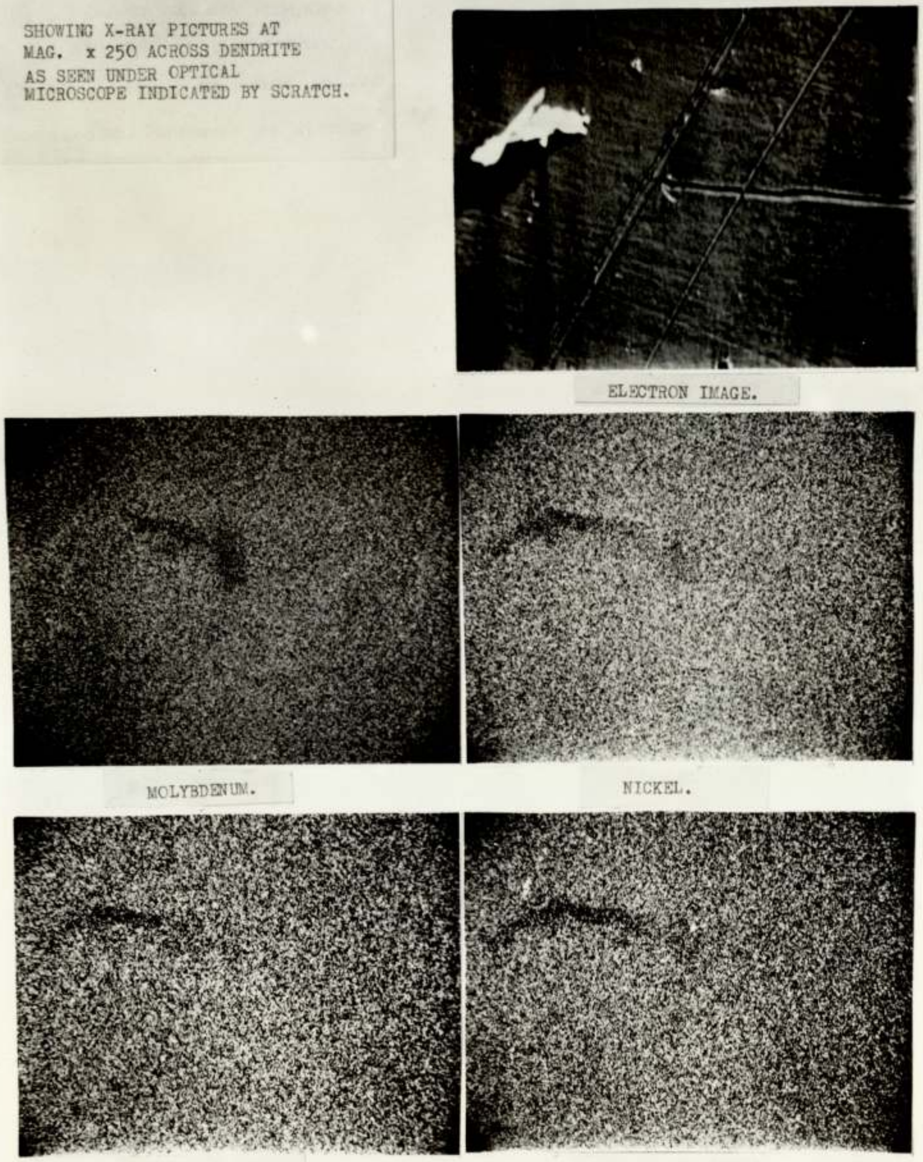
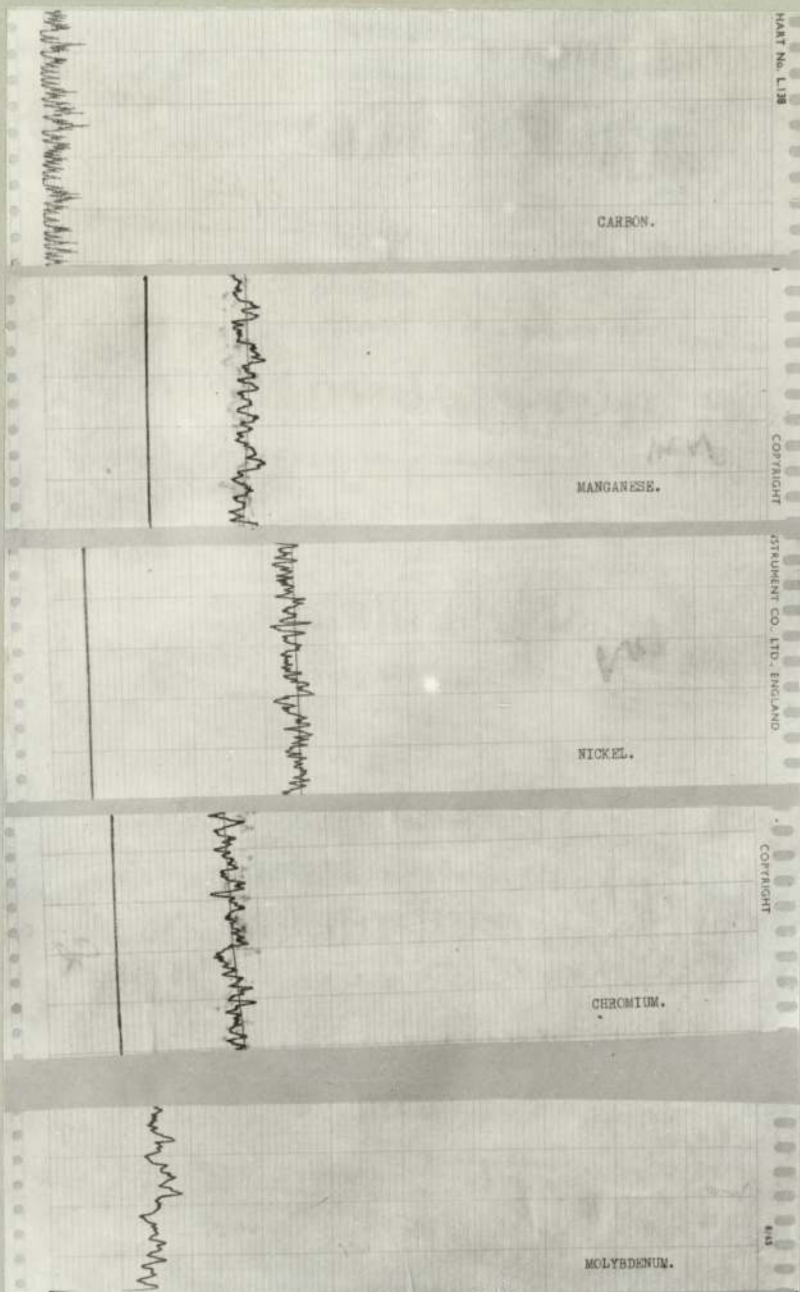


Fig -46-

MICRO-PROBE SCANNING TRACES.



Approx. x 1/3.

-Fig -47-