An investigation of aluminum-iron alloys from powders as possible electrical strip.

J. F. Dalgleish
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AN INVESTIGATION OF ALUMINUM-IRON ALLOYS
FROM POWDERS AS POSSIBLE ELECTRICAL STRIP

A Thesis
Submitted to the Faculty of Graduate Studies through the Department of Engineering Materials in Partial Fulfilment of the Requirements for the Degree of Master of Applied Science at the University of Windsor

by

J. F. Dalgleish

Windsor, Ontario
1967
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ABSTRACT

A study of some of the magnetic properties of low Al-Fe strip produced from powders has been conducted.

The mixed powder alloys were compacted and sintered at 1300°C. Cold rolling (90 pct. reduction) with one intermediate stress relief anneal was used to produce 0.011 in. thick strip. Various final annealing times and temperatures were employed in an attempt to develop a preferred orientation.

Material containing 1.75 wt pct. Al and 2.70 wt pct. Al, annealed at 1200°C or 1300°C for 12 hours under low dewpoint H₂, possessed a large-grained Goss texture, (110)[001]. The 60 cps power loss (15,000 Gauss peak) is less than 1 watt/lb and the DC magnetostriction (15,000 Gauss) has an average value of about $7.0 \times 10^{-6}$ in./in. although the results are widely scattered. Secondary recrystallization appears to begin at about 1100°C. The tensile strength of the material is, roughly, double that of pure iron.

While such Al-Fe strip is mechanically and magnetically inferior to grain-oriented Si-Fe alloy strip of commercial production, improvement of the texture and reduction of the inclusion content should produce a material magnetically equivalent to grain-oriented Si-Fe strip.
ACKNOWLEDGMENTS

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Acknowledgment is also due to Peace River Mining and Smelting Ltd. who aided this work both technically and financially and to the Westinghouse Corporation (Pittsburgh) whose technical assistance was extremely valuable.

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Due to the fact that aluminum has very much the same effect as silicon on the properties of iron, it has been suggested\textsuperscript{1,2} that Al-Fe alloys might be valuable as electrical strip. At room temperature (20° C) aluminum increases the resistivity of iron approximately linearly from 9.7 $\mu$ohm-cm by 10 $\mu$ohm-cm for each one weight pct.\textsuperscript{*} of Al added, up to approximately six weight pct. of aluminum. Silicon is only slightly more effective (11.6 $\mu$ohm-cm/one pct. Si) over the same alloy range. Aluminum and silicon both decrease the saturation magnetization of iron (1710 Gauss) linearly by about 50 Gauss/one pct. up to about six pct. of the alloying element.

Magnetic properties such as permeability, coercive force, anisotropy constant, core loss and magnetostriction have been measured on both single crystals and strip material and found to be comparable to Si-Fe alloys. In addition, Al-Fe alloys containing up to 10 pct. aluminum can be cold rolled, while an alloy of about 5 pct. Si is brittle.

The intention of this work was to evaluate low Al-Fe alloys as possible electrical strip. The properties of power loss and magnetostriction have been measured on material treated to develop a preferred orientation. The other important magnetic properties, coercive force and permeability, can in general be related to power loss: an increase in

\textsuperscript{*} Unless otherwise stated, all percentages are in weight per cent.
permeability implies a decrease in coercive force while a lowering of the power loss is accompanied by a decreased coercive force. Some mechanical tests have also been made.

The oxidation of aluminum makes it difficult to introduce large quantities of this element to an iron melt. The virtues of powder metallurgy methods for the production of Al-Fe alloys are considerable. They may be summarized as follows;

(a) Careful control of the sintering atmosphere restricts oxidation.
(b) Liquid phase sintering in this system aids in homogenization and reduces the porosity of the as-sintered material.
(c) Composition can be easily controlled.
(d) High purity is achieved and such elements as carbon, nitrogen, vanadium, chromium, titanium and magnesium, which can greatly influence magnetic properties at trace levels of concentration, can be controlled.

A. Magnetostriction

Magnetostriction is the name given to the effect whereby a ferromagnetic material changes dimensions in a magnetic field. A recent report attributes transformer vibration and noise to;

(a) variations in permeability in each lamination.
(b) design characteristics which cause flux transfer from lamination to lamination.
(c) magnetostriction, which is the exclusive cause of longitudinal vibration of a laminate and contributes largely to the vibratory oscillation occurring across a transformer arm.

Magnetostriction changes the volume of a transformer strip as well as the length. Generally speaking, the volume magnetostriction has a
small effect provided that the magnetic intensity is not above saturation. With this limitation in mind we can neglect volume magnetostriction and consider linear magnetostriction exclusively.

Each crystallographic direction has a saturation magnetostriction, $\lambda$, which represents the fractional change in length of the crystal when magnetized to saturation in that direction. These are expressed as $\lambda_{100}$ for the saturation magnetostriction in the [100] direction, $\lambda_{110}$ for the saturation magnetostriction in the [110] direction, etc. In general, the saturation magnetostriction of a cubic single crystal is expressed as $^{4,5}$:

$$
\lambda = \Delta l/l = h_1(\alpha_1 \beta_1 + \alpha_2 \beta_2 + \alpha_3 \beta_3 - 1/3) + h_2(\alpha_1 \beta_1 \beta_2 + \alpha_2 \beta_3 \beta_3 + \alpha_3 \beta_3 \beta_1) + h_3 s
$$

$$
+ h_4(\alpha_1 \beta_1 + \alpha_2 \beta_2 + \alpha_3 \beta_3 + 2s/3 - 1/3)
$$

$$
+ 2h_5(\alpha_1 \alpha_2 \beta_1 \beta_2 + \alpha_2 \alpha_3 \beta_3 \beta_3 + \alpha_3 \alpha_1 \beta_3 \beta_1)
$$

where

$$
s = \alpha_1 \alpha_2 + \alpha_2 \alpha_3 + \alpha_3 \alpha_1
$$

$$
\lambda = \text{relative fractional change in length of the crystallographic direction with direction cosines } \beta_1, \beta_2, \beta_3 \text{ with respect to the crystal axes, and when the crystal is magnetically saturated in the crystallographic direction with direction cosines } \alpha_1, \alpha_2, \alpha_3 \text{ with respect to the crystal axes.}
$$
$h_1 \ldots h_5 = \text{magnetostriction constants}$

$h_1 = 3/2(\lambda_{100})$

$h_2 = 3/2(\lambda_{111})$

If the magnetostriction is measured on the (110) crystal plane in the [001] direction (this is the single crystal equivalent to the Goss texture which is developed in commercial Si-Fe transformer material) equation (1) becomes:

$$\lambda = h_1(1 + 3\cos2\theta)/6 + h_3(7 - 4\cos2\theta - 3\cos4\theta)/32$$

$$+ h_4(9 + 20\cos2\theta + 3\cos4\theta)/48$$

(2)

where $\theta$ is the angle between the [001] direction and the direction of the applied field.

With the field applied exactly in the [001] direction $\theta = 0$, then:

$$\lambda = 2/3(h_1 + h_4)$$

(3)

For convenience the saturation magnetostriction measured in the (110) plane and along the [001] direction is called $\lambda_G$. For Si-Fe alloys experimental results fit the equation which is derived by using only the first two terms of equation (1) - implying that $h_4$ is negligible. This leads from equation (3) to:

$$\lambda_G = 2/3(h_1)$$
It follows that;

$$
\lambda_G = \lambda_{100}
$$  \hspace{1cm} (4)

Figure 1 is a plot of $\lambda_{100}$ versus weight per cent of Si and Al. The silicon alloy results are taken from Carr and Smoluchowski$^6$ and Hall$^7$ and the aluminum alloy results from Hall$^7$. These results demonstrate that the saturation magnetostriction $\lambda_{100}$ is, at worst, the same for low alloy concentrations; at best, that the Al-Fe system has a slightly lower $\lambda_{100}$ up to approximately 5 atomic pct. Al. This implies that $\lambda_G$ for low alloy concentrations of Al in Fe will produce equivalent or lower saturation values than Si-Fe alloys of the same concentration level in single crystals.

Equation (4) can be used to describe the saturation magnetostriction of a Goss texture, (110)[001], polycrystalline strip provided that it is recognized that in a polycrystalline material other factors may also affect the magnetostriction. Measurements$^9$ of single crystal (110)[001] magnetostriction in a polycrystalline matrix have shown that as the orientation of the surrounding grains away from the [001] direction in the rolling direction (and magnetizing direction) increases, the magnetostriction is no longer adequately described by equation (4). A second term must be included to account for the elastic intergranular activity.

The domain structure of the magnetized material can be used to qualitatively discuss the magnetostriction$^4,10,11$. The formation of domains in a crystal reduces the magnetic free energy of the crystal. A domain, consisting of all of the atoms with their magnetic moments aligned in the same direction, will grow or diminish in size through domain
boundary migration during magnetization. In the bcc system there are basically only two domain boundary conditions, 180° and 90°, representing
the change in direction of the magnetic moment in neighbouring domains
across a boundary. During magnetization the crystal lattice is deformed
in the direction of domain magnetization with those domains in the
direction of magnetization growing at the expense of less favourably
oriented domains. In a single crystal or in a polycrystalline material
with an orientation having [001] in the direction of magnetization, most
of the domains will be aligned along the length of the sample, since
[001] is a direction of easy magnetization. This implies that most of
the domains will have 180° walls with adjacent domains magnetized in
opposite directions. If only 180° walls are present, at the end of each
domain a magnetic free pole exists, giving the material a high demagneti-
zizing energy. Ninety degree walls at the ends of domains with 180°
boundaries permit the direction of domain magnetization to change without
a free pole condition. Considering only the 180° domain walls, the
application of an external field causes the domains with magnetic moments
in the direction of magnetization to grow by domain wall movement at the
expense of the grains with magnetic moments opposite to the direction of
the applied field. Only a 180° change of flux direction occurs and no
magnetostriction results. At high flux densities, when 90° rotations
are forced and 180° domain walls that are not exactly in the direction of
magnetization are rotated into the direction of magnetization, will the
magnetostriction increase. This explains why the plot of magnetostriction
versus flux density is either exponential or hyperbolic rather than linear.
Under AC excitation the magnetostriction tends to be larger than its DC
value for the same peak flux density due to the increased effectiveness

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of the 90° domain walls. As the value of the magnetostriction drops in a material, the difference between AC and DC values becomes negligible. A material with low DC magnetostriction will have low AC magnetostriction; and a material with zero DC magnetostriction will have zero AC magnetostriction.

It can be appreciated from the above discussion that as the size of the domains increases (i.e. the number of domain walls decreases) and as the orientation is perfected and the number of inclusions, impurities or voids (which oppose domain wall migration) decreases, the magnetostriction will approach zero even at relatively high inductions. Several papers have reported9,12,13,14,15 low magnetostriction at quite high induction levels in both single crystal and polycrystalline Si-Fe alloys. These same reports also demonstrate the dependence of magnetostriction on the direction of magnetization, shape and size of the sample (which affect the domain pattern), and grain size (which also affects the domain structure). With all of these dependent variables it is not surprising that a great deal of variation in magnetostriction for samples taken from the same sheet of Si-Fe material has been found.

No references have been found for magnetostriction measurements on polycrystalline Al-Fe alloys with low Al content.

If negligible magnetostriction were the only requirement for electrical strip then both the Si-Fe and Al-Fe systems could satisfy the need. The polycrystalline saturation magnetostriction (assuming random orientation) of both alloy systems is plotted in Figure 2. Again for low alloy concentrations the Al-Fe system has lower magnetostriction than the corresponding Si-Fe material. These results were obtained from the equation7;
Unfortunately, the remaining structure sensitive electrical and magnetic properties are not compatible with a randomly oriented grain structure.

B. Power Loss

The power dissipated in a magnetic material arises from two sources: the hysteresis loss and the eddy current loss.

The magnetically irreversible nature of a ferromagnetic material gives rise to a hysteresis loop. When a ferromagnetic material is magnetized by a field, $H$, a flux density, $B_H$, is induced into the material. When $H$ is returned to zero the flux density remains at a value $B_r$ such that $0 < |B_r| < |B_H|$. $B_r$ is called the remanent magnetization (or remanence), and in the case where $B_H$ is sufficient to saturate the material, $B_r$ is referred to as the retentivity. To return $B$ to zero, a magnetizing field, $H$, must be applied in the opposite direction. When $B = 0$ this value of $H$ denoted, $H_c$, is called the coercive force, or, for saturation conditions, coercivity. It can be seen that in one hysteresis cycle, more energy is put into the material than is recovered. This excess energy is dissipated as heat in the specimen and is called the hysteresis loss, $W_h$. The magnitude of $W_h$ is directly proportional to the area of the hysteresis loop. The hysteresis loss per cycle is assumed to be independent of frequency.

$$W_h = (A_h) \cdot f$$  \hspace{1cm} (6)

where

$A_h = \text{area of hysteresis loop}$

$f = \text{frequency}$
Domain wall displacement is responsible for the hysteresis loss, and any impediment to the motion of the domain walls (i.e., grain boundaries, inclusions and voids) will increase the hysteresis loss. The coercive force, \( H_c \), is proportional to the field necessary for a domain wall to traverse any imperfection.

The application of an alternating field to the material induces eddy currents due to flux change. The direction of eddy current flow is such that the secondary magnetic field generated by the eddy currents opposes the applied field; that is, the eddy currents flow parallel to the surface of the material and perpendicular to the direction of the applied field. The heat generated by eddy current flow represents the eddy current loss. Rigorous mathematical evaluation of the eddy current loss is not possible because of the complex nonlinear relationship between the magnetizing field, \( H \), and the flux density, \( B \), as determined by the hysteresis loop. By making the erroneous assumption that the permeability \( (\mu = B/H) \) of the material is both constant and uniform, it is possible to express the eddy current loss, \( W_e \), as:

\[
W_e = K \frac{2^2 2}{\pi t \rho B f} \tag{7}
\]

where

- \( K \) = constant
- \( t \) = thickness of strip
- \( B \) = flux density
- \( f \) = frequency of excitation
- \( \rho \) = resistivity

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Because of the assumptions made in deriving equation (7) the measured eddy current loss is always greater than the predicted value. The difference between the two values is referred to as the anomalous loss.

The total power loss, \( W_p \), is the sum of the hysteresis loss (6) and the eddy current loss (7):

\[
W_p = W_h + W_e
\]  

(8)

The power loss is sometimes referred to as the iron loss. When the material is in the form of a magnetic core for either a transformer or rotating electrical apparatus, the power loss is referred to as the core loss. When a single strip of alloyed material is being evaluated, the term power loss is most appropriate. As the operating frequency increases the eddy current loss term becomes increasingly dominant in equation (8). Examination of equation (8) suggests that a material with the following properties will have the lowest power loss:

(a) relatively high resistivity
(b) a large grain structure
(c) a preferred orientation having an easy direction of magnetization in the direction of the applied field
(d) low impurity and inclusion levels
(e) no internal strains
(f) a thin section

Some of these properties are also applicable to the reduction of magnetostriction.
"Hipersil" which represents about the best commercially produced transformer strip is a 3.25 pct. Si in Fe alloy strip, 0.011 in. thick. This grain-oriented material has a (110)[001], Goss, texture with an average grain diameter of 3 to 5 mm. The power loss (or core loss) averages about 0.5 watts/lb with a positive linear DC magnetostriction of about $5.0 \times 10^{-6}$ in./in. measured at 15,000 Gauss. The grain size of the material varies from as large as 8 mm in diameter to only 0.5 mm. This variation in grain size suggests a variation in the degree of orientation and is reflected in the variability of both the power loss and magnetostriction, which deviate from their average values by as much as 25 pct. Such a scatter in the constitution of the material demonstrates the influence of domain pattern and domain wall motion, both of which are highly structure sensitive.

C. Preferred Orientation

The foregoing discussion has established the desirability of a preferred orientation. The Goss texture, (110)[001], which is the only commercially developed texture to date, is produced in 3 pct. to 3.25 pct. Si-Fe alloys by means of the following procedure:

(a) hot rolling to 0.080 in. - 0.100 in.
(b) cold rolling to approximately 0.025 in.
(c) stress relief anneal in the range $800^\circ$C to $900^\circ$C
(d) cold rolling to 0.011 in. - 0.014 in.
(e) short decarburizing anneal to reduce carbon content to about 0.005 pct. (recrystallizes the material)
(f) final anneal of several hours at approximately $1200^\circ$C to produce a Goss texture by secondary recrystallization
The final cold reduction is about 50 pct., which has been shown \(^{17}\) to produce the minimum power loss values. The \(\gamma\) (gamma) loop in the Si-Fe phase diagram closes at approximately 2.5 pct. Si so that at compositions above this Si content, cooling from a high temperature annealing process does not cause a phase change. The deformation texture after cold rolling is mainly \((100)<011>\) with minor components of \((112)<110>\), \((111)<110>\) and \((111)<112>\) possible. The primary recrystallization texture is usually related to the deformation texture by rotations of \(25^\circ\) to \(35^\circ\) about \((110)\) poles. The remaining components of the primary recrystallization texture can be related to rotations about \((100)\) poles. The Goss texture grows from any of these textures by secondary recrystallization provided that primary grain growth is inhibited either by the specimen thickness effect or by inclusions. In some cases \(^{16}\) the presence of inclusions is necessary for secondary recrystallization to occur.

The other texture that has been developed \(^{18,19,20,21,22}\) is the cube texture, \((100)[001]\), which grows out of the same primary recrystallization texture, provided, again, that primary grain growth has been inhibited. The thickness effect is the inhibiting mechanism in thin-gauge material which recrystallizes to the cube texture. (A third texture known as the Littmann texture, approximately \((120)<001>\), has received little attention.) The cube texture, which to date has been limited to laboratory production, has better magnetic properties perpendicular to the rolling direction than the Goss texture.

Many papers have been published \(^{17,18,20,22,23,27}\) to show that minimum surface energy is the driving force for secondary recrystallization and that the surface energy conditions are affected by variations in furnace atmosphere, temperature, heating rate, sample composition, and
inclusion content in the sample. Orientations which have a minimum surface energy under a given set of the above conditions will grow. In a high purity atmosphere (inert or reducing) or in a vacuum, the \{110\} planes have the lowest surface energy. Adsorption of impurity atoms from the atmosphere, which tend to decrease the surface energy, have a greater effect on the \{100\} planes than on the \{110\} planes; thus, under impure annealing conditions the cube texture is favoured. If too impure an atmosphere is used, grain growth will be inhibited completely. This effect can be overcome by increasing the annealing temperature.\(^{24}\)

A paper by Kochnov and Gol'dshteyn\(^{25}\) drew several significant conclusions;

(a) The course of primary recrystallization is not affected by the degree of deformation (although the texture that develops will change as the amount of deformation is altered).

(b) The degree of deformation has a considerable effect on the starting temperature of secondary recrystallization.

(c) Furnace atmosphere does not influence primary recrystallization provided that it is protective.

Under industrial conditions (hydrogen atmosphere with some impurities) in which a Goss texture develops, grains with a \((110)[001]\) orientation are the first to nucleate during primary recrystallization and by the time that primary recrystallization has been completed they are two to three times the size of the average recrystallized grain. Processes that require more than one cold reduction have a better chance of producing \((110)[001]\) primary recrystallized grains because a grain with this orientation when cold rolled, recrystallizes to the same orientation. Most stress relief annealing treatments are sufficiently above the
primary recrystallization temperature (approximately 575°C) to permit recrystallization to take place. Thus, after one rolling and stress relief anneal a certain amount of the material will have recrystallized to (110)[001]. The second reduction and recrystallization will produce additional (110)[001] oriented grains without the loss of those produced with the first rolling. Primary grains with this orientation act as secondary recrystallization nuclei for the development of Goss texture.

Walker and Howard developed a Goss texture in 3.25 pct. Si-Fe strip using powders. The pure metal powders were mixed and compacted. Sintering temperatures and the dewpoint of the sintering atmosphere were varied. After sintering, all samples were subjected to the same rolling schedule and final anneal. Figure 3, which is taken from their results, illustrates that the development of secondary recrystallization to the Goss texture is dependent on the moisture content of the sintering atmosphere and that, generally, an increase in moisture content is required for secondary recrystallization as the sintering temperature is raised. The minimum secondary recrystallization temperature was approximately 1100°C. This is in agreement with the conditions for oxidation of silicon in moist hydrogen and suggests that silica acts as the inhibitor of primary grain growth, hence permitting secondary recrystallization to occur during final annealing. Assuming that secondary recrystallization begins at a temperature sufficiently high for the decomposition or dissolution of dispersed inclusions it would follow that a lower inclusion content would be found in materials with a Goss texture than in a randomly oriented material of the same silicon content. However, this is not the case. The authors conclude that the inhibiting silica particles are so finely dispersed as to be invisible
under the optical microscope. Grain-coarsening of low-carbon steel that has been deoxidized with aluminum and annealed above $A_{c3}$ is used as circumstantial evidence for such a conclusion.

The same authors (Walker and Howard$^{26}$) also developed a cube texture in a 3.25 pct. Si-Fe alloy prepared from powders. A Goss textured material was reduced 85 pct. by cold rolling, and annealed at approximately $900^\circ$C in pure hydrogen. As the original Goss texture improved, the final cube texture deteriorated.

The Al-Fe phase diagram for low aluminum content, shown in Figure 4, is very similar to the Si-Fe phase diagram. The closing of the $\gamma$ loop at approximately 1 pct. Al means that for alloys with more than 1 pct. Al high temperature annealing is possible without allotropic transformations occurring on cooling.

The cold rolling textures of Al-Fe alloys are similar to those of Si-Fe$^{27,28}$. It should, therefore, be possible to develop a Goss texture in Al-Fe alloys. Bozorth et al$^{29}$ and Foster and Pavlonic$^{30}$ have developed cube texture in a 3 pct. Al in Fe alloy.
II EXPERIMENTAL PROCEDURE

A. Production and Processing of Strip

A series of iron (99 pct. Fe; 0.01 pct. C; 0.14 pct. Mn) aluminum (99.5 pct. Al; 0.5 pct. Fe and Si) alloys containing, nominally, 1 pct., 2 pct., 3 pct. and 4 pct. by weight of aluminum were mixed from the pure metal powders in batches of approximately 100 grams by mortar and pestle. The Tyler screen analyses of the powders are:

<table>
<thead>
<tr>
<th>Screen Size</th>
<th>Fe (pct.)</th>
<th>Al (pct.)</th>
</tr>
</thead>
<tbody>
<tr>
<td>100 mesh</td>
<td>1.9</td>
<td>1</td>
</tr>
<tr>
<td>150 mesh</td>
<td>16.2</td>
<td>1</td>
</tr>
<tr>
<td>200 mesh</td>
<td>22.5</td>
<td>6</td>
</tr>
<tr>
<td>250 mesh</td>
<td>7.5</td>
<td></td>
</tr>
<tr>
<td>270 mesh</td>
<td></td>
<td>6</td>
</tr>
<tr>
<td>325 mesh</td>
<td>24.5</td>
<td>11</td>
</tr>
<tr>
<td>Through 325 mesh</td>
<td>27.4</td>
<td>75</td>
</tr>
</tbody>
</table>

From these mixes and from the "pure" iron powder, 13-gram samples were compacted at 80,000 psi in a Haller die - D1002. All compacts were sintered for twelve hours at 1300°C in a prepurified-grade hydrogen atmosphere (Dewpoint -50°C; O₂ <3 ppm). The dimensions of the sintered material were 1.25 in. x 0.5 in. x 0.185 in. x 0.010 in. - the final dimension being the thickness. The first cold reduction of from 50 pct.
to 55 pct. was performed on a lightly lubricated, two-high, 2.25 in. diameter hand roll. The edges, which suffered some cracking, were trimmed on a cut-off saw before the samples received a 15 minute stress relief anneal at 900°C under prepurified-grade hydrogen. Final cold rolling on the same mill to 0.011 in. ±0.001 in. (93 pct. to 95 pct. total reduction) produced a strip approximately 1 in. wide by 8 in. long. No edge cracking was experienced in the final cold reduction. By grinding the ends of the samples to a slightly oval shape after the first cold reduction, the material could be rolled into a straight strip.

Final annealing was carried out for twelve hours at temperature - either 1200°C or 1300°C under prepurified-grade hydrogen. The strips were insulated from one another with a fine alumina powder with a maximum of six strips annealed at once - the top strip being a dummy sample to prevent warping of the test samples and to ensure that each test sample was exposed to the same atmospheric conditions. This long annealing time was used in order to improve homogeneity and achieve the largest grain size possible. While commercial grain-oriented material has a relatively small grain size, by encouraging substantial grain growth it can be determined whether or not a preferred orientation is developing. If an orientation is found, the grain size can be reduced by shortening the time at temperature.

On heating, the furnace reached 1200°C in approximately two and one half hours with an additional hour necessary to attain 1300°C. From either temperature it required only about twenty minutes to cool to 1000°C with an added time of three hours to drop to 400°C. The samples were then pulled from the hot zone and allowed to cool to below 100°C before being exposed to the air.
Even with the relatively pure atmosphere used an oxide coat formed on the samples. In most cases it was fairly uniform. Before power loss and magnetostriction measurements were taken, this oxide was removed with concentrated HCl.

For comparative purposes a few samples were given final annealing treatments of six hours at $900^\circ$C and three hours at $1100^\circ$C. Wet chemical analysis of a representative portion of the samples was made to determine the exact aluminum content.

Power loss measurements were made on each strip, then the strain gauges were attached and DC magnetostriction values were determined. The samples were cut for X-ray studies following these measurements. Separate samples were used for the mechanical tests.

B. Measurement of Magnetic Properties

1. DC Magnetostriction

In recent years several types of apparatus have been suggested\textsuperscript{31,32} for measuring magnetostriction, all of which require the use of more than one strip for a single measurement. This fact alone makes them unsuitable for our experiments. With the exception of an optical lever apparatus, which is of limited accuracy, the strain gauge technique (originally introduced by J.E. Goldman in 1947) is the only method permitting evaluation of a single strip of material. The apparatus and calculations as outlined in Westinghouse Research Report 64-141-450-R2\textsuperscript{33} have been adapted to suit our requirements.

Figure 5 illustrates the magnetic circuit used. The specimen under test was placed between the two "C" magnets with the air gap set at a minimum width to allow free movement of the specimen. A non-magnetic stainless steel strip was mounted just beneath the magnet and a strain
gauge mounted on each side of each strip. Figure 24 shows the actual apparatus. Figure 6 illustrates the electrical circuit. The gauges were connected to form a Wheatstone bridge. The gauges on the sample (R₁ and R₃) formed one pair of opposite bridge arms that compensated for any bending stresses in the sample, while the other arms of the bridge (R₂ and R₄) on the stainless steel strip compensated for any differential expansion between the gauges and the test strip. By placing the gauges fairly close together, using temperature compensated gauges, having a magnet whose cross-section was much larger than that of the strip (i.e. negligible heating of the magnet at high induction levels of the strip), and covering the magnet to eliminate ambient temperature changes, there was insignificant bridge imbalance due to thermal effects.

For bridge balancing and calibration a potentiometer was paralleled with one of the bridge arms. The bridge was fed by a mercury cell and the null detector was a Hewlett Packard 419A DC Null Voltmeter.

The induction level in the test specimen was determined by placing a 100 turn air-gap compensated search coil around the sample and calibrating it against a Pye Scalamp Fluxmeter (cat. no. 8834). The DC magnetostriction was measured at 15,000 Gauss. The field strength did not vary appreciably along the length of the strip. All leads were twisted and shielded where possible.

The magnetostriction was determined by using the formula;

\[
\frac{d_1}{1} = \frac{R_1}{2(GF_1)} \cdot \frac{dR_b}{R_b^2} \quad \text{(in./in.)} \quad (9)
\]
where

\[ \frac{dl}{l} = \text{magnetostriction along the length of the sample (in./in.)} \]
\[ R_1 = \text{strain gauge resistance (ohms)} \]
\[ GF_1 = \text{gauge factor of strain gauge} \]
\[ R_b = \text{potentiometer setting (ohms)} \]
\[ dR_b = \Delta R = \text{change in potentiometer for bridge balance (ohms)} \]

The development of this formula appears in Appendix A.

Using Micro-Measurements Precision Strain Gauges (EA-06-500BH-120), which have a nominal resistance of 120 ohms and a gauge factor of approximately 2.1, equation (9) reduces to;

\[ \frac{dl}{l} = 28.5(\frac{\Delta R}{R_b})^2 \text{ (in./in.)} \]  \hspace{1cm} (10)

The following experimental procedure was employed.

1. With the mercury cell connected, the bridge was balanced using \( R_b \) and the value of \( R_b \) was noted.
2. The magnet was energized and the current adjusted to give the required flux density. The power was turned off.
3. The sample was demagnetized using 60 cps in the magnet with an oscilloscope serving as a null detector. Complete demagnetization was achieved when the bridge balanced without changing \( R_b \).
4. The magnet was again energized and the bridge rebalanced with \( R_b \).
5. Using \( \Delta R \), the difference in \( R_b \) before and after magnetization of the sample, and \( R_b \), the average value of the potentiometer before and after magnetization, the magnetostriction of the sample was calculated by equation (10).
6. The direction of magnetization of the sample was reversed and the reading was taken again. The average value of d1/1 was used.

7. The polarity of the magnetostriction was determined by using a Baldwin Lima Hamilton SR-4 Strain Indicator Type N in conjunction with a Baldwin Lima Hamilton Switching and Balancing Unit. While these instruments could measure the strain in only one gauge at a time, by magnetizing the sample above its saturation value, the polarity could be determined. This method assumed only that the forced magnetostriction was of the same polarity as the saturation magnetostriction. This is always the case.

The apparatus has a sensitivity of better than 1 x 10^-7 in./in.

2. Power Loss

The ASTM standard core loss test (A-34) requires the use of many Epstein strips (3 cm x 30 cm) from the same sample in order to evaluate the material. As our samples are to be evaluated individually, another method must be used. The practice of winding a single strip into a coil before final annealing, thus creating a magnetic core on which power loss measurements can be made, is not suitable since we wish to take magnetostriction readings on the same samples.

The transient heating curve method as described in Westinghouse Research Report 10-0604-1-R4 was employed.

The power dissipated in a specimen is determined by the transient heating curve produced when a magnetic field is turned on and off. The electrical circuit for the measurement of temperature change is shown in Figure 7. The thermocouples, 0.005 in. copper-constantan wire, which have an output of 41 uvolts per °C, are made by simply twisting wires together, due to the fact that heating by flame or welding breaks the
copper wire. The thermocouple is cut to the smallest size possible to avoid creating a large thermal mass. Condenser paper, 0.001 in. thick, is used to electrically insulate the thermocouples from the strip while maintaining good thermal contact. By subtracting the voltage output of the room temperature thermocouple which is placed in the vicinity of the strip, the amplified signal is just that of the temperature rise in the sample. A Hewlett Packard 419A DC Null Voltmeter was used as the amplifier of Figure 7, and it fed a 33-01-06 Bausch and Lomb Laboratory Recorder. The same magnetic circuit as described for the magnetostriction measurements was used with 60 cps power.

The peak induction level in the sample was determined with the circuit of Figure 8, using the following equation;

\[
\text{Gauss/mm def} = \left(\frac{C R}{N_2 A}\right) \times 10^8 \times \text{volts/mm}
\]

(11)

where

\[C = \text{capacitance (farads)}\]
\[R = \text{resistance (ohms)}\]
\[N_2 = \text{turns in search coil}\]
\[A = \text{cross-sectional area of sample (cm}^2\)]
\[\text{volts/mm} = \text{oscilloscope range setting}\]

See Appendix B for the development of this formula. Substituting in (11);

\[C = 4.9 \mu\text{farads}\]
\[R = 130.0 \text{ ohms}\]
\[N = 10 \text{ turns}\]
\[\text{oscilloscope setting} = .01 \text{ volts/mm}\]
Equation (12) was used to determine a peak flux density of 15,000 Gauss.

The specimen can be shown to obey the heat flow equation;

\[ P + KV \theta = cP(\partial \theta / \partial t) \]  \hspace{1cm} (13)

at all times, where

- \( P \) = power generated per unit volume in specimen (cal/cc -sec)
- \( K \) = thermal conductivity of specimen (cal/cm-sec-\(^\circ\)C)
- \( c \) = specific heat of specimen (cal/gm-\(^\circ\)C)
- \( \rho \) = density of specimen (gm/cc)
- \( \theta \) = temperature of specimen (\(^\circ\)C)
- \( t \) = time (sec)

This equation is evaluated for a rectangular specimen by applying the boundary conditions for a sphere of approximately the same volume. In doing so, the following results were obtained. Assuming perfect or near perfect insulation of the sample under non-steady-state conditions;

\[ P = cP[\partial \theta] / [\partial t]_{t_0} \]  \hspace{1cm} (14)

\[ \text{cal/cc-sec} \]
where
\[
\frac{\partial \theta}{\partial t} \bigg|_{t_0} = \text{initial slope of transient heating curve (°C/sec)}
\]
(see Figure 9)

With imperfect insulation and non-steady-state conditions;

\[
P = c_p \left[ \frac{\partial \theta}{\partial t} \bigg|_{t_1^-} - \frac{\partial \theta}{\partial t} \bigg|_{t_1^+} \right] \quad \text{(cal/cc-sec)} \tag{15}
\]

where
\[
\frac{\partial \theta}{\partial t} \bigg|_{t_1^-} = \text{slope of transient heating curve just before magnetic field is removed (°C/sec) (see Figure 9)}
\]
\[
\frac{\partial \theta}{\partial t} \bigg|_{t_1^+} = \text{slope of transient heating curve just after magnetic field is removed (°C/sec) (see Figure 9)}
\]

Using the following values;

1 calorie = 4.184 x 10^7 ergs
1 gram = 2.204 x 10^{-3} lb

\[
\text{ergs/sec x } 10^{-7} = \text{watts}
\]

the power loss, P, from equation (14) and equation (15) can be expressed as;

\[
P = 1.90 \times 10^3 \times c \left[ \frac{\partial \theta}{\partial t} \bigg|_{t_0} \right] \quad \text{(watts/lb)} \tag{16}
\]

or

\[
P = 1.90 \times 10^3 \times c \left[ \frac{\partial \theta}{\partial t} \bigg|_{t_1^-} - \frac{\partial \theta}{\partial t} \bigg|_{t_1^+} \right] \quad \text{(watts/lb)} \tag{17}
\]

As our samples were not insulated, equation (17) is the more applicable. However, as the assumed boundary conditions can hardly be considered equivalent to the thin strip of material, it was not expected that the
power loss readings would be accurate. Indeed, they were not. Difficulty arose first in duplicating the cooling curve of the samples. While the heating curve duplicated very well, the loss of heat from the sample after the magnetic field was removed varied from one run to another, yielding a variable slope. It was decided, therefore, to employ equation (16) which used only the reproducible part of the time-temperature response. Appendix C discusses the derivation and applicability of this equation.

Samples whose power loss had been measured by the ASTM standard method were evaluated using equation (16). The results of this evaluation are listed in Table 1 and plotted in Figure 10.

<table>
<thead>
<tr>
<th>Material</th>
<th>Average ASTM Power Loss (watts/lb)</th>
<th>Average Recorded Power Loss (watts/lb)</th>
<th>Min. Recorded Power Loss (watts/lb)</th>
<th>Max. Recorded Power Loss (watts/lb)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>0.51</td>
<td>1.35</td>
<td>1.20</td>
<td>1.50</td>
</tr>
<tr>
<td>2</td>
<td>1.70</td>
<td>5.75</td>
<td>5.50</td>
<td>6.00</td>
</tr>
<tr>
<td>3</td>
<td>4.15</td>
<td>18.5</td>
<td>15</td>
<td>22</td>
</tr>
</tbody>
</table>

Figure 10 served as a calibration chart for the power loss measurements. The scatter in the results increased with increased losses. Provided that the power loss was less than about 2.0 watts/lb the graph could be used with reasonable accuracy. The materials listed in Table 1 and described below were not of our production.

Material 1 is "Hipersil", 0.011 in., strip which is grain-oriented 3.25 pct. silicon alloy.

Material 2 is a 0.018 in. strip of 1.3 pct. Si which is non-oriented.
Material 3 is a 0.035 in. strip of 1.3 pct. Si which is also non-oriented.

The procedure for determining the power loss can be summarized as follows:

1. The amplifier was calibrated on the scale to be used by putting in a known DC signal and measuring the output.
2. Power was supplied to the magnet and the induction level set using the circuit previously described. Power was then switched off.
3. When thermal equilibrium had been achieved, the strip recorder was started and power applied to the magnet producing a transient heating curve. Power was then turned off and the sample allowed to cool.
4. The reading was repeated. The sample did not have to be demagnetized between readings as the state of magnetization of the sample did not enter into the derivation of the formula.
5. The sample was demagnetized using the oscilloscope as a null detector.
6. The initial slope of the heating curve was measured.

\[
\left(\frac{\Delta \theta}{\Delta t}\right)_{t_0} = \frac{V}{C_v \times 41 \times 10^{-6} \times \Delta t} \quad \text{\(^{\circ}\!\!\text{C/sec}\)} \quad (18)
\]

where

\(V\) = voltage increase in time \(\Delta t\)

\(C_v\) = voltage gain of amplifier
Output of thermocouple = $41 \times 10^{-6}$ volts/°C

7. Equation (16) was used to determine the "measured" power loss. Specific heat, $c$, varies from 0.108 cal/gm-°C for pure iron to approximately 0.112 cal/gm-°C for 4 pct. Al in Fe. An average value of 0.110 cal/gm-°C was used for all calculations.

8. Figure 10 was used to determine the actual power loss. Although this method requires the use of a correction factor, we have persisted in its use because the results are reproducible and the time-temperature method of power loss determination is the only one that can be used on a single strip of material.

3. Orientation Determination

X-ray examination was used to evaluate the degree of preferred orientation in the sample. The standard method of displaying a preferred orientation is by means of a pole figure. This method is not applicable to large-grained material as many grains must be exposed to the X-ray beam. Back reflection Laue photographs can be used to evaluate large grains but this method requires precise positioning of the sample with respect to the X-ray beam. Often several photographs must be taken of a single grain to determine its orientation with any accuracy. If a relatively large number of grains are to be examined, time required for analysis by either method is prohibitive. Moreover, as will be seen later, both large and small grained material are encountered in the results so that neither pole figure determination nor Laue photographs are suitable for studying all of the samples.

The method described below was developed to qualitatively determine the texture of the strip. The three high intensity Bragg reflections for
this bcc material are the 110, 200 and 211 reflections, having relative intensities of 100, 19 and 30 respectively. If many grains had a (110) plane in the surface of the sheet, the intensity of the 110 reflection relative to the 200 and 211 intensities would be larger than that found in a randomly oriented sample. Similarly, if the material had a (100) plane in the sheet surface it would be more difficult to detect the 110 reflection. The only two preferred orientations that commonly develop in this type of material have either the (110) plane (Goss texture) or the (100) plane (cube texture) parallel to the sheet surface so that examination of the three high intensity peaks determines which, if either, of the textures has developed.

A circular piece of each strip, approximately 0.8 in. in diameter, was cut from the strip and placed in the goniometer of the diffraction unit and set at the approximate Bragg angle for the 110 reflection (2\(\theta\) = 52°) as determined by the peak intensity of the recorder. The sample was then rotated in its own plane (i.e. with the sheet normal as axis of rotation) to maximize the peak. Once this had been accomplished the sample was rotated in its own plane 360° and the number and intensity of each peak noted during rotation. The procedure was repeated for the 200 reflection (2\(\theta\) = 77°) and the 211 reflection (2\(\theta\) = 99°). In many instances, particularly for the 110 reflection, the recorded intensity was above the background intensity (as determined by changing 2\(\theta\) by a few degrees) for 360° of rotation. This was an indication that either a large number of randomly oriented grains were being irradiated or, if only a few grains or perhaps one grain was irradiated, that the reflecting plane in question paralleled the plane of the sample. If the remaining reflections could be easily detected, the first condition existed.
The following code was used to classify the samples:

(a) X-ray reflection characteristics

A - high intensity 110 peaks alone were found and the intensity level was consistently above the background

B - 110 reflections were easily found but were not as intense as in A; 211 peaks were also detectable

C - 110 and 200 reflections were detectable but not 211

D - all three reflections were easily detected

E - 200 was the only reflection

(b) Apparent grain size

Removal of the oxide coat after annealing left the samples in a macro-etched condition so that the apparent grain size could be estimated as;

1 - very large grains; a maximum of two or three grains per sample

2 - smaller grains but still easily visible to the eye

3 - microscopic grain size

Thus, using a letter and a numeral the relative perfection of the orientation could be determined.

4. Mechanical Tests

Vickers hardness and tensile tests were made on several strips to compare the mechanical properties of the Al-Fe alloys to those of pure iron and grain-oriented Si-Fe. The tensile specimens were ground on a Tensilgrind apparatus. The grinding marks on the edges of the samples were smoothed with 2/0 emery paper so that any stress raisers ran the length of the sample rather than across the width. The tensile specimens were pulled on an Instron testing machine using a D load cell and a 1 in. gauge length extensometer. The yield strengths were determined using a 0.5 pct. offset on the stress-strain curves not exhibiting an ultimate yield strength.
III RESULTS

A. Chemical Analysis

Wet chemical analysis results are listed in Table 2.

Table 2 Wet Chemical Analysis

<table>
<thead>
<tr>
<th>Nominal wt pct. Al</th>
<th>Average wt pct. Al</th>
<th>Deviation</th>
<th>Average pct. Al Lost</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>0.75</td>
<td>±0.08</td>
<td>25.0</td>
</tr>
<tr>
<td>2</td>
<td>1.75</td>
<td>±0.10</td>
<td>12.5</td>
</tr>
<tr>
<td>3</td>
<td>2.70</td>
<td>±0.14</td>
<td>10.0</td>
</tr>
<tr>
<td>4</td>
<td>3.50</td>
<td>±0.20</td>
<td>12.5</td>
</tr>
</tbody>
</table>

Table 2 is the result of three different analyses performed on some of the magnetostriction and core loss samples. Two of the tensile test samples had Al contents outside these limits: 1.43 pct. (nominally 2 pct.) and 2.27 pct. (nominally 3 pct.). This suggested the possibility that some of the magnetostriction and core loss samples did not fit into Table 2, although those actually tested (15 pct. of all samples) did. The aluminum is believed to have been lost during mixing because of the fine aluminum powder size and during sintering due to oxidation of aluminum.

B. Magnetostriction and Power Loss

Tables 3 and 4 list the magnetostriction and power loss results for the 1200°C and 1300°C annealing treatments respectively. The magnetostriction results of Tables 3 and 4 are plotted in Figures 11 and 12.
Table 3  DC Magnetostriction and 60 cps Power Loss for Samples Annealed 12 Hours at 1200°C

<table>
<thead>
<tr>
<th>Wt pct. Al</th>
<th>DC Magnetostriction ($x 10^{-6}$ in./in.)</th>
<th>Power Loss (watts/lb)</th>
</tr>
</thead>
<tbody>
<tr>
<td>0</td>
<td>-6.7</td>
<td>1.06</td>
</tr>
<tr>
<td></td>
<td>-5.7</td>
<td>1.12</td>
</tr>
<tr>
<td></td>
<td>-3.4</td>
<td>2.00</td>
</tr>
<tr>
<td>0.75</td>
<td>-4.2</td>
<td>1.32</td>
</tr>
<tr>
<td></td>
<td>-2.3</td>
<td>1.06</td>
</tr>
<tr>
<td></td>
<td>-2.8</td>
<td>1.36</td>
</tr>
<tr>
<td></td>
<td>-3.4</td>
<td>1.67</td>
</tr>
<tr>
<td>1.75</td>
<td>4.4</td>
<td>0.54</td>
</tr>
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<td></td>
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<td></td>
<td>5.2</td>
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</tr>
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<td></td>
<td>2.2</td>
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</tr>
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<td></td>
<td>10.7</td>
<td>0.60</td>
</tr>
<tr>
<td></td>
<td>4.4</td>
<td>0.59</td>
</tr>
<tr>
<td>2.70</td>
<td>0.6</td>
<td>0.59</td>
</tr>
<tr>
<td></td>
<td>0.4</td>
<td>0.77</td>
</tr>
<tr>
<td></td>
<td>1.1</td>
<td>0.74</td>
</tr>
<tr>
<td></td>
<td>4.4</td>
<td>0.84</td>
</tr>
<tr>
<td></td>
<td>1.8</td>
<td>0.88</td>
</tr>
<tr>
<td></td>
<td>20.4</td>
<td>0.50</td>
</tr>
<tr>
<td></td>
<td>12.2</td>
<td>0.93</td>
</tr>
<tr>
<td>3.50</td>
<td>5.2</td>
<td>1.05</td>
</tr>
<tr>
<td></td>
<td>1.1</td>
<td>1.02</td>
</tr>
</tbody>
</table>

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Table 4  DC Magnetostriction and 60 cps Power Loss for Samples Annealed 12 Hours at 1300°C

<table>
<thead>
<tr>
<th>Wt. pct. Al</th>
<th>DC Magnetostriction ( \times 10^{-6} ) in./in.</th>
<th>Power Loss (watts/lb)</th>
</tr>
</thead>
<tbody>
<tr>
<td>0</td>
<td>-3.2</td>
<td>2.00</td>
</tr>
<tr>
<td></td>
<td>-4.4</td>
<td>0.99</td>
</tr>
<tr>
<td></td>
<td>-5.6</td>
<td>1.81</td>
</tr>
<tr>
<td></td>
<td>-5.3</td>
<td>1.48</td>
</tr>
<tr>
<td>0.75</td>
<td>-2.8</td>
<td>1.95</td>
</tr>
<tr>
<td></td>
<td>-3.6</td>
<td>1.40</td>
</tr>
<tr>
<td></td>
<td>-4.5</td>
<td>1.24</td>
</tr>
<tr>
<td></td>
<td>-3.7</td>
<td>1.48</td>
</tr>
<tr>
<td>1.75</td>
<td>5.1</td>
<td>1.25</td>
</tr>
<tr>
<td></td>
<td>4.4</td>
<td>1.46</td>
</tr>
<tr>
<td></td>
<td>5.5</td>
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</tr>
<tr>
<td></td>
<td>7.5</td>
<td>0.95</td>
</tr>
<tr>
<td></td>
<td>7.1</td>
<td></td>
</tr>
<tr>
<td></td>
<td>7.8</td>
<td></td>
</tr>
<tr>
<td>2.70</td>
<td>1.6</td>
<td>1.06</td>
</tr>
<tr>
<td></td>
<td>1.3</td>
<td>0.86</td>
</tr>
<tr>
<td></td>
<td>2.6</td>
<td>0.90</td>
</tr>
<tr>
<td></td>
<td>12.0</td>
<td>0.90</td>
</tr>
<tr>
<td></td>
<td>16.3</td>
<td></td>
</tr>
<tr>
<td></td>
<td>1.3</td>
<td></td>
</tr>
<tr>
<td>3.50</td>
<td>5.8</td>
<td>1.04</td>
</tr>
<tr>
<td></td>
<td>1.8</td>
<td>0.93</td>
</tr>
<tr>
<td></td>
<td>6.2</td>
<td>0.96</td>
</tr>
<tr>
<td></td>
<td>1.9</td>
<td>0.71</td>
</tr>
</tbody>
</table>
Table 5 lists the magnetostriction results for the samples annealed six hours at 900°C and three hours at 1100°C. No power loss measurements were taken for the 900°C samples.

Table 5 Magnetostriction and Power Loss for Samples Annealed at Low Temperatures

<table>
<thead>
<tr>
<th>Wt pct. Al</th>
<th>900°C DC Magnetostriction (x 10⁻⁶ in./in.)</th>
<th>1100°C DC Magnetostriction (x 10⁻⁶ in./in.)</th>
<th>1100°C Power Loss (watts/lb)</th>
</tr>
</thead>
<tbody>
<tr>
<td>0</td>
<td>-3.1</td>
<td>-4.6</td>
<td>1.67</td>
</tr>
<tr>
<td>0.75</td>
<td>-4.6</td>
<td>6.1</td>
<td>1.22</td>
</tr>
<tr>
<td>1.75</td>
<td>-6.1</td>
<td>5.4</td>
<td>1.26</td>
</tr>
<tr>
<td>4.0</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>2.70</td>
<td>4.7</td>
<td>10.0</td>
<td>0.99</td>
</tr>
<tr>
<td></td>
<td></td>
<td>13.6</td>
<td>0.83</td>
</tr>
<tr>
<td>3.50</td>
<td>8.5</td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

Figure 13 is a plot of all of the power loss readings from Tables 3, 4 and 5. The line is drawn through the average value of the 1200°C and 1300°C samples. Figure 14 records the magnetostriction of the low temperature annealed material.

A peak flux density of 15,000 Gauss for the power loss measurement as determined by equation (11) is accurate to ±300 Gauss (2 pct.). The major source of error is the uncertainty in the calibration figure. Using the extreme values of the recorded power loss listed in Table 1, the power loss measurements have a possible error of 6 pct. However, this error
does not include the variation in power loss of the individual samples used for calibration.

The error in the magnetostriction values caused by possible variations in the terms of equation (10) is approximately 5\%\(\text{pct.}\). This assumes, however, that the magnetization level is exactly 15,000 Gauss, which it is not. The flux density for the magnetostriction readings is limited in precision to ±500 Gauss due to the sensitivity of the fluxmeter. As the DC magnetostriction varies either exponentially or hyperbolically with flux density, a variation of this magnitude will create a considerable error. Using such a plot of magnetostriction versus flux density for a "Hipersil" strip it is estimated that this error is 15\%\(\text{pct.}\). Summing these errors, a possible error of 20\%\(\text{pct.}\) exists in the magnetostriction readings.

C. Grain Size

As expected, the largest grain sizes occurred in the material annealed at high temperature. Grain size differences that occurred with changing aluminum content were not affected by the annealing temperature as far as the 1200\(^\circ\text{C}\) and 1300\(^\circ\text{C}\) anneals were concerned. The pure iron samples had an equiaxial grain structure with the average grain diameter varying from 4 mm to 1 mm. The ASTM macro-grain size numbers for these values are M6.5 and M10 respectively. The edge areas of all the pure iron samples were composed of smaller grains having diameters of from 0.33 mm to 0.28 mm, which corresponds to M13.5 to M14 on the ASTM macro-grain size chart. A few of the pure iron samples had a region of elongated grains between the edge grains and the equiaxial centre grains. Figure 17 shows such a sample. This effect was present for both high annealing temperatures. The small edge crystals were a phenomenon of the pure iron samples and, to a lesser extent, the 0.75\%\(\text{pct.}\) Al-Fe alloy.
The 0.75 pct. Al samples had a small equiaxial grain structure with an average diameter of approximately 0.074 mm. This corresponds to the ASTM Austenite Grain Size No. 6. Figure 18 is representative of this structure. This grain size corresponds to ASTM micro-grain size 4.5. Figure 19 shows the few small edge grains of this alloy.

Table 6 Grain Sizes of a 2.70 pct. Al Strip Annealed 12 Hours at 1300°C

<table>
<thead>
<tr>
<th>Approx. Grain Diameters (mm)</th>
<th>Areas (mm$^2$)</th>
</tr>
</thead>
<tbody>
<tr>
<td>20 x 20</td>
<td>400</td>
</tr>
<tr>
<td>50 x 5</td>
<td>250</td>
</tr>
<tr>
<td>30 x 12</td>
<td>360</td>
</tr>
<tr>
<td>14 x 8</td>
<td>112</td>
</tr>
<tr>
<td>12 x 7</td>
<td>84</td>
</tr>
<tr>
<td>13 x 7</td>
<td>91</td>
</tr>
<tr>
<td>6 x 8</td>
<td>48</td>
</tr>
<tr>
<td>22 x 11</td>
<td>242</td>
</tr>
<tr>
<td>36 x 15</td>
<td>540</td>
</tr>
<tr>
<td>40 x 16</td>
<td>640</td>
</tr>
<tr>
<td>21 x 6</td>
<td>126</td>
</tr>
<tr>
<td>6 x 5</td>
<td>30</td>
</tr>
<tr>
<td>10 x 25</td>
<td>250</td>
</tr>
<tr>
<td>2 x 2</td>
<td>4</td>
</tr>
<tr>
<td>50 x 25</td>
<td>1250</td>
</tr>
<tr>
<td>13 x 5</td>
<td>65</td>
</tr>
<tr>
<td>15 x 16</td>
<td>240</td>
</tr>
<tr>
<td><strong>Average</strong></td>
<td><strong>280 (mm$^2$)</strong></td>
</tr>
</tbody>
</table>

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The alloys containing 1.75 pct., 2.70 pct. and 3.50 pct. Al had extremely large grains. As the samples were only one inch in width, the growth of many of the grains was limited to the length of the strip, effectively producing elongated grains. Affixing an average grain diameter to such a structure is deceptive and for an estimation of "average" grain size, the areas of the grains were considered. For example, Table 6 lists all seventeen grains found in one sample of 2.70 pct. Al. The average area of 280 mm² corresponds to M2.5. For these alloys, values of M4 to M1 were obtained. Figure 20 is a typical sample. An indication of the inclusion content is given by Figure 21. The higher Al alloys had an inclusion content that corresponded to ASTM (E45-63) Background Classification C or D.

None of the samples treated at 900°C experienced large grain growth. The pure iron samples had an average grain diameter of approximately 0.068 mm. This corresponds to the grain size of the 0.75 pct. Al samples treated at high temperature (i.e. M4.5). All of the aluminum alloys had average grain diameters of approximately 0.027 mm yielding a micro-grain size of about 7.5. A typical sample is shown in Figure 22.

A complete series of alloys was not treated at 1100°C. The average grain diameter of the small grains in the higher aluminum alloys (1.75 pct. and 2.70 pct. Al) varied from 0.025 mm to 0.038 mm, which is equivalent to ASTM micro-grain sizes 7.5 and 6.5 respectively. However, in these alloys a few large grains were found. Figure 23 shows the growth of a large grain into the smaller grains. These large, irregularly shaped grains were in the range M2 to M4: that is, they are as large as the average grain size of the material annealed at higher temperatures. It was difficult to estimate what percentage of the samples had a large grain.
structure as only one part (1 \text{ in.} \times 1 \text{ in.}) of the strip was polished. The large-grained part of the polished sample incorporated from 20 \text{ pct.} to 50 \text{ pct.} of the area.

These results are summarized in Table 7. The macro-grain sizes, prefaced by M, are all \#0.5.

<table>
<thead>
<tr>
<th>Wt. pct. Al</th>
<th>900°C for 6 Hrs.</th>
<th>1100°C for 3 Hrs.</th>
<th>1200°C for 12 Hrs.</th>
<th>1300°C for 12 Hrs.</th>
</tr>
</thead>
<tbody>
<tr>
<td>0</td>
<td>4.5</td>
<td>M6.5 to M10</td>
<td></td>
<td></td>
</tr>
<tr>
<td>0.75</td>
<td>7.5</td>
<td>4.5</td>
<td>4.5</td>
<td></td>
</tr>
<tr>
<td>1.75</td>
<td>7.5</td>
<td>7.5*</td>
<td>M4 to M1</td>
<td></td>
</tr>
<tr>
<td>2.70</td>
<td>7.5</td>
<td>6.5*</td>
<td>M4 to M1</td>
<td></td>
</tr>
<tr>
<td>3.50</td>
<td>7.5</td>
<td></td>
<td>M4 to M1</td>
<td></td>
</tr>
</tbody>
</table>

* These samples have some large grains - M2 to M4.

D. Preferred Orientation

Using the code previously described the results of Table 8 were obtained. The results are listed with positions relative to Table 3 (on the left) and Table 4 (in the centre). From Table 8 it is apparent that the small-grained structures of pure iron and the 0.75 \text{ pct.} Al alloy are fairly randomly oriented for the high temperature material. In two pure iron samples, in which all of the reflections were not detected (E and C), it is suspected that only one grain was irradiated during rotation of the sample. The large-grained alloys have a Goss texture, (110)[001], the quality of which varies appreciably.
Table 8 Preferred Orientation Evaluation

<table>
<thead>
<tr>
<th>Wt pct. Al</th>
<th>Orientation Code (1200°C samples)</th>
<th>Orientation Code (1300°C samples)</th>
<th>Orientation Code (1100°C samples)</th>
</tr>
</thead>
<tbody>
<tr>
<td>0</td>
<td>D2</td>
<td>E2</td>
<td>D2</td>
</tr>
<tr>
<td></td>
<td>D2</td>
<td>D2</td>
<td>D2</td>
</tr>
<tr>
<td></td>
<td>D2</td>
<td>C2</td>
<td>D2</td>
</tr>
<tr>
<td>0.75</td>
<td>D3</td>
<td>D3</td>
<td>D3</td>
</tr>
<tr>
<td></td>
<td>D3</td>
<td>D3</td>
<td>D3</td>
</tr>
<tr>
<td></td>
<td>D3</td>
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</tr>
<tr>
<td></td>
<td>D3</td>
<td>D3</td>
<td>D3</td>
</tr>
<tr>
<td>1.75</td>
<td>D1</td>
<td>A1</td>
<td>C1</td>
</tr>
<tr>
<td></td>
<td>A1</td>
<td>C1</td>
<td>D1</td>
</tr>
<tr>
<td></td>
<td>C1</td>
<td>C1</td>
<td></td>
</tr>
<tr>
<td></td>
<td>D1</td>
<td>A1</td>
<td></td>
</tr>
<tr>
<td></td>
<td>A1</td>
<td>C1</td>
<td></td>
</tr>
<tr>
<td></td>
<td>A1</td>
<td>A1</td>
<td></td>
</tr>
<tr>
<td>2.70</td>
<td>C1</td>
<td>A1</td>
<td>A1</td>
</tr>
<tr>
<td></td>
<td>C1</td>
<td>B1</td>
<td>B1</td>
</tr>
<tr>
<td></td>
<td>C1</td>
<td>A1</td>
<td></td>
</tr>
<tr>
<td></td>
<td>C1</td>
<td>A1</td>
<td></td>
</tr>
<tr>
<td></td>
<td>B1</td>
<td>B1</td>
<td></td>
</tr>
<tr>
<td></td>
<td>A1</td>
<td></td>
<td></td>
</tr>
<tr>
<td>3.50</td>
<td>A1</td>
<td>B1</td>
<td></td>
</tr>
<tr>
<td></td>
<td>C1</td>
<td>C1</td>
<td></td>
</tr>
<tr>
<td></td>
<td>A1</td>
<td>C1</td>
<td></td>
</tr>
<tr>
<td></td>
<td>C1</td>
<td>A1</td>
<td></td>
</tr>
</tbody>
</table>

Note: The apparent grain size of the 1100°C samples are listed here as being very large (1), since the grain boundaries of the small grains are not visible and, in the presence of a large grain, small grains appear to be a single grain.

E. Mechanical Tests

1. Tensile Results

The three oriented Si-Fe samples exhibited a yield point which occurred at approximately 0.5 pct. elongation. The iron and Al-Fe alloys did not
exhibit a yield point and a 0.5 pct. offset was used in estimating the yield strength. The ultimate tensile strengths are based on the original cross-sectional area of the sample. Only the 1.75 pct. and 2.70 pct. Al in Fe alloys were tested as they showed the most promise, magnetically speaking. The ultimate strength is also the failure strength as the samples failed by tearing. Figure 15 is a plot of the average values of Table 9.

<table>
<thead>
<tr>
<th>Wt pct. Al</th>
<th>Yield Strength (psi)</th>
<th>Ultimate Tensile Strength (psi)</th>
<th>Pct. Elongation at Ultimate Strength</th>
</tr>
</thead>
<tbody>
<tr>
<td>0</td>
<td>9,500</td>
<td>19,300</td>
<td>14.4</td>
</tr>
<tr>
<td>0</td>
<td>10,200</td>
<td>20,500</td>
<td>17.5</td>
</tr>
<tr>
<td>0</td>
<td>12,300</td>
<td>24,300</td>
<td>18.5</td>
</tr>
<tr>
<td>0</td>
<td>10,700</td>
<td>23,700</td>
<td>20.5</td>
</tr>
<tr>
<td>1.75</td>
<td>17,600</td>
<td>34,500</td>
<td>11.5</td>
</tr>
<tr>
<td>1.75</td>
<td>17,200</td>
<td>31,700</td>
<td>50.0+</td>
</tr>
<tr>
<td>2.70</td>
<td>21,100</td>
<td>37,500</td>
<td>13.5</td>
</tr>
<tr>
<td>2.70</td>
<td>17,600</td>
<td>33,500</td>
<td>36.0</td>
</tr>
<tr>
<td>2.70</td>
<td>20,400</td>
<td>33,500</td>
<td>13.3</td>
</tr>
<tr>
<td>2.70</td>
<td>20,200</td>
<td>34,500</td>
<td>16.5</td>
</tr>
<tr>
<td>2.70</td>
<td>26,000</td>
<td>43,200</td>
<td>39.5</td>
</tr>
<tr>
<td>2.70</td>
<td>26,200</td>
<td>42,100</td>
<td>16.5</td>
</tr>
<tr>
<td>3.25 pct. Si</td>
<td>53,800</td>
<td>57,000</td>
<td>10.0</td>
</tr>
<tr>
<td>3.25 pct. Si</td>
<td>46,800</td>
<td>48,700</td>
<td>6.3</td>
</tr>
<tr>
<td>3.25 pct. Si</td>
<td>45,300</td>
<td>48,600</td>
<td>8.3</td>
</tr>
</tbody>
</table>

2. Vickers Hardness

Table 10 lists the Vickers hardness (100 gm load) taken on the samples given a high temperature annealing treatment. Figure 16 is a plot of the average values of Table 10.
<table>
<thead>
<tr>
<th>Wt pct. Al</th>
<th>Vickers Hardness</th>
<th>Average</th>
</tr>
</thead>
<tbody>
<tr>
<td>0</td>
<td>70.5 78.8</td>
<td>75.3</td>
</tr>
<tr>
<td></td>
<td>68.8 79.5</td>
<td></td>
</tr>
<tr>
<td></td>
<td>71.9 82.2</td>
<td></td>
</tr>
<tr>
<td>0.75</td>
<td>115 103</td>
<td>111</td>
</tr>
<tr>
<td></td>
<td>118 113</td>
<td></td>
</tr>
<tr>
<td></td>
<td>111 105</td>
<td></td>
</tr>
<tr>
<td>1.75</td>
<td>125 131</td>
<td>124</td>
</tr>
<tr>
<td></td>
<td>121 121</td>
<td></td>
</tr>
<tr>
<td></td>
<td>130 113</td>
<td></td>
</tr>
<tr>
<td>2.70</td>
<td>157 135</td>
<td>138</td>
</tr>
<tr>
<td></td>
<td>144 142</td>
<td></td>
</tr>
<tr>
<td></td>
<td>126 124</td>
<td></td>
</tr>
<tr>
<td>3.50</td>
<td>139 142</td>
<td>143</td>
</tr>
<tr>
<td></td>
<td>146 158</td>
<td></td>
</tr>
<tr>
<td></td>
<td>138 137</td>
<td></td>
</tr>
<tr>
<td>**********</td>
<td>3.25 pct. Si</td>
<td>252</td>
</tr>
<tr>
<td></td>
<td>237 287</td>
<td></td>
</tr>
<tr>
<td></td>
<td>245 258</td>
<td></td>
</tr>
<tr>
<td></td>
<td>230 253</td>
<td></td>
</tr>
</tbody>
</table>
IV DISCUSSION OF RESULTS

A. Samples Given a High Temperature Annealing Treatment

None of the results showed any significant difference between the properties of the samples annealed at 1200°C and those annealed at 1300°C. The results of Table 7 suggest that alloys containing 1.75 pct. Al or more undergo secondary recrystallization at 1200°C or 1300°C to a large-grained Goss texture, (110)[111]. Pure iron and 0.75 pct. Al in iron annealed at 1200°C or 1300°C do not possess a large grain size and are relatively randomly oriented. Some preferred orientation is obtained in the primary recrystallization of the latter samples as they recrystallize from a cold-worked material, but it is not nearly as well developed as in the large-grained samples. Reference to Figure 4 demonstrates that both high annealing temperatures are within the limits of the γ and α + γ region of the Al-Fe phase diagram for pure iron and 0.75 pct. Al in Fe. On cooling, α grains nucleate in the high temperature matrix and a small grain size results.

It is suspected that the irregular grain pattern found along the edges of some of the pure iron and 0.75 pct. Al samples is a result of temperature gradients or cold working. The edges of the pure iron samples cool more quickly, hence encouraging undercooling, many nuclei, and a small grain size. That this phenomenon occurred to a lesser extent in the 0.75 pct. Al alloy can be explained by the fact that the transformation temperature is higher. This alloy, on cooling more quickly than pure
iron through the transition region, had a higher nucleation rate across
the entire sample, resulting in a smaller but more uniform grain size
than the pure iron samples. The large difference in grain size between
the edge grains and the centre grains is beyond that expected from a
variation in nucleation rate during cooling. It is known that the amount
of cold work prior to annealing affects the final grain size. Again, the
large difference in grain size is not anticipated for these heavy reductions.

It is felt, therefore, that the unusual grain pattern of the pure
iron and 0.75 pct. Al alloy may be produced by a combination of the two
effects; non-uniform cooling and cold working. Further testing would be
required to isolate the cause or causes of this effect.

On the other hand, the alloys of 1.75 pct. Al and more did not
undergo an allotropic transformation on cooling and retained their high
temperature structures. The inhibition of primary grain growth, which is
required for secondary recrystallization, appears to have been accomplished
by the inclusions in the samples, most of which would be aluminum oxide.
Some aluminum oxide would be present on the aluminum powder mixed with the
iron. On sintering, this oxide is not reduced. Indeed, it is quite
likely that the O₂ in the sintering atmosphere would increase the oxide
c Content somewhat, as the surface of the Al-Fe alloys was oxidized. The
fact that the pure iron samples were not oxidized precludes the formation
of an iron oxide during heat treatments.

The suggestion that the pure iron and 0.75 pct. Al alloy were
randomly oriented was enhanced by the magnetostriction readings. The
measured magnetostriction was negative, which agrees with the poly-
crystalline magnetostriction of Figure 2. If these alloys had a preferred
orientation, then the magnetostriction would have been positive as
suggested by Figure 1. On the other hand, the positive magnetostriction of the 1.75 pct., 2.70 pct. and 3.50 pct. Al alloys suggests the presence of a preferred orientation. The wide scatter in these readings can be attributed to several sources. The X-ray results show that the preferred orientation varied widely from sample to sample and a variation in magnetostriction was, therefore, bound to occur. The (110) plane rotated out of the surface of the sample is usually greater than the rotation of the [001] direction from the rolling direction. Both of these variations, nevertheless, increased the magnetostriction in Si-Fe alloys and can be expected to do the same here. Of course, the fact that the strips are not perfectly straight and that the strain gauges could not be applied exactly in the rolling direction also caused the magnetostriction readings to vary and, generally, to increase. The variation in composition detected by analysis is not considered a significant cause of the scatter in the readings. If it were, then a definite trend in the magnetostriction versus composition curve would result. This is not to say that such a variation is not present, but its effects here are masked by other variables. Compositional variations within a given sample will be quite important, however. Internal stresses, which increase the magnetostriction, are induced by heterogeneity. Variations of up to 11 pct. Al from one end of a sample to another have been found in the as-sintered condition; but after an additional 12 hour annealing treatment, the alloys should be fairly homogeneous.

It should be noted that the plots of Figure 1 and Figure 2 are for saturation magnetostriction, which occurs around 20,000 Gauss. As our readings were taken at 15,000 Gauss, it is not permissible to say that these figures represent a theoretical value for either a preferred or a
random orientation. It is entirely possible that a strain gauge could have been installed incorrectly or damaged during installation. The only check that can be made on the installed gauge is a resistance measurement of the gauge; but if the resistance is correct, the gauge may still be inaccurate. As all of the installed gauges had the correct resistance, none were rejected, creating the possibility of erroneous results on a few samples.

The fact that the variation in magnetostriction readings of the randomly oriented samples is much less than that of the textured samples suggests that the degree of perfection of the orientation and the domain pattern, which depends upon the orientation, are far more significant in determining the magnetostriction of a sample than are the other variables discussed above. The inclusion content being relatively high in the Goss textured samples will also increase the magnetostriction by creating Neel spikes in the domain patterns, opposing domain wall migration and creating internal stresses in the structure. The complete elimination of oxide inclusions (which in this material should accommodate nearly all inclusions) might not be desirable as the inclusions are probably necessary to permit secondary recrystallization by inhibiting primary grain growth. It is possible, however, that the specimen thickness effect would perform this function, allowing, at least theoretically, the removal of all inclusions.

There is one apparent anomaly in the magnetostriction results of the specimens annealed at high temperatures. The saturation magnetostriction, rises with increased aluminum content (Figure 1). This would suggest that the average magnetostriction value of the 3.50 pct. Al samples should be greater than that of the 2.70 pct. Al samples. This is not the case.
Due to the large variations encountered in the readings, it is felt that if additional strips were tested, the average value of the magnetostriction readings would be directly proportional to the alloying content.

Even with this large spread of values, about 60 pct. of the high temperature samples had magnetostriction values equal to or less than \(5.0 \times 10^{-6}\) in./in., which is as good as that of a grain-oriented Si-Fe material.

Power loss of the samples annealed at 1200°C and 1300°C decreased with increasing aluminum content. This effect had two causes. The first is the increase in resistivity produced by Al. The increased resistivity reduces the eddy current flow, lowering the power loss. The second effect is the preferred orientation of the higher Al alloys, which results in a larger domain size with the 180° domain walls parallel to the rolling direction. The slope of the line of average power loss values is greatest across the transition region from random to preferred orientation (i.e. 0.75 pct. Al to 1.75 pct. Al), suggesting that the preferred orientation is very effective in reducing the power loss. The spread of data is again quite marked and, in this case, the randomly oriented samples have as large a spread as the textured material. A large variation in power loss readings is often found among samples taken from the same piece of strip. It is not unusual, therefore, that by taking readings on individual samples heat treated in different batches such a variation is recorded. The factors which affected the magnetostriction results - compositional variations, inhomogeneity, inclusion content, domain pattern and grain size (all of which are not independent variables) - also affect the power loss readings. As the power loss is also dependent upon the thickness of the samples (equation (8)), the variation in the thickness,
0.011 in. *0.001 in. (approximately 10 pct.), will also scatter the readings. From these results it appears that a power loss of less than 1 watt/lb can be expected in the Goss textured alloys. In fact, if a material with the same grain size and degree of preferred orientation as the 3.25 pct. Si-Fe alloy were produced in Al-Fe, the power loss would be substantially below 1 watt/lb.

The mechanical properties of the textured aluminum alloys are significantly better than the pure iron samples. The addition of 2.70 pct. Al approximately doubles both the yield strength and the ultimate tensile strength of pure iron. The 2.70 pct. Al sample had an average yield strength of 23,200 psi as determined by a 0.5 pct. offset and an ultimate tensile strength of 38,300 psi. However, a Goss textured 3.25 pct. Si alloy had an average yield strength of 48,600 psi and an average ultimate tensile strength of 51,400 psi. These values are approximately double the aluminum alloy values.

Similarly, while the addition of 2.70 pct. Al nearly doubled the Vickers hardness of pure iron from 75.3 to 138, the grain-oriented 3.25 pct. Si alloy is nearly twice as hard, having a Vickers hardness of 252.

While electrical strip is not submitted to severe tensile loading during operation, it is important to be able to shear and stamp electrical sheet without severe edge deformation, which would affect its performance. These characteristics improve with increased strength and hardness. It should be concluded, therefore, that the silicon alloy is better suited for these operations.

In summary, the 2.70 weight per cent Al in Fe, Goss textured alloy produced here has an average linear DC magnetostriction (at 15,000 Gauss) of approximately $7.0 \times 10^{-6}$ in./in. and a 60 cps power loss (at a peak
flux density of 15,000 Gauss) of less than 1 watt/lb. While these values represent an improvement over a non-oriented silicon-iron alloy, they are not as good as the silicon-iron Goss textured strip of the same thickness (0.011 in.). Mechanically the 2.70 weight per cent Al-Fe alloy is inferior to the silicon material.

B. Samples Given a Low Temperature Annealing Treatment

None of the material annealed at 900°C underwent secondary recrystal-
лизация. As no phase change occurred during cooling, the average grain size of the 0.75 pct. Al samples was roughly the same as in the higher Al alloys. The slightly increased grain size of the pure iron samples can be attributed to the lack of oxide inclusions, which retard normal grain growth. Due to the small grain size of this material, power loss values, if taken, would have been significantly above the average values of the Al-Fe material annealed at high temperature. If a purely random orientation existed in this material, then according to Hall, the 1.75 pct. Al alloy should have had zero magnetostriction. The scatter of values obtained at this composition is a result of the primary recrystal-
лизация texture.

The presence of some large grains in the alloy treated at 1100°C implies that secondary recrystallization occurs at this temperature. As the microstructure also contains small grains, 1100°C must be close to the minimum temperature for secondary recrystallization. This is in agreement with the minimum temperature determined for powdered Si-Fe strip. An incomplete secondary recrystallization texture suggests that the inclusion density through the material is not constant. In areas of high inclusion concentration secondary recrystallization (i.e. abnormal grain growth) is inhibited. This inhibition is overcome by using higher
annealing temperatures. It is difficult to interpret the magnetostriction values for these alloys. The X-ray results suggest that the large grains do tend toward (110)[001]. As the grain size of the material under the strain gauge is not known, it can only be assumed that both large and small grains contribute to the magnetostriction readings. In any event, the intergranular activity that accompanies misalignment from (110)[001] will increase the magnetostriction, yielding values somewhat above those of the high temperature material.

Figure 13 illustrates that the power loss values of the 1100°C samples are above the average values of the high temperature material, reflecting the presence of the small grains.

The aluminum-iron alloys annealed at low temperatures (i.e. 900°C and 1100°C) would not be acceptable as electrical strip.
A serviceable, Goss textured, 2.70 weight per cent Al-Fe electrical strip has been produced whose properties, both magnetic and mechanical, are inferior to the standard Goss textured, 3.25 weight per cent Si-Fe material. However, this Al-Fe alloy has lower power loss than pure iron and some non-oriented silicon-iron alloys, although the latter two materials in a non-oriented condition may have lower magnetostriction. In addition, this aluminum alloy has an insulating oxide coat in the as-annealed condition.

The material could be improved magnetically by perfecting the texture, which is of low quality. A more orthodox rolling schedule (i.e. a large reduction of about 75 per cent, then a final reduction of about 50 per cent) should improve the orientation. A lower final annealing temperature of approximately 1150°C should be sufficient for secondary recrystallization. A more uniform inclusion density would be desirable, particularly at this reduced annealing temperature, to insure uniform nucleation of secondary recrystallization grains. As the inclusions are almost exclusively aluminum oxide, produced by the oxidation of aluminum during sintering, a more uniform inclusion density would be achieved by better mixing of the powders.

The extensive heat treatment used demonstrates that a Goss texture will develop. Trial runs would establish the annealing temperature and time at temperature required to produce the desired grain size. It seems
possible from the grain sizes encountered at 1100°C that a larger grain size than occurs in the "Hipersil" material could be achieved. The oxide inclusion content of the material can be minimized by using relatively clean aluminum powder and a hydrogen atmosphere low in \( \text{O}_2 \) and \( \text{H}_2\text{O} \). However, the aluminum powder used in this work was dessicated and a high purity hydrogen employed. Vacuum sintering and/or annealing would probably be necessary to effect a significant reduction of aluminum oxide in the sample. Such a reduction would improve the magnetic properties.
Figure 1: $\lambda_{100}$ for Al-Fe and Si-Fe Single Crystals
Figure 2: $\lambda_p$ for Al-Fe and Si-Fe Randomly Oriented Material
Figure 3: Dewpoint versus Sintering Temperature (Walker and Howard²⁴)
Figure 4: Al-Fe Phase Diagram (Hansen^47)

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Figure 5: Magnetic Circuit for Magnetostriiction Measurements

- Stainless Steel Strip
- Laminated Core
- Test Specimen
- Search Coil
- Coils
- R₁, R₂, R₃, R₄
Figure 6: Electrical Circuit for Magnetostriction Measurements
Figure 7: Electrical Circuit for Temperature Measurements
Figure 9: Theoretical Time-Temperature Plot

\[ \left( \frac{\partial^2 \theta}{\partial t^2} \right)_{t_1} \]

\[ \left( \frac{\partial \theta}{\partial t} \right)_{t_1} \]

\[ \theta(t) \]

Time

Temperature

\( t_0 \)

\( t_1 \)
Figure 10: Power Loss Calibration Curve

Approx. linear to 18.5 (measured) at 4.15 (actual)
Figure 11: DC Magnetostriiction of Strips Annealed at 1200°C
Figure 12: DC Magnetostriiction of Strips Annealed at 1300°C
Figure 13: Power Loss

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Figure 14: DC Magnetostriction of Strips Annealed at 900°C and 1100°C

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Figure 15: Tensile Strengths

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Figure 16: Vickers Hardness
Figure 17: Macrostructure of a Pure Iron Strip Annealed at 1300°C (3X)

Figure 18: Microstructure of a 0.75 wt pct. Al Strip Annealed at 1300°C (200X)
Figure 19: Macrostructure of a 0.75 wt pct. Al Strip Annealed at 1300°C (3X)

Figure 20: Macrostructure of a 2.70 wt pct. Al Strip Annealed at 1300°C (3X)
Figure 21: Microstructure of a 2.70 wt pct. Al Strip Annealed at 1300°C (200X)

Figure 22: Microstructure of a 1.75 wt pct. Al Strip Annealed at 900°C (200X)
Figure 23: Microstructure of a 2.70 wt pct. Al Strip Annealed at 1100°C (100X)

Figure 24: Magnetostriction Apparatus
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APPENDIX A

DERIVATION OF MAGNETOSTRICTION FORMULA

Referring to Figure 5, $R_1$, $R_2$, $R_3$ and $R_4$ are the nominal resistances of the strain gauges. $R_1$ and $R_3$ are the active gauges. The gauge factors for the active gauges are:

$$ GF_1 = \frac{dR_1}{R_1} \quad \text{and} \quad GF_3 = \frac{dR_3}{R_3} $$

The equation for bridge balance is:

$$ R_1 R_3 = R_2 R_4 $$

where

$$ R_4' = \frac{R_4 R_b}{R_4 + R_b} $$

Differentiating equation (20) to obtain the effect of small changes in the active gauge resistances and $R_b$:

$$ R_3 dR_1 + R_1 dR_3 = R_2 dR_4' $$

$$ dR_4' = \frac{(R_4 + R_b) R_4 dR_b - R_4 R_b dR_b}{(R_4 + R_b)^2} $$
\[ dR'_4 = \frac{R_4^2 dR_b}{(R_4 + R_b)^2} \]

Substituting into equation (21);

\[ R_3 dR_1 + R_1 dR_3 = \frac{R_2 R_4^2 dR_b}{(R_4 + R_b)^2} \]  \hspace{1cm} (22)

From equation (19);

\[ dR_1 = R_1 (GF_1) d1/l \]

and

\[ dR_3 = R_3 (GF_3) d1/l \]

Substituting the above in equation (22);

\[ R_3 R_1 (GF_1) d1/l + R_1 R_3 (GF_3) d1/l = \frac{R_2 R_4^2 dR_b}{(R_4 + R_b)^2} \]

Using identical strain gauges;

\[ R_1 = R_2 = R_3 = R_4 \quad \text{and} \quad GF_1 = GF_3 \]
Therefore;

\[ 2(GF_1)R_1^2 \frac{d1/1}{1} = \frac{R_1^3 dR_b}{(R_1 + R_b)^2} \]

Assuming \( R_b \gg R_1; \)

\[ 2(GF_1)R_1^2 \frac{d1/1}{1} = \frac{R_1^3 dR_b}{(R_b)^2} \]

\[ d1/1 = \frac{R_1 dR_b}{2(GF_1)R_b^2} \]  \( \text{(9)} \)
APPENDIX B
AC INDUCTION CALIBRATION

Referring to the output circuit of Figure 7, the voltage across the oscilloscope, $V_o$, is;

$$V_o = \frac{1}{C} \int I \, dt$$  \hspace{1cm} (23)

Assuming that the resistance of the search coil is negligible in comparison to $R$;

$$V_o = \frac{1}{C} \int \frac{V_{in}}{R} \, dt$$

where $V_{in}$ is the voltage induced into the coil.

$$V_o = \frac{1}{RC} \int V_{in} \, dt = \frac{N_2}{RC} \int \frac{d\phi}{dt} \, dt = \frac{N_2 \phi}{RC}$$

where

$\phi = \text{magnetic flux}$

Therefore;

$$\phi = \frac{V_o RC}{N_2}$$  \hspace{1cm} (24)
Dividing both sides of equation (24) by the cross-sectional area of the sample, \( A(\text{cm}^2) \);

\[
\frac{\Phi}{A} = \frac{B}{N_2A} = \frac{\text{RC}}{N_2A} \cdot \frac{V}{\text{volts}}
\]

where

\( B = \) magnetic flux density

\[
\frac{B}{\text{mm}} = \frac{\text{RC}}{N_2A} \cdot \frac{\text{volts}}{\text{mm}} \quad (25)
\]

In the cgs system;

\[
\begin{align*}
1 \text{ farad} & = 10^{-9} \text{ emu} \\
1 \text{ ohm} & = 10^9 \text{ emu} \\
1 \text{ volt} & = 10^8 \text{ emu} \\
1 \text{ Gauss} & = 1 \text{ emu}
\end{align*}
\]

Then equation (25) becomes;

\[
\text{Gauss/mm def} = (\text{CR}/N_2A) \times 10^8 \times \text{volts/mm} \quad (11)
\]
APPENDIX C
POWER LOSS MEASUREMENT

The initial part of the following is a summary of a portion of Westinghouse Research Report 10-0604-1-R4.

The use of equation (16) requires that the initial part of the actual transient heating curve of the specimen be linear. The power, $P$, generated in a specimen subjected to an alternating field of angular velocity, $\omega$, heats the specimen according to:

$$v^20 + \frac{P}{K} = \frac{cp}{K} \frac{\partial \theta}{\partial t}$$

(26)

where

$$P = P_0 \left[ 1 + \sum_{i=1}^{\infty} a_i \sin(\omega t + \phi_i) \right]$$

The values of $a_i$ and $\phi_i$ depend upon the waveform of the flux density, which is a function of many other factors. Although $P$ varies with time, compared to the time of measurement and the response of the circuit of Figure 6, it can be considered equal to its average value.

The purely theoretical heating curve of the specimen must be calculated from equation (26) for a laminated specimen having a rectangular cross-section. The individual considerations of specimen size and insulation makes this extremely difficult.

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Instead, the calculations are made for a sphere whose volume is approximately that of the rectangular section and using experimental values of power loss (0.34 watts/lb) and final temperature rise (0.793°C) from the rectangular specimen (1 in. x 1 in. x 0.5 in.). It was determined that the transient heating curve is linear to within 5 pct. of the initial value;

\[
\frac{\partial \theta}{\partial t} t_o = \frac{p}{c_p}
\]

for at least 12 seconds after application of the field. For power generations greater than 0.32 watts/lb the slope is constant for a greater period.

As our samples had a larger surface to volume ratio than the rectangular specimen mentioned above, the heat lost to the atmosphere will tend to decrease the linear portion of the transient heating curve. However, as the power loss is greater than 0.32 watts/lb, the linear portion of the heating curve should not be too severely affected. Indeed, the recorded heating curves were approximately linear for at least 10 seconds.

The inability to reproduce results using equation (17) suggests that, beyond the linear portion of the heating curve, large, erratic heat conduction to the air makes the time-temperature curve inconsistent, particularly during cooling.
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