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ROLLING RECRYSTALLISATION TEXTURE OF COMMERCIAL GRADES OF LOW CARBON STEELS

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SUMMARY

The work in progress concerning the study of rolling-recrystallisation textures in commercial grades of low-carbon steels is outlined, the influence of aluminium nitride precipitation and recovery phoenomena being the main lines of research. Results to date indicate that the retention of the {100}<110> component of the rolling texture is more favoured during recrystallisation as the prior recovery times and temperatures are increased. Textural variations through the sheet thickness are also being studied, results to date being presented.

Contents

I. Introduction

The importance of anisotropy, caused by crystallographic preferred orientation, upon the forming characteristics of commercially produced, annealed-low-carbon steel sheet has long been recognised[1] such that interest has been shown in the factors thought to influence texture development. Annealing textures arise during the recrystallisation and grain growth of materials exhibiting marked deformation textures, consequently many workers have studied the development of rolling textures in b.c.c. metals[2, 3, 4, 5, 6]. Such work has recently been reviewed by Dillamore and Roberts[7]. Most workers agree that the $\{100\}\langle 110\rangle$ orientation spread about the rolling direction constitutes the main feature of the texture $\{112\}\langle 110\rangle$ frequently being highly developed. Other components have been variously described as spreads about $\{111\}\langle 110\rangle$ and $\{111\}\langle 112\rangle$. Haesner and Weik[3] described the minor components as resulting from two fibre textures, one being considered as a $\langle 110\rangle$ fibre axis rotated 35° from the sheet normal towards the rolling direction and the other as a limited fibre texture of increasing spread around the rolling direction as axis.

The use of ideal orientations to describe texture is convenient, but can be misleading as large orientation spreads are frequently encountered. Close examination of the pole figures of the different workers is essential when comparing results.

Two theories of rolling texture development give answers which compare reasonably well with experimental results, namely those of Calnan and Clews[8], and Dillamore and Roberts[9].

Calnan and Clews[8] suggest that in order to maintain grain boundary continuity during deformation at least four slip systems must operate simultaneously in any given grain, boundary constraints moving the effective stress axis to one of the corners of the unit triangle. They suggest that orientations which simultaneously satisfy the requirements of a tensile stress in the rolling direction and a compressive stress parallel to the sheet normal will constitute major components in the rolling texture. Rolling textures were thus suggested from predicted tensile and compression textures. Three types of slip system were considered $\{110\}\langle 111\rangle$ $\{112\}\langle 111\rangle$ $\{123\}\langle 111\rangle$, giving a total of 48. Each of the three types were considered individually and the predictions obtained from each did not vary significantly. The predicted tension texture was $\langle 110\rangle$ spread towards $\langle 311\rangle$ whilst the compression texture was suggested to have a strong $\langle 111\rangle$ component plus a weaker one corresponding to $\langle 100\rangle$. The predicted rolling texture has $\{100\}\langle 110\rangle$ and $\{111\}\langle 110\rangle$ as major components.

A major flaw in this theory is that the stable texture is assumed to be a combination of those obtained for tension and compression, no account being taken as to how the end points are arrived at. Dillamore and Roberts[7] have pointed out that all the possible starting grain orientations cannot, during rotations to form a tension texture, simultaneously satisfy the requirement of an orthogonal compressive stress; in fact only 20% can do so. They suggest that a better approximation to the rolling process is to assume a biaxial stress system and then to determine the slip systems that would operate under combined stresses. In reality the stress system is tri-axial but since the process approximates to one of plane strain it is suggested that the state of yielding is equivalent to that obtained by considering a tensile and compressive stress equal in magnitude parallel to the sheet normal and rolling direction respectively.

Unlike Calnan and Clews[8], Dillamore and Roberts[9] suggest that multiple slip need only be invoked in close proximity to grain boundaries, the main body of all grains being deformed by duplex slip. They also noted that the slip systems {112}<111> and {123}<111> are geometrically equivalent to different proportions of primary and cross slip on {110}<111>. Using these assumptions and by considering the suggested biaxial stress system they predicted a texture showing a spread between the orientations (001) [1$\bar{1}$0] and (112) [1$\bar{1}$0]. This prediction differs from that of Calnan and Clews[8] in that the orientation {111}<110> is not suggested. This component is not universally reported in the literature and Dillamore and Roberts[9] did not find it in cold rolled varadium.

Much work is reported in the literature concerning the rolling-recrystallisation textures of b.c.c. metals[3, 7, 10, 11, 12, 13]. Ibe and Lucke[14] suggested from their results that a rotation of close to 27° about a <110> pole common to both deformation and recrystallisation texture could adequately account for the majority of orientation changes that occur during annealing. Dillamore[15] noting this observation, developed a theory of rolling-recrystallisation texture formation based on a criterion of oriented growth. In predicting possible nuclei orientations Dillamore[15] increased the angle found by Ibe and Lucke and considered a range of between 20° and 60° in order to simultaneously satisfy the growth relationships between a growing grain and two components of the deformed matrix. The nuclei most favourably oriented for growth into the various components of the deformation texture were then suggested to produce orientations corresponding to the main features of the recrystallisation texture. From this analysis it is reported that the principal components of the rolling-recrystallisation textures are accounted for and that the temperature dependence of annealing textures is qualitatively explained.

II Experimental Work

The research programme is designed to study rolling-recrystallisation textures in commercial grades of low carbon steel. Coupons of hot-rolled rimming, silicon-balanced and an aluminium-killed steel have been supplied for this work, pit analyses being given in Table I. Final analyses are not yet available. The killed and rimming steel were supplied at a nominal gauge of $^1/_4''$, and the balanced steel of $^5/_{16}''$.

Current theories of texture development suggest that the relative intensities of the various components commonly found in the rolling-recrystallisation textures of low carbon steels are, for a given annealing temperature, dependent upon the magnitude of the recrystallisation temperature. Factors such as prior recovery treatments and time at annealing temperature are also considered to be of importance in that relative boundary displacements during growth will be influenced.

It is proposed to vary the recrystallisation temperature of the killed steel by controlling the level of aluminium and nitrogen in solution prior to cold rolling. This will be effected by suitable quenching and tempering operations. Brine quenching from above 1300°C will be used, tempering operations being conducts at lower temperatures (- 700°C) for various times. 'Recrystallisation' temperatures are to be determined using hardness measurements and isochronal annealing treatments at different temperatures. The annealing time will be short, 5 minutes, in order to avoid effects caused by the precipitation of aluminium nitride. For a material of a given 'recrystallisation' temperature, increased annealing treatments will be given at various ratios of the 'recrystallisation' temperatures. Pole figures and electron microscopy will be used to study the end results.

The effect of recovery upon rolling-recrystallisation textures is being studied in both the killed and rimming steels, some results being given in this report. This work is to be extended to include studies of the dislocation configurations that results from cold work and recovery. Transmission electron microscopy will be used.

The pole figures given in this report were obtained from samples of the stock material that had been thermally cycled four times through the α/γ phase change. It was thought that this treatment would help to randomise any hot-rolling textures which may have been present. During the final cycle the materials were slowly cooled from 900°C. Rolling was carried out using $5^1/_4''$ diameter rolls, the reductions per pass being between 0.010" and 015".

Air-recirculation furnaces were used for recovery treatments, final annealing at 680°C (30 mins.) being carried out in a resistance wound tube furnace through which dry argon was passed.

200 pole figures were obtained, 0 - 75° out from centre, using the Schulz reflection technique on a Siemen's texture unit.

Textural variation through the sheet thickness after rolling, and rolling and recrystallisation, are also being studied.

III Results and Discussion

i. Textural variations through the sheet thickness

During rolling the stress system acting upon a workpiece surface is dependent upon the frictional conditions pertaining to the roll-metal interface, increasing friction favouring deformation of the surface layer by shear in the rolling direction.

The depth in the through thickness direction below which surface conditions no longer influence deformation texture formation is not well documented since variations in rolling conditions are thought to have a significant influence. It is likely that the shape of the slip-line field is of prime importance, this being determined by variables, many of which are inter-related, such as the shape of the friction hill, roll diameter, reduction per pass, nominal sheet thickness and the flow strength and work hardening capabilities of the material.

Through thickness textural variations, particularly in the annealed condition, are of significance both industrially, where forming properties may be affected, and in the laboratory where the problem of texture representation is to be considered.

In the current research programme consideration is being given to this source of texture variation but major work, involving slip line field analysis, is not being conducted. Textures are being determined, both after rolling and after rolling and annealing (30 mins. 680°C in dry Argon) in samples of the killed steel cold rolled from 0.25″ to 70% reduction in increments of $\simeq$ 0.012″ between dry $5^1/_4$″ diameter rolls. Pole figures are being determined at approximately 0.003″ increments from the sheet surface.

Figures 1 and 2 give the deformation textures at the sheet surface and middle respectively. In figure 2 an intense region at the centre of the pole figure shows a spread about the rolling direction of the orientation {100}<110>. The intensity falls from $3\frac{1}{2}$ times random at the centre of the pole figure to 2 times random 30° out from the centre

The remainder of the texture may be described by the fibre components observed by Haesnner and Weik[3] the maximum intensities recorded in this case being between 2 and $2\frac{1}{2}$ times random.

The surface texture is shown in figure 1. Again the {100}<110> component and its spread about the rolling direction predominates but the whole texture is less well developed than that at the centre of the sheet. At the middle of the pole figure the intensity is $2\frac{1}{2}$ times random but as the spread increases to 23° about the rolling direction the intensity drops to 2 times random.

The spread in texture, both at the surface and at the centre of the material, makes exact analysis of the pole figures difficult but it would appear that the relative intensity of $\{111\}$, compared to the rest of the pole figure is higher at the surface of the material than at its centre. This may be of importance during the development of recrystallisation textures.

The effect of increasing the amount of cold reduction upon the development of surface texture may be gauged by reference to figure 3, a 200 pole figure of the surface of the killed steel after 85% cold reduction. Compared to the results shown in figures 1 and 2 the texture is more highly developed. The intensity at the centre of the pole figure is $3\frac{1}{2}$ times random and the spread about the rolling direction is such that for orientation of up to 40° out the intensity level is greater than 2 times random. The fibre texture is more marked but evidence of the onset of its depletion is shown by the presence of intense regions not observed in material rolled to 70% reduction. These new peaks correspond to a spread about the $\{112\}\langle 110\rangle$ orientation.

Further results are required for intermediate depths below the sheet surface in order to map the deformation textural variations through the sheet thickness. The effect on material annealed for $\frac{1}{2}$ hour at 680°C has also to be studied. A separate report on this aspect of the work will be given on completion.

II Rolling texture development (inside textures)

The rolling textures developed in low carbon steels and other b.c.c. metals are well known. The main object in including such a study in the present work is to characterise the starting materials to be used in recovery and recrystallisation experiments.

i. Aluminium-killed steel

The well-developed texture shown in figure 4 is typical of the cold rolling textures produced at high deformations in b.c.c. metals. This particular sheet was cold rolled 90%. The texture consists of two major orientations: $\{100\}\langle 110\rangle$ spread about the rolling direction plus a component spread about $\{211\}\langle 110\rangle$. Comparison with figure 2 shows that the $\{211\}\langle 110\rangle$ orientation does not show such a marked relative intensity, compared to the $\{100\}\langle 110\rangle$ component at the smaller amounts of cold work (70%).

ii. Rimming steel

Figure 5 shows the 200 pole figure at 95% cold reduction. The result obtained is similar to that shown in figure 4 except that in this case the intense regions are more marked. This is almost certainly due to the higher rolling reduction given to the rimming steel.

III Recrystallisation textures

i. A large number of factors influence the development of preferred orientations during the commercial production of annealed low-carbon steel sheet. These may be sub-divided into the two groups below, interaction between factors listed under the different groups frequently being of importance.

(a) Process variables

1) Casting technique.
2) Soaking temperature.
3) Soaking time.
4) The initial and finishing hot rolling temperatures.
5) Cooling rate during water cooling after hot rolling.
6) Coiling temperature.
7) Total amount of cold reduction.
8) Reduction per pass.
9) Rolling variables, (e.g. magnitude of front or back tensions. lubrication conditions etc.)
10) Time at ambient temperature prior to annealing.
11) Heating rate to annealing temperature.
12) Annealing temperature.
13) Annealing time.
14) Furnace atmosphere.

(b) Material variables

1) Solid solution composition prior to annealing.
2) The abundance type, disperson etc. of secondary phases such as carbides and sulphides.
3) Cold worked texture
4) Grain size.

For a steel of any given composition the material variables (b) above are controlled by the process variables (1) to (10) and for a given annealing treatment the recrystallisation texture will be mainly determined by the cold worked texture, material variables influencing recrystallisation mechanisms being able to cause modifications. For a predetermined degree of cold reduction the 'inside' cold rolling texture of low carbon steel is not significantly influenced by rolling variables, the only factor of importance being the texture in the material prior to cold rolling. This will depend upon hot-mill processing, high temperatures and low strain-rates favouring the production of a randomly oriented material. Whiteley and Wise(16) found that for both rimming and aluminium-killed steels a reduction in hot mill finishing temperature from 860° to 770° altered the hot rolling texture from that of random to one having a {110} component in the plane of the sheet. Subsequent cold-rolling followed by annealing showed that the randomly-oriented hot-rolled material gave a higher {111} components in the final recrystallisation texture. Work by Atkinson[17] et al has shown this component to be {111}<110>.

The effects of other process variables are not thought to influence the formation of the 'inside' rolling texture. Surface textures, and the depth to which they penetrate the sheet should be dependent upon process variables Nos. 7, 8 and 9 in that these should influence slip line field configuration. Available information showing the effects of such factors is very scanty.

The influence of annealing temperature upon rolling-recrystallisation textures developed in b.c.c. metals has been reported in the literature [3, 7, 10, 11, 12, 13].

At temperatures just above that required for recrystallisation the major components of the rolling texture tend to be retained but there is a tendency for a <110> fibre texture to form about the rolling direction, {111}<110> and {112}<110> usually predominating. At intermediate temperatures above the recrystallisation temperature the component {111}<112> forms at the expense of the others. At still higher temperatures a further major component is formed which, according to Stablein and Möller[10], may be described as {335}<7,12,3>.

It is known[17] that for a given aluminium-killed steel annealed at a set temperature the intensity of the {111}<110> component is enhanced the lower is the hot-band coiling temperature. Low coiling temperatures favour the retention of aluminium and nitrogen in solid solution, such that after subsequent cold work aluminium nitride will be precipitated during the final annealing treatment. The precipitation of aluminium nitride during the annealing process is thought to elevate the recrystallisation temperature such that the formation of the {111}<110> component is favoured.

As is indicated in the foregoing discussion the rolling-recrystallisation textures obtained in commercial grades of low carbon steel sheet are, in general terms, reasonably well understood, although much more information is needed concerning the recrystallisation processes involved and their kinetics. It is considered that effects due to precipitation, elements in solid solution and recovery will be of importance in determining the relative intensities of the various components commonly found in recrystallisation textures. The work at Cranfield is designed to study some of these factors.

ii) Recovery

As a result of recovery the amount of strain energy in a given crystal decreases with time at recovery temperatures and since recovery is orientation dependent the relative driving forces for the growth of the various major components of a recrystallisation texture will probably be influenced by prior recovery treatment. Results obtained to date indicate that this is true.

Figure 6 is the 200 pole figure obtained from the killed steel after 90% cold reduction followed by an anneal in dry argon for 30 minutes at 680°C. The specimen, approximately 0.025″ thick, was placed in the furnace at temperature in order to minimise recovery effects prior to the onset of recrystallisation. Certain features of the cold worked texture are unaltered by annealing in that the $\{100\}\langle 110\rangle$ component and its spread about the rolling direction is observed to be retained, but it is to be noted that the intensity is diminished (C.f. Fig. 4) to 2 times random at the centre falling to $1^3/_4$ times random 20° out. Other components may be described as spreads about the $\{111\}\langle 112\rangle$ and $\{111\}\langle 110\rangle$ orientations. The effect of recovery prior to annealing upon the recrystallisation texture shown in Fig. 6 is given in Fig. 7. The recovery treatment was for 10 minutes, at 210°C. A significant difference is observed in that the $\{100\}\langle 110\rangle$ component is much more intense ($3\frac{1}{2}$ times random at the centre of the spread) and is not much reduced from that observed in the cold rolled material (see Fig. 4). The $\{111\}\langle 110\rangle$ component is much reduced, its intensity now being less than random. The influence of this recovery treatment is surprising considering the low temperature at which it was carried out. Raising the recovery temperature to 380°C (Fig. 8) produces similar results to those obtained at 210°C except that the intensity of $\{100\}\langle 110\rangle$ is increased slightly to 4 times random, i.e. equal to that in the cold worked material.

From these results it would appear that prior recovery favours the retention of the $\{100\}\langle 110\rangle$ component of the rolling texture. The work of Hu[18] on the recrystallisation of rolled silicon/iron single crystals has shown that crystals having the orientation $\{100\}\langle 110\rangle$ do not change their orientation during rolling or during subsequent annealing at 700°C. Recrystallisation was not observed, crystal softening occuring purely by recovery. Extrapolation of Hu's results to polycrystalline materials leads to the possibility that $\{100\}\langle 110\rangle$ cold worked textural components need a large driving force in order to effect recrystallisation. The effect of recovery will be to increase the driving force required and hence the probability of recrystallisation is further diminished.

The reason why the $\{100\}\langle 110\rangle$ component and its spread about R.D. disappears at higher recrystallisation temperatures is not understood but two possibilities exist:-

a) Recrystallisation occurs due to the large thermal activation.

b) It is absorbed by other growing components of higher boundary mobility.

It is thought that the former will be more favourable for the larger orientation spreads, the latter being more important at the ideal orientation $\{100\}\langle 110\rangle$.

Pole figures of the rimming steel after various recovery treatments prior to annealing are not yet available. The recrystallisation texture, after 95% cold reduction plus 30 minutes at 680°C is given in Fig. 9. Compared to the recrystallisation texture of the killed steel, Fig. 6, it is observed that the $\{111\}\langle 211\rangle$ component shown in Fig. 9 has a higher relative intensity. This is probably due to the greater amount of cold reduction given to the rimming steel (95% compared to 90%) such that its recrystallisation temperature could be lower than that of the killed steel, the formation of $\{111\}\langle 211\rangle$ thus being more favoured. An interesting feature of Fig. 9 is the presence of intense peaks $17\frac{1}{2}^\circ$ out from the centre of the pole figure. Similar peaks were observed by Stablein and Moller[10] in silicon-iron alloys annealed at 'intermediate temperatures'.

Work on the effects of recovery is to continue, pole figures measurements and electron microscopy being used as experimental techniques. The inter-relationships between the degree of cold work, recovery treatments and final annealing time and temperature are to be studied. Results obtained to date, particularly at 210°C are to be verified.

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Table I

Steel Compositions (Pit Analysis)

Steel	C	S	P	Mn	Cu	Ni	Sn	N_2	Si
Si. Balanced	.100	.018	.034	.39	.032	.021	.008	.0022	.018
Al. killed	.067	.018	.014	.34	.284	.021	.007	.0034	-

The pit analysis of the rimming steel is not known.

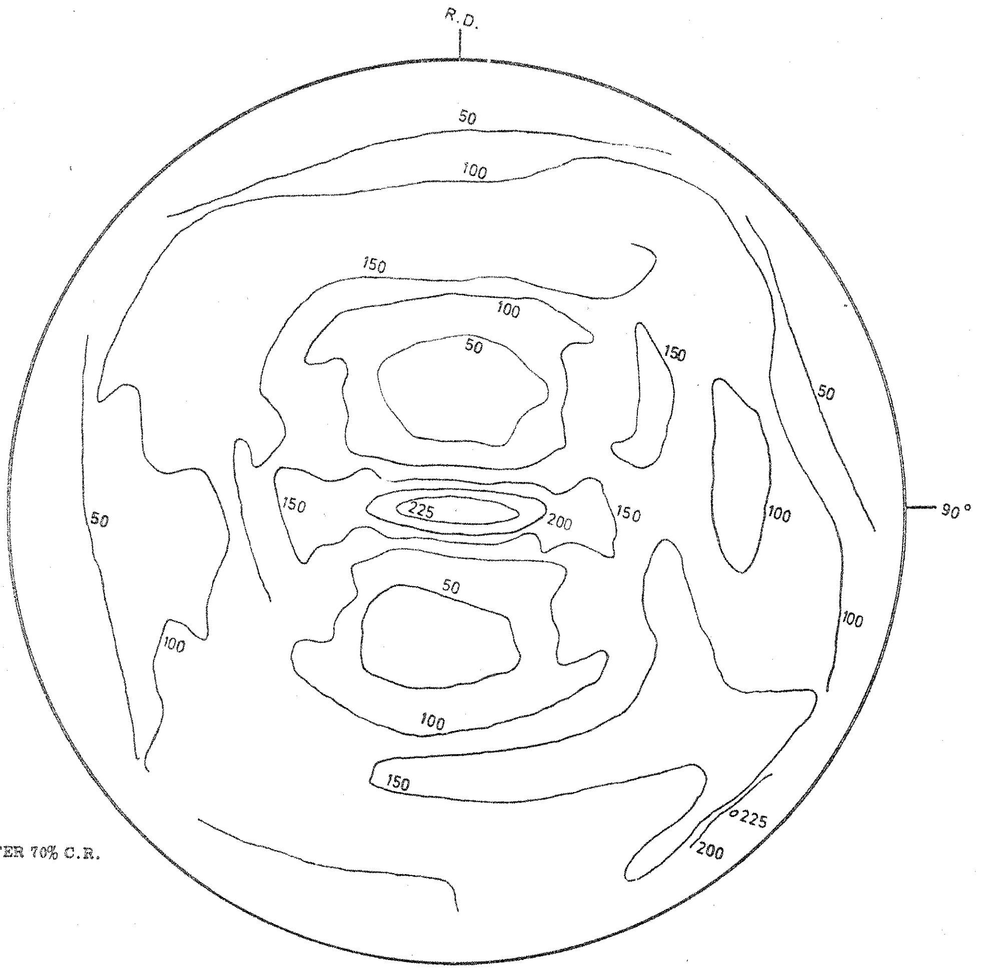

FIG. 1
Al KILLED STEEL.
SURFACE TEXTURE AFTER 70% C.R.
200 POLE FIG.

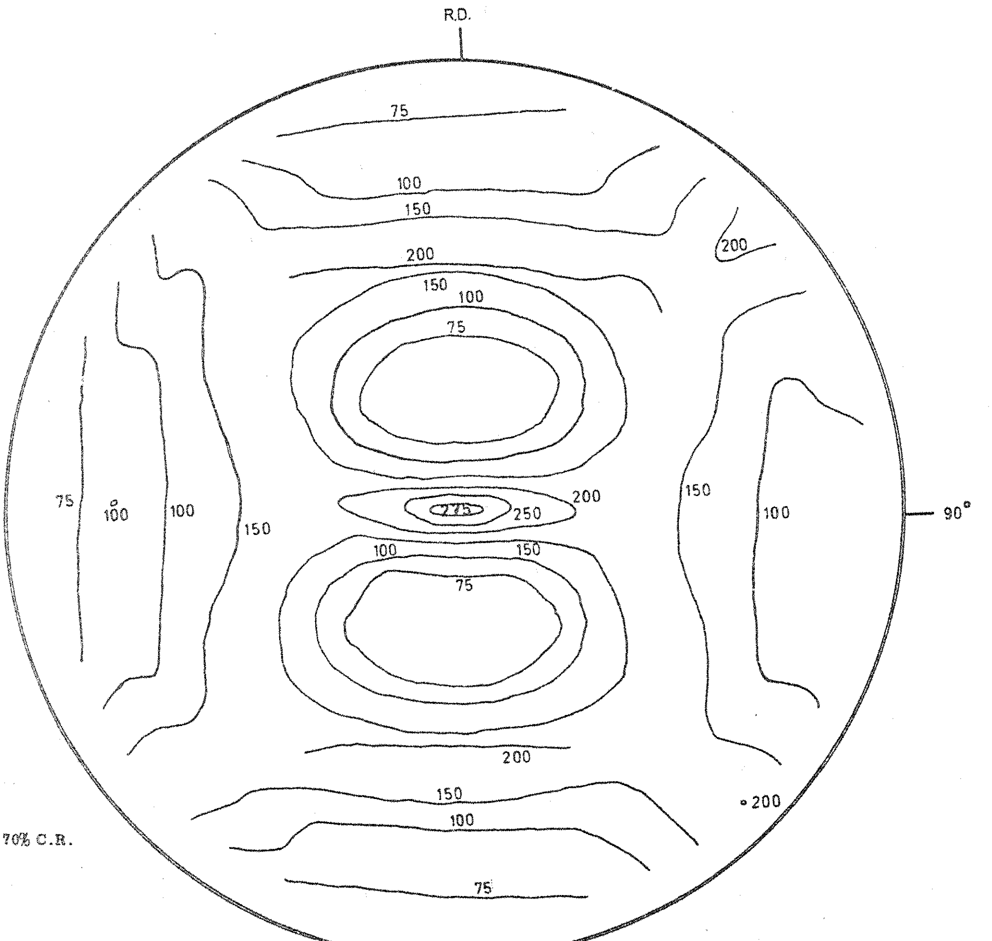

FIG.2
Al KILLED STEEL.
INSIDE TEXTURE AFTER 70% C.R.
200 POLE FIG.

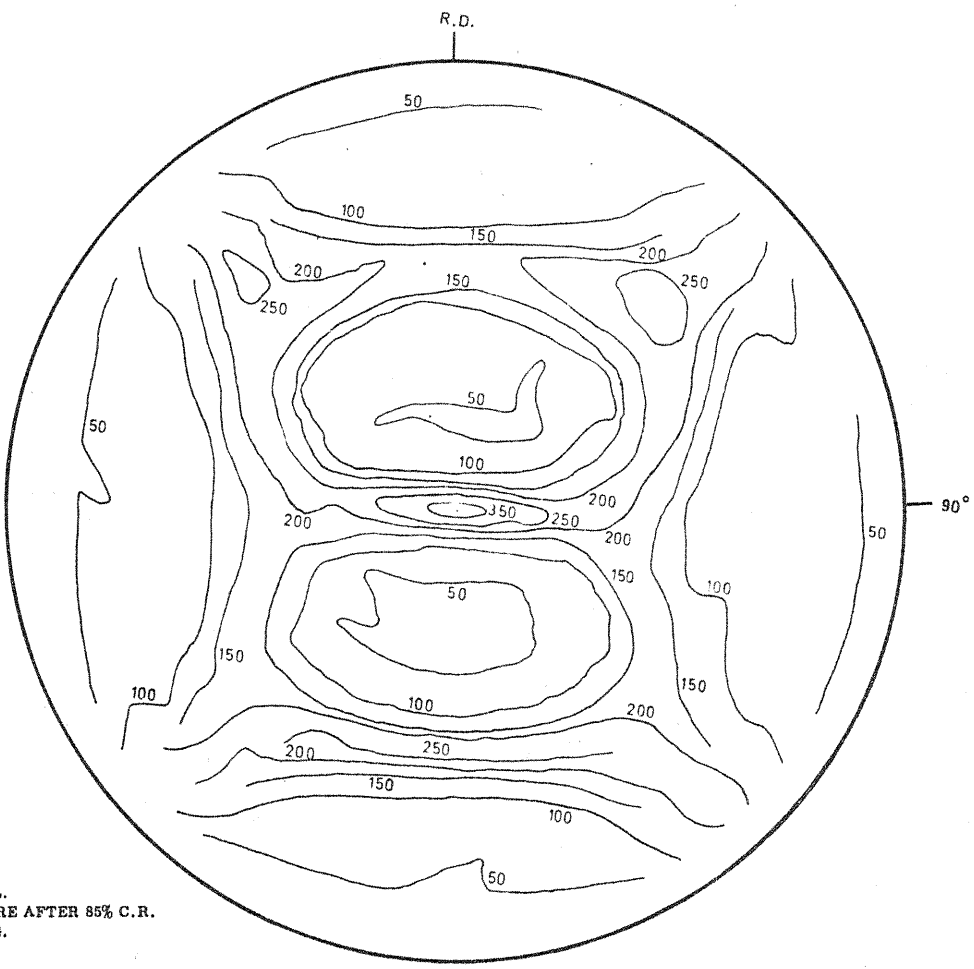

FIG. 3
Al KILLED STEEL.
SURFACE TEXTURE AFTER 85% C.R.
200 POLE FIG.

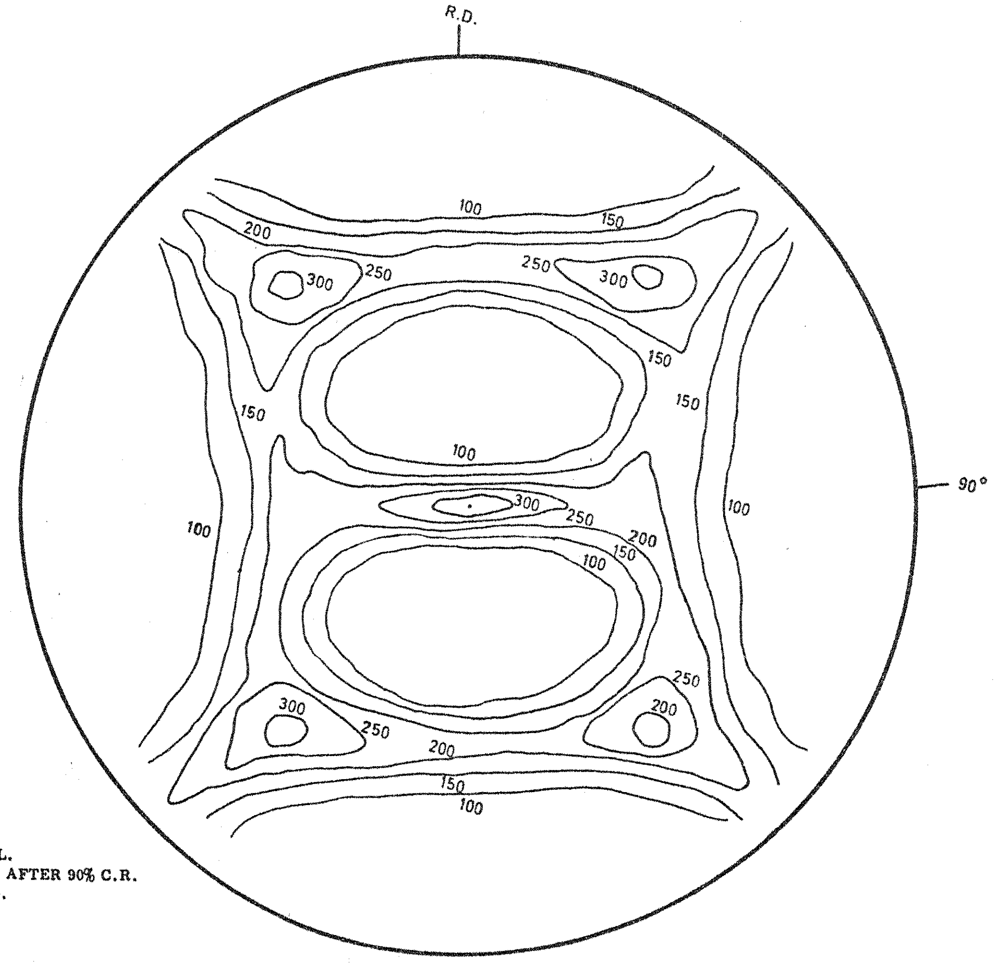

FIG. 4
Al KILLED STEEL.
INSIDE TEXTURE AFTER 90% C.R.
200 POLE FIG.

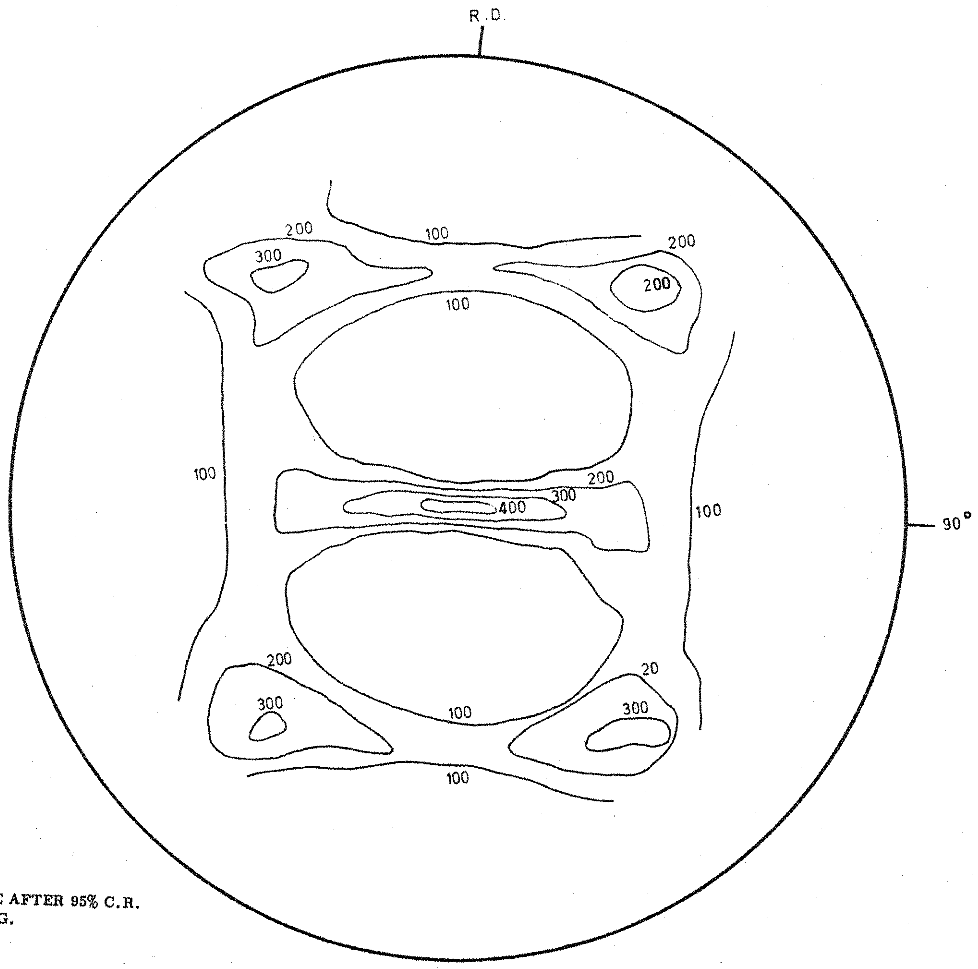

FIG. 5
RIMMING STEEL
INSIDE TEXTURE AFTER 95% C.R.
200 POLE FIG.

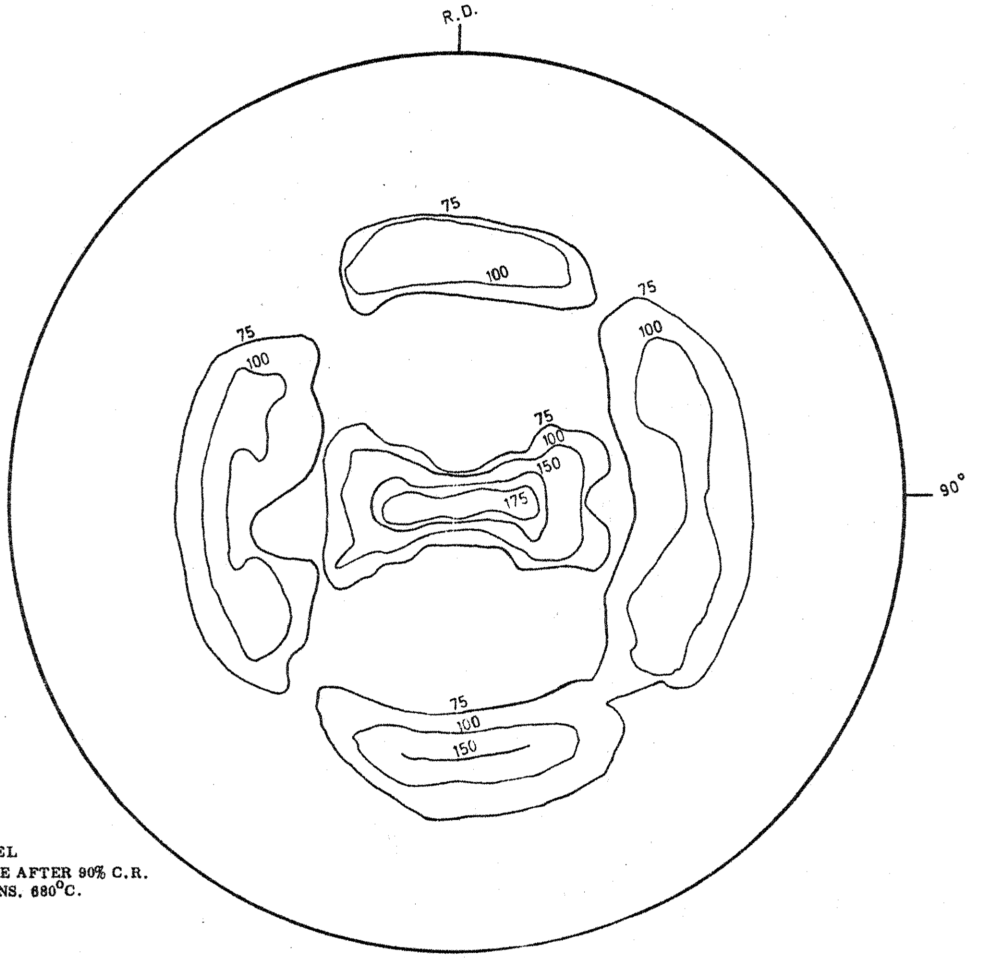

FIG. 6
Al KILLED STEEL
INSIDE TEXTURE AFTER 90% C.R.
PLUS 30 MINS. 680°C.

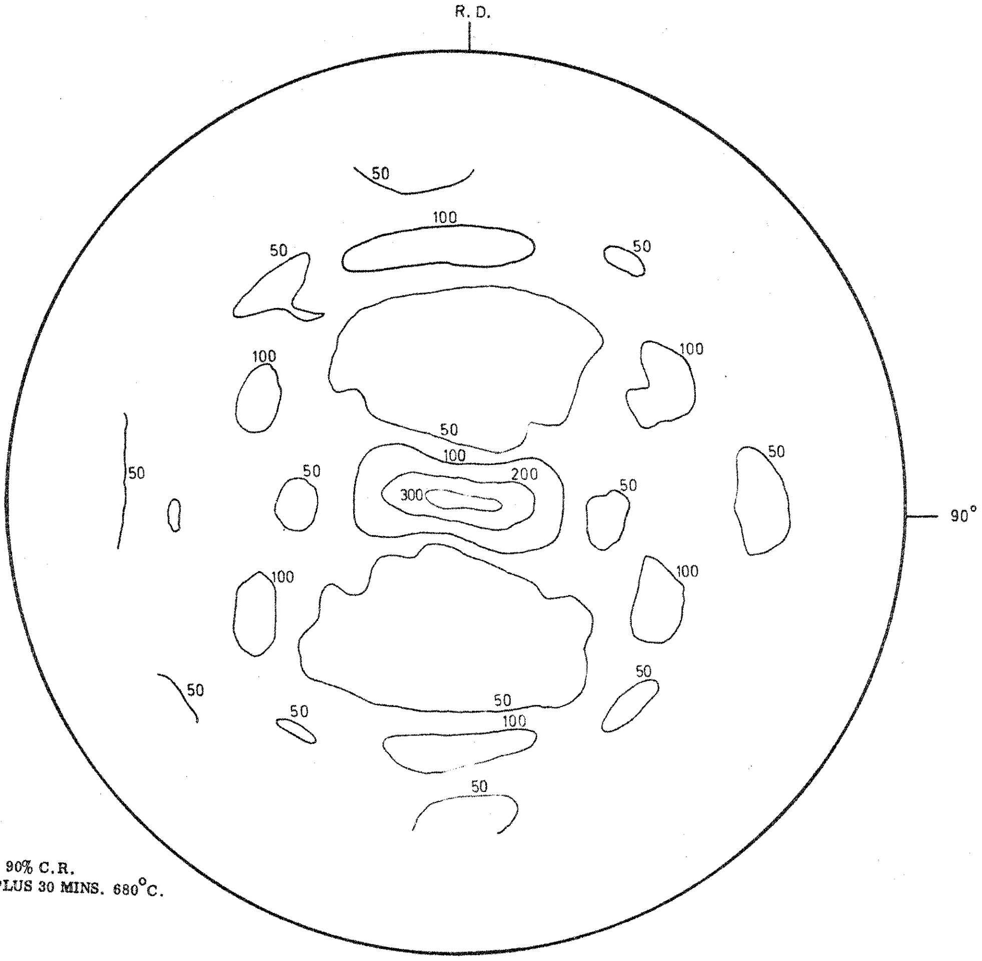

FIG. 7
Al KILLED STEEL.
INSIDE TEXTURE AFTER 90% C.R.
PLUS 10 MINS. 210°C PLUS 30 MINS. 680°C.

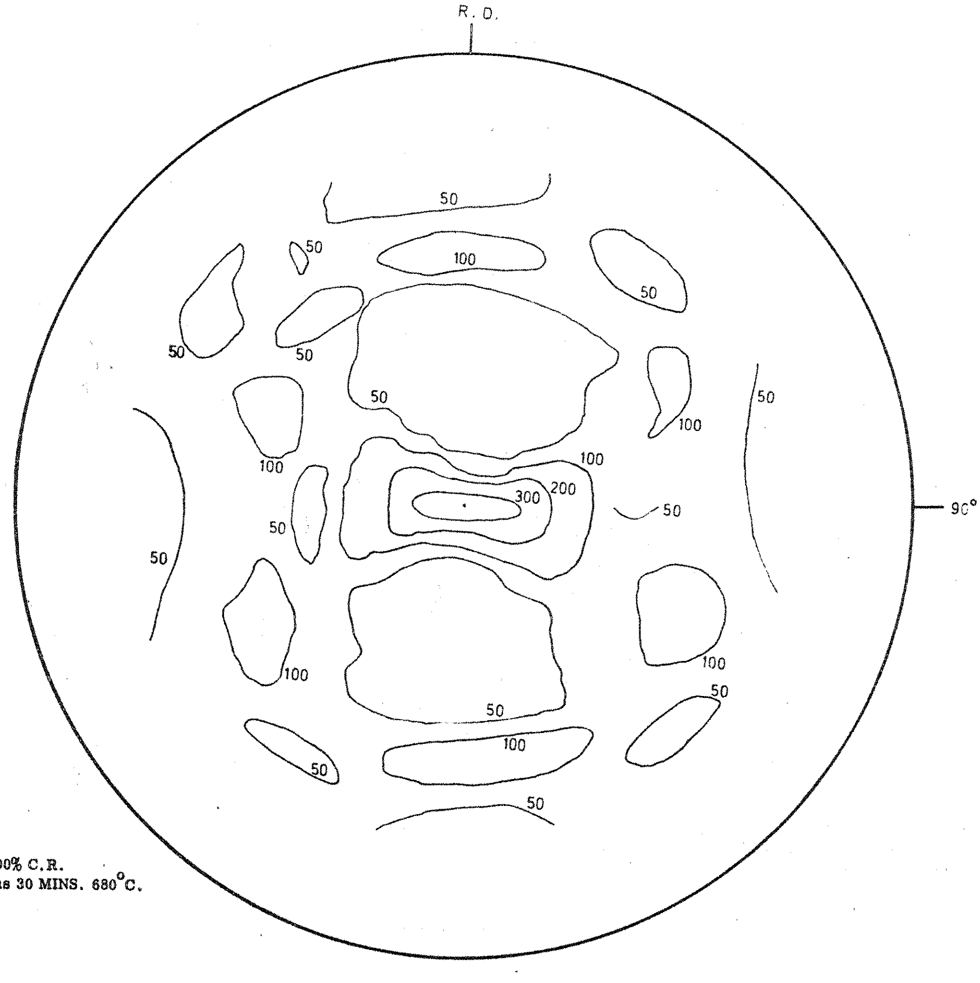

FIG. 8
Al KILLED STEEL.
INSIDE TEXTURE AFTER 90% C.R.
PLUS 10 MINS. 380°C plus 30 MINS. 680°C.

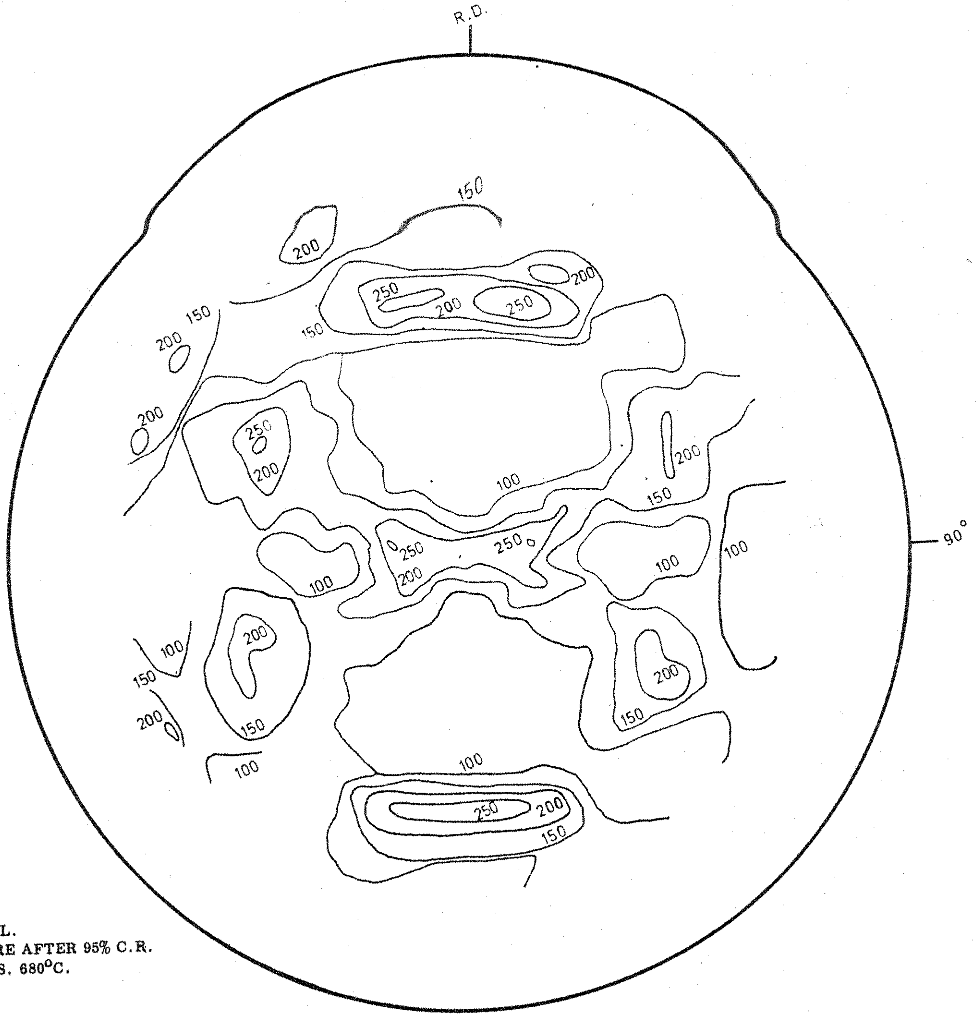

FIG. 9
RIMMING STEEL.
INSIDE TEXTURE AFTER 95% C.R.
PLUS 30 MINS. 680°C.