Microstructural characterization of ferromagnetic materials using magnetic NDE techniques

Rajiv Ranjan
Iowa State University

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MICROSTRUCTURAL CHARACTERIZATION OF FERROMAGNETIC MATERIALS USING MAGNETIC NDE TECHNIQUES

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Microstructural characterization of ferromagnetic materials using magnetic NDE techniques

by

Rajiv Ranjan

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DEDICATION

To my parents,
Sudha and Rameshwar Yadav
GENERAL INTRODUCTION—

The mechanical properties of ferromagnetic materials depend strongly on their microstructural features such as grain size, dislocation density and type, size and density of precipitates [1]. For this reason, the steel industries, etc., spend a great deal of effort in determining and ensuring the desired microstructure. In practice, the microstructure is usually determined by optical or electron microscopy where samples are cut from the actual product and are observed after special sample preparation. Conceivably, these methods are time consuming and a hundred percent inspection is probably impossible considering the fact that the delay in preparing the samples would affect the product's net output. Thus, there is an interest to develop nondestructive on-line techniques capable of rapid evaluation of microstructures. It has been recently recognized [2] that magnetic NDE methods have a great advantage over other NDE (e.g., ultrasonic and x-ray) methods because of their high sensitivity to a change of several microstructural features. In addition, the magnetic NDE techniques, as will be discussed later, provide ample micromagnetic information [3–5]. This information can provide a link between magnetic domain models [6, 7] and the magnetization curve [8, 9].

In the present study, specific magnetic NDE methods, such as the magnetic Barkhausen noise (MB), the acoustic Barkhausen noise (AB) and the hysteresis (B-H) curves, have been used to study the microstructures of polycrystalline nickel and steels.
Explanation of the Dissertation Format

This dissertation begins with a general review of magnetic NDE methods, followed by a detailed description of experimental techniques. Section I contains a report on a study of the effect of dislocations on magnetic properties of polycrystalline nickel. This report is a compilation of two papers, one of which has been published in Review of Progress in Quantitative NDE and the other one has been accepted by the Journal of Applied Physics.

Section II deals with grain size measurements using magnetic and acoustic Barkhausen noises. This report has been accepted by the Journal of Applied Physics.

Section III contains a report on a study of the effect of carbide precipitation and hardness of steels using acoustic and magnetic Barkhausen noises. This report has been accepted by Acta Metallurgica.

The Appendix gives the design details of magnetic Barkhausen sensors. This is part of a paper which received a student paper award at the SIMS (Symposium for Innovation in Measurement Sciences) 1986 meeting of the Instruments Society of America.
GENERAL REVIEW OF MAGNETIC NDE TECHNIQUES

In a ferromagnetic material, which is in the demagnetized state, the magnetic domains are oriented such that the net magnetic energy (which includes the magnetocrystalline anisotropy energy, the magnetostatic energy, the magnetoelastic energy and the domain wall energy) is minimized [10]. As a magnetic field is applied to a ferromagnetic polycrystalline material, some of the domains which are oriented favorably with respect to the applied magnetic field direction grow at the expense of the others by translation of domain walls and rotation of the spins. As a result of this, the net magnetization is increased. The domain walls carry a stress field (due to the exchange energy, magnetocrystalline anisotropy, and magnetoelastic coupling) which interacts with the strain fields of the different structural defects mentioned earlier. In other words, it is the magnetostrictive coupling (and also the magnetostatic coupling if the defect is larger than the thickness of the domain wall) between the structural defects and the domain walls that affect the motion of the domain walls and, therefore, the ferromagnetic properties [7]. It is this interaction between the defects and domain walls which is responsible for the irreversible motion of the domain walls during magnetization. This is also the origin of magnetic hysteresis and generation of magnetic and acoustic Barkhausen noises.

In 1919, Barkhausen [11] found that magnetization changes in a piece of iron induced a voltage in a pick-up coil which produced "noise"
in an earphone, which was called "Barkhausen noise". Incidentally, this experiment gave the first evidence of the presence of magnetic domains in a ferromagnetic material. In 1975, Lord [12] reported on an observation of acoustic emission generated during magnetization of nickel. Its origin was attributed to changes in the elastic energy associated with the Barkhausen effect. Due to its nature of generation, this phenomenon is called acoustic Barkhausen noise (AB) or magnetomechanical acoustic emission. With the same logic, the above-mentioned "Barkhausen noise" is now called the magnetic Barkhausen noise (MB) to clearly distinguish it from the acoustic Barkhausen noise.

MB is caused by abrupt changes in magnetization with changing magnetic field, principally by the motion of 180° domain walls. Simultaneously, some non-180° domain wall motion will have to occur due to the presence of closure domains. However, contributions of the non-180° domain walls to MB are smaller than those of the 180° domain walls for two reasons: (i) the average velocity of 180° walls is larger than that of the non-180° walls, as the non-180° walls have to accommodate the change in magnetostrictive strain during their motion, and (ii) the volume swept out by 180° walls is larger than that by non-180° walls [3]. AB, on the other hand, is stress waves generated by the local changes of the magnetostrictive strains associated with irreversible translation of non-180° domain walls. This was shown by Kusangi et al. [13] who showed very little AB activity in cobalt as compared to nickel and steel since cobalt
contains fewer non-180° domain walls due to its high crystalline anisotropy.

MB, AB and the magnetic properties derived from magnetic hysteresis curves have been separately used for microstructural characterization of ferromagnetic materials [14-20]. The simultaneous observation of MB, AB and the magnetic hysteresis curves, however, should have a number of advantages. It should provide a means of measuring the interaction between various types of domain walls and the different microstructural features such as dislocations, grain boundaries and precipitates. At the same time, it also should enable one to select the most sensitive technique for a particular application since the sensitivity of the signals varies from one microstructural feature to the other [4].
EXPERIMENTAL TECHNIQUES

The block diagram of the experimental setup is shown in Figure 1. Long cylindrical samples were placed along the axis of a large solenoid. The magnetic field was controlled from an IBM PC personal computer via a Kepco SN-121 digital-to-analog converter and a Kepco BPO 72-5M programmable power supply. The MB signals were detected by having a search coil surrounding the sample undergoing magnetization. Discontinuous motion of 180° domain walls results in a small local induction change, consequently generating a field which propagates within the material before arriving at the surface and thereby, inducing a time varying voltage in the search coil, as depicted in Figure 2(a). The attenuation (eddy current loss) experienced by the propagating field will depend on the relative location of the Barkhausen discontinuity w.r.t and the search coil. Therefore, the induced voltage in the search coil is given by

\[ V(t) = G(t) \times M(t) \times S(t) \]  

(1)

where \( G(t) \), \( M(t) \) and \( S(t) \) are explained in Figure 2(a). In a first order approximation, both \( G(t) \) and \( M(t) \) can be assumed to be constant for a constant magnetizing frequency. \( S(t) \), the source response function, depends on the rate of induction change (dB/dt) due to the sudden motion of magnetic domain walls. As a result, the induced MB signal will contain pulses of varying amplitude and frequency, and these variations will depend on the depth at which the event occurred,
Figure 1. Schematic diagram of the experimental set-up
Figure 2. A schematic illustration of (a) magnetic Barkhausen noise and (b) acoustic Barkhausen noise.
as well as on the detailed features of the signal. Pulse height and frequency spectrum analysis [21, 22] have been used to aid in the separation of the signal features. The detailed design considerations of a MB sensor are explained in the Appendix.

AB, as stated earlier, is the stress waves generated by changes in the magnetostrictive strains during non-180° wall translation. The stress waves propagate spherically from the source and cause transient elastic displacements at the surface of the specimen, as shown in Figure 2(b). These are detected by a piezoelectric transducer cemented to the sample surface. The signals may again be analyzed for spectral content and pulse height distribution [23, 24]. The hysteresis curves, i.e., the variation of magnetic induction, B, with field H, were measured using a modified Walker MH-10 hysteresisgraph. The output from a second search coil was connected to a MF-3A integrating voltmeter to determine B, while the output from the Hall probe was connected to a MG-3A gaussmeter to determine H.
SECTION I. A STUDY ON THE EFFECT OF DISLOCATIONS ON THE MAGNETIC PROPERTIES OF NICKEL USING MAGNETIC NDE METHODS
INTRODUCTION

Magnetic domain walls carry a stress field (due to the exchange energy, the magnetocrystalline energy and magnetoelastic coupling) which interacts with strain fields of dislocations. It is this magnetostrictive coupling between dislocations and domain walls that affects the movement of the domain walls and therefore, the magnetic properties of plastically deformed ferromagnetic materials \[1\]. Dislocations not only interact with the domain walls, but also change the size and shape of magnetic domains as has been observed by the magneto-optical Kerr-effect \[2, 3\]. These changes are caused by the presence of the internal stresses surrounding the dislocations. The effect of these stresses on the fluctuations of the spontaneous magnetization direction can be obtained from the micromagnetic equations using the minimization of the total Gibbs free energy \[4\]. The fluctuations of the spontaneous magnetization affect the domain structure and thus, the dynamics of the domain walls generating MB and AB during magnetization.

Rudyak \[5\] has suggested a direct connection between the residual magnetization and the Barkhausen effect. His simultaneous measurements of the residual magnetization and the number of Barkhausen jumps in nickel show that the two are linearly related to one another and are strongly influenced by the defect structure. Kondorskii's theory \[6\] of the hysteresis suggests that it is mainly caused by the delay in the displacement of domain walls and the delay in the growth of domain
nuclei as H is applied. Kameda and Ranjan [7] have evidence that the
aforementioned parameters directly affect AB and MB signals. The
present work was carried out with the objective to understand the
interaction of different types of domain walls, viz., 180° and
non-180° walls, with dislocations.

This report presents observations on the variations of magnetic
and acoustic Barkhausen noises and the magnetization of plastically
deformed pure nickel during different stages of annealing in which a
number of microstructural features, such as dislocation density and
arrangement, grain size and distribution, change in a well-known way.
EXPERIMENTAL PROCEDURE AND RESULTS

Pure nickel samples were swaged (~70% R.A.) to a final diameter of 1/2 inch and a length-to-diameter ratio equal to 12.0. These samples were then annealed at different temperatures for the same lengths of time (isochronal anneal). The magnetization (B-H) curves, and the magnetic and acoustic Barkhausen signals were recorded for a 0.1 Hz magnetic field ($H_{\text{max}} \approx \pm 100$ Oe) with a triangular wave form. In these experiments, the samples were initially fully demagnetized to achieve an equilibrium magnetic condition at the beginning of each experiment. Failure to do this altered the results.

The microstructures of the annealed nickel samples are shown in Figure 1 and their hardness values are shown in Figure 2. Examples of a magnetization curve, and the magnetic and acoustic Barkhausen signals are shown in Figure 3, as observed on the plastically deformed sample (solid line) and the sample annealed at 350°C for 3 hrs. (dotted line). The magnetic and acoustic Barkhausen noises obtained are shown only for those portions of the magnetization curve denoted by arrows. Noticeable is the pronounced multiple peak structure in both the acoustic and magnetic Barkhausen noises.

The hysteresis curves for the samples are shown in Figure 4. The hysteresis loss, that is the area under the B-H curves shown in Figure 4, has been plotted in Figure 5 as a function of annealing temperature. The total number of counts of MB and AB signals over a predetermined threshold is shown in Figures 6 and 7, respectively.
Figure 1. Optical micrographs of the nickel samples (samples were annealed for 3 hrs. at the indicated temperatures)
Annealing time = 3 hrs.

Figure 2. Hardness (DPH) of the annealed nickel samples as a function of annealing temperature.
Figure 3. Magnetic hysteresis curves (top) and the accompanying count rate plots of magnetic (middle) and acoustic Barkhausen noise (bottom) for deformed (−) and 350°C annealed (---) nickel samples.
Figure 4. Magnetic hysteresis curves of nickel samples
Annealing time = 3 hrs.
Frequency of the Magnetic Field
\( \approx 0.1 \text{ Hz} \)
\( \Delta H \approx \pm 110 \text{ Oe} \)

Figure 5. Hysteresis loss of nickel samples as a function of annealing temperature
Figure 6. Total number of counts of MB signals as a function of annealing temperature.

Annealing time = 3 hrs.

Frequency of the Magnetic Field
≈ 0.1 Hz
ΔH ≈ ±110 Oe

Annealing Temp. (°C)

Total Number of Counts (X 10^3)

as deformed
Figure 7. Total number of counts of AB signals as a function of annealing temperature.
DISCUSSION

Microstructural Aspect

During isochronal or isothermal annealing, the microstructure of the material undergoes changes consisting of recovery, followed by recrystallization and finally, grain growth [8]. During recovery, some of the dislocations annihilate each other and, therefore, the dislocation density of the material decreases by a factor of about ten. Moreover, the dislocations rearrange into a low energy honeycomb type configuration (polygonization). Sample 2 in Figure 1 has undergone such recovery. As the dislocation density stays basically constant during recovery [9], in contrast to recrystallization, the hardness of the material remains almost constant [8].

During recrystallization, new strain free grains nucleate and grow into the cold worked matrix [8] with increasing time or temperature. In Figure 1, samples 3 and 4 have partially undergone recrystallization with sample 4 being more recrystallized than sample 3, due to the higher annealing temperature. Sample 5 is almost fully recrystallized. Since the new strain free grains grow and consume the cold worked matrix, the dislocation density of the material drops by a factor of about $10^4$ at the completion of recrystallization [10]. As a result, the major drop in hardness takes place during recrystallization, as can be seen in Figures 2.

During grain growth, some thermodynamically favorable grains grow at the expense of others [8]. As may be seen in Figure 1, some grains
of sample 5 and most of the grains of sample 6 have undergone grain growth. The hardness drop during grain growth is relatively small, as shown in Figure 2.

Comparing Figures 2 and 5, it can be seen that the hysteresis loss (area under the B-H curve) and the hardness both decrease with increasing annealing temperature. It thus appears that a measurement of the hysteresis loss in nickel may be useful to predict hardness.

Examination of the dashed hysteresis curve in Figure 3 shows a "wasp-waist" shape that develops after a 350°C anneal. This suggests that two independent processes may occur during the magnetization, $H < H_c$, and the second active in the later stages of magnetization, $H > H_c$. It is also interesting to note that when this "wasp-waist" shape develops, that portion of the acoustic Barkhausen noise occurring at $H < H_c$ drops substantially (see Figure 3). Noting that both of these effects occur in the recrystallization region, it is tempting to speculate that the effects are related to the formation of strain free, recrystallized grains, having different magnetic properties than the cold worked ones. The "wasp-waist" hysteresis curve would then be a consequence of a superposition of two regular hysteresis curves, one for the strain free grains and the other for the cold worked unrecrystallized grains. As schematically shown in Figure 8(a), the magnetization path with increasing field can, therefore, be envisioned to consist of two different magnetization paths, curve I (---) and curve II (--x--). Magnetization curve I is assumed to be due to the strain free recrystallized grains (with low dislocation density) and curve II is
Figure 8. Magnetic hysteresis, magnetic and acoustic Barkhausen noise records of nickel annealed at 350°C. The solid curves (-) represent an idealization based on many experiments. The dashed curves I and II represent idealized components for annealed and deformed nickel.
due to the cold worked grains (with high dislocation density). If this speculation is correct, the acoustic and magnetic Barkhausen signals (observed on material with "wasp-waist" shaped magnetization curves) should show the characteristics found on the fully recrystallized and the unrecrystallized material. The acoustic and magnetic Barkhausen signals shown in Figures 8(b) and 8(c) can be interpreted in a similar fashion. Here, the solid curves represent an idealization, based on many runs, of the data obtained after the 350°C anneal. The dashed curves I and II represent hypothetical components, similar to those observed individually on the deformed and annealed specimens, but simplified and shifted in H to correspond to the two steep portions of the hysteresis curve. Thus, the triply peaked magnetic Barkhausen structure can be interpreted as the superposition of simpler processes in the two classes of grains. Moreover, the major loops of the magnetic hysteresis curve at low field amplitudes, shown in Figure 8(a) (.-.-.-.), resemble the one of fully annealed (500°C and 600°C) samples. This, once again, supports our speculation that the strain free recrystallized grains are magnetized during the early part of magnetization, i.e., $H<H_c$ and then, the rest of the cold worked grains are magnetized.

Figures 2, 5 and 7 show that the total number of counts of AB signals and the hysteresis loss follow the same trend as hardness. However, the total number of MB signal counts in Figure 6 does not follow the hardness curve, as indicated by the slope at low temperature and the high temperature maximum. Since the AB signal is mainly due to
the sudden local change in magnetostrictive strain associated with the irreversible translation of non-180° domain walls, the results imply that the non-180° domain walls interact strongly with dislocations. On the other hand, 180° domain walls do not seem to show such a strong interaction with dislocations. Theoretical calculations [11] of the magnetoelastic interaction between dislocation domain walls indicate a stronger interaction with non-180° domain walls than with 180° domain walls, consistent with our experimental results.

Figure 6 shows a "valley" in the 300°C to 500°C annealing temperature region which is the recrystallization region. This has also been observed in iron [12]. Two competitive processes occur during recrystallization, namely, a drop in dislocation density and a nucleation of small recrystallized grains. As the dislocation density decreases, the number of pinning points for 180° domain walls decreases and therefore, the MB signal decreases. With the nucleation of small recrystallized grains, the average grain size decreases and the density of 180° domain walls increases [13]. It has also been observed [14] that the MB signals increase with decreasing grain size in nickel. Thus, the two above-mentioned processes have an opposite effect on the MB signal and thereby result in a "valley" in the recrystallization region, as shown in Figure 6. However, such a valley was not observed in the AB signals as the grain size change does not affect AB signals as strongly as it does MB signals [14]. The "valley" in the MB signal of nickel might be used for nondestructive estimation of the extent of
recrystallization in ferromagnetic materials by attaching the MB sensor directly to the material undergoing annealing and running real time experiments.
CONCLUSIONS

Results show that AB signals and hysteresis loss in nickel follow the same trend during annealing as hardness. Thus, they can be used to nondestructively measure the extent of deformation in nickel. Results of MB signals show a "valley" in the recrystallization region. This is due to a combined effect of change in grain size and dislocation density during recrystallization. The difference in the sensitivities of AB and MB signals to change in dislocation density is due to the difference in the interaction of 180° and non-180° walls with dislocations. Results support the theoretical calculations [11] which show that non-180° domain walls interact more strongly with dislocations than 180° domain walls.
REFERENCES


SECTION II. GRAIN SIZE MEASUREMENT USING MAGNETIC AND ACOUSTIC BARKHAUSEN NOISES
INTRODUCTION

The relationship between grain size and magnetic properties has been an area of great interest for a long time. It has generally been observed that an increase in grain size causes a decrease in the hysteresis losses [1]. From the metallurgical point of view, the grain size is affected by the amount of impurities and second-phase particles during recrystallization, as well as grain growth. Impurities and second-phase particles act as nucleating sites and thereby, make it easier for new grains to form during recrystallization. On the other hand, they inhibit the grain boundary migration by pinning the grain boundaries, slowing down grain growth. Therefore, the higher the density of impurities and second-phase particles, the smaller the grain size will be. From the magnetic point of view, impurities and second-phase particles, as well as grain boundaries, tend to pin domain walls and act as nucleation sites for closure domains [2]. Thus, to separate the individual contributions of such defects to the magnetic properties by a changing microstructure is not a simple matter. Certainly, individual attempts to provide correlations have been made such as by Adler and Pfeiffer [3], e.g., who have suggested that the coercive force in 47.5% NiFe is inversely proportional to the grain diameter.

In general, grain boundary effects on magnetic properties are explained in two ways. Theoretical calculations [4] have been made based on the Globus model [5] of domain wall pinning at grain boundaries.
to study the effect of grain size distribution on the magnetization curve. The model assumes spherical grains which consist of two domains separated by a 180° domain wall. When a low magnetic field is applied, the domain walls remain pinned at the grain boundary and the centers of the walls will bow out. As the applied field is further increased, the walls get unpinned and jump within the grain until they are repinned. Since this is a discontinuous process and involves only 180° domain walls, it implies that a change in grain size should only affect the MB signals and not the AB signals.

Goodenough [2] used a different model to calculate the critical magnetic field required for nucleation of magnetic domains at grain boundaries. At the grain boundaries, so-called "magnetic free poles" tend to be generated due to the change in crystallographic orientation across the grain boundary. This results in the generation of closure domains (reverse spikes) at the grain boundary. As depicted in Figure 1, the grain boundary surface pole density is \( W^* = I_S (\cos \theta_1 - \cos \theta_2) \), where \( \theta_1 \) and \( \theta_2 \) are the angles between the spontaneous magnetization vector \( I_S \) on either side of the grain boundary and its normal. The magnetic energy associated with the "magnetic free poles" is reduced by the formation of closure domains and a redistribution of the remaining poles [6]. At the knee of the hysteresis curve, these closure domains act as the nucleation sites for both 180°, as well as non-180° domain walls. Therefore, one would expect that both MB and AB will change with grain size.
Figure 1. Depiction of free pole distribution at grain boundary
Attempts have been made to observe the effects of free surface poles on the closure domain structure [7]. On the basis of these observations, predictions were made regarding the closure domains inside the material, i.e., at the grain boundary. Craik and McIntyre [8] followed this concept to calculate the magnetostatic energy. However, the utility of this approach is limited since a free sample surface is not representative of the grain boundary surface due to several reasons:

(i) In commercial materials, the grain boundary composition and, therefore, the magnetic properties of the grain boundary can be very much different from the bulk. This is mainly due to grain boundary segregation effects as the grain boundary acts as a sink for impurities and defects.

(ii) The state of stress inside the material is triaxial, whereas on the free surface, it is biaxial (plane stress). Since the domain structure depends on the state of stress, it can be very different under these conditions.

Thus, the domain structure at the grain boundary could be widely different from that near the free surface. The simultaneous detection of Barkhausen noises discussed in the present report, therefore, can be useful as an important nondestructive tool to study the domain structure within the bulk material.

Otala and Säynäjäkangas [9], Säynäjäkangas and Otala [10], Säynäjäkangas [11] and Titto et al. [12] have studied extensively the effect of grain size on MB noise in steel. Their pulse-height distribution results of the MB noise show that the number of large
amplitude pulses increases with increasing grain size. Their results also show that the total number of MB noise pulses increases with increasing grain size in commercial steel [12]. In subsequent work [13, 14], the total number of the MB noises, as well as AB pulses, was reported to increase with increasing grain size in decarburized steel in agreement with references [9-12]. This indicates that the Barkhausen activity of both 180° domain walls and non-180° domain walls (90° in steel) increases in general with increasing grain size in steels.

This report presents results of the grain size effect on the MB and AB noises in nickel, which are compared with the aforementioned results in steel. Finally, micromagnetic mechanisms for the MB and AB noises will be discussed.
EXPERIMENTAL PROCEDURE AND RESULTS

An as-cast nickel rod with the composition given in Table 1 was swaged to a final diameter of 1/2 inch after a series of interrupted annealing treatments. Samples of length-to-diameter ratio of 12.0 were cut and annealed under ultrahigh vacuum at different temperatures for the same length of time. The heat treatment, grain size and hardness of the samples are given in Table 2.

The composition of three groups of decarburized steel specimens are shown in Table 3. These samples were cut from sheets which had undergone the following production sequence; hot rolling, cold rolling, annealing, temper rolling and decarburization annealing. In all cases, the carbon content was less than 0.005 wt%.

MB signals were obtained using a magnetic field cycled at a frequency of 0.05 Hz (triangular) with a maximum amplitude of 200 Oe. The MB count rates, as a function of H, for decarburized steels and nickel are shown in Figures 2 and 3, respectively.

AB signals for nickel and decarburized steel samples were obtained at frequencies of 0.05 Hz and 0.1 Hz (both triangular), respectively.

<table>
<thead>
<tr>
<th>Table 1. Composition of nickel samples</th>
</tr>
</thead>
<tbody>
<tr>
<td>C (at. ppm)</td>
</tr>
<tr>
<td>-------------</td>
</tr>
<tr>
<td>42</td>
</tr>
</tbody>
</table>
Table 2. Heat treatment, grain size and hardness of the samples

<table>
<thead>
<tr>
<th>Sample No.</th>
<th>Heat treatment (after swaging)</th>
<th>Average grain dia. (µm)</th>
<th>Matrix hardness (DPH)</th>
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</thead>
<tbody>
<tr>
<td>1</td>
<td>Annealed at 500°C for 3 hrs.</td>
<td>18</td>
<td>74</td>
</tr>
<tr>
<td>2</td>
<td>Annealed at 600°C for 3 hrs.</td>
<td>51</td>
<td>67</td>
</tr>
<tr>
<td>3</td>
<td>Annealed at 700°C for 3 hrs.</td>
<td>118</td>
<td>64</td>
</tr>
<tr>
<td>4</td>
<td>Annealed at 800°C for 3 hrs.</td>
<td>238</td>
<td>65</td>
</tr>
</tbody>
</table>

Table 3. Composition of three groups of decarburized steel specimens

<table>
<thead>
<tr>
<th></th>
<th>G5</th>
<th>G6</th>
<th>G7</th>
</tr>
</thead>
<tbody>
<tr>
<td>C (wt.%)</td>
<td>&lt; 0.005</td>
<td>&lt; 0.005</td>
<td>&lt; 0.005</td>
</tr>
<tr>
<td>Mn (wt.%)</td>
<td>0.62</td>
<td>0.62</td>
<td>0.55</td>
</tr>
<tr>
<td>P (wt.%)</td>
<td>0.13</td>
<td>0.09</td>
<td>0.017</td>
</tr>
<tr>
<td>Si (wt.%)</td>
<td>0.06</td>
<td>0.20</td>
<td>0.95</td>
</tr>
<tr>
<td>S (wt.%)</td>
<td>0.02</td>
<td>0.015</td>
<td>0.012</td>
</tr>
<tr>
<td>Al (wt.%)</td>
<td>--</td>
<td>0.22</td>
<td>0.22</td>
</tr>
<tr>
<td>Grain size (µm)</td>
<td>72</td>
<td>87</td>
<td>115</td>
</tr>
<tr>
<td>Hardness (DPH)</td>
<td>$172 \pm 5$</td>
<td>$175 \pm 2$</td>
<td>$176 \pm 1$</td>
</tr>
</tbody>
</table>
Magnetic Barkhausen Signal

Freq. of the magnetic field = 0.05 Hz

Signal A is for increasing field.

Signal B is for decreasing field

H(oe)

Figure 2. MB signal count rate of decarburized steel as a function of the field
Figure 3. MB signal count rate of nickel as a function of the field.
For decarburized steels, the higher frequency was required in order to obtain a good signal to noise ratio. The AB count rates, as a function of H, for decarburized steels and nickel are shown in Figures 4 and 5, respectively. Note that the count rates in Figures 2-5 are averages of 25 sweeps. For comparison, the magnetic hysteresis curves of these nickel samples are shown in Figure 6.

An optical micrograph of one of the decarburized samples is shown in Figure 7(a) and of a nickel sample in Figure 7(b). As may be seen, the decarburized sample contains some precipitates which are mainly MnS. The density of precipitates was lowest in group 7 samples and highest in group 5 samples, as may be expected from their compositions (Table 3). As can be seen, the decarburized steels had a relatively high phosphorous concentration which we assume to segregate at the grain boundaries. This segregation would be highest in the group 5 samples and lowest in group 7 samples. In contrast, nickel does not show any precipitates, as expected from the low impurity content.
Acoustic Barkhausen Signal

Freq. of the magnetic field = 0.1 Hz

Signal A is for increasing field

Signal B is for decreasing field

Figure 4. AB signal count rate of decarburized steel as a function of the field
Acoustic Barkhausen Noise:

\[ \dot{N}(x \times 10^4 \text{ Sec}^{-1}) \]

Grain Size = 18 \( \mu \text{m} \)

Grain Size = 51 \( \mu \text{m} \)

Grain Size = 118 \( \mu \text{m} \)

Grain Size = 238 \( \mu \text{m} \)

Figure 5. AB signal count rate of nickel as a function of the field.
Figure 6. Hysteresis curves of samples with different grain sizes
Figure 7. Micrographs of (a) decarburized steel and (b) nickel.
DISCUSSION

Nickel Samples

Results in Figures 3 and 5 indicate that the MB, as well as AB signals in nickel are, to a large degree, sensitive to grain size. Noticeable is the double peak structure observed in each case. Unaffected by the grain size is the smaller peak in the MB signals. The results are summarized in Figure 8, where the ratios of the heights of the second peak to the first one are plotted as a function of grain size. The figure shows that for MB noise, only the second peak drops with increasing grain size. For AB noise, however, both the peaks drop with increasing grain size such that their peak height ratios remain constant. As Table 4 indicates, the separation in the external fields required to produce the two peaks increases with increasing grain size for both MB and AB signals.

In the following, we will be comparing the MB and AB signals of nickel in Figures 3 and 5, respectively, with the corresponding hysteresis curves in Figure 6. It will be argued that the origin of the first peak should be attributed to nucleation of domain walls and that the second peak should be due to the irreversible translation and annihilation of domain walls.

For small-grained materials, the nucleation rate of domains should be higher due to a larger density of available nucleation sites, i.e., grain boundaries. Moreover, small grains would also increase the number of obstacles which domain walls encounter during translation.
Figure 8. Plot of the peak height ratios as a function of grain size.
Table 4. Positions of peaks in MB noise and AB noise in Ni

<table>
<thead>
<tr>
<th>Sample No. with grain size</th>
<th>Position of the first peak ($H_1$)</th>
<th>Position of the second peak ($H_2$)</th>
<th>$\Delta H = H_1 - H_2$</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Magnetic Barkhausen Noise</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>1 18 $\mu m$</td>
<td>-2 0e</td>
<td>+12 0e</td>
<td>14 0e</td>
</tr>
<tr>
<td>2 51 $\mu m$</td>
<td>-3.5 0e</td>
<td>+11.5 0e</td>
<td>15 0e</td>
</tr>
<tr>
<td>3 118 $\mu m$</td>
<td>-7 0e</td>
<td>+16 0e</td>
<td>23 0e</td>
</tr>
<tr>
<td>4 238 $\mu m$</td>
<td>-8.5 0e</td>
<td>+11 0e</td>
<td>19.5 0e</td>
</tr>
<tr>
<td>Acoustic Barkhausen Noise</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>1 18 $\mu m$</td>
<td>0 0e</td>
<td>10 0e</td>
<td>10</td>
</tr>
<tr>
<td>2 51 $\mu m$</td>
<td>-3 0e</td>
<td>10.5 0e</td>
<td>13.5 0e</td>
</tr>
<tr>
<td>3 118 $\mu m$</td>
<td>-5 0e</td>
<td>11.5 0e</td>
<td>16.5 0e</td>
</tr>
<tr>
<td>4 238 $\mu m$</td>
<td>-4.5 0e</td>
<td>11 0e</td>
<td>15.5 0e</td>
</tr>
</tbody>
</table>
Therefore, large Barkhausen signals in small-grained materials would be expected due to the above mechanism. This is consistent with the observed trends in Figures 3 and 5.

The lack of MB and AB signals at field $H = H_c$, in Figures 3 and 5, respectively, is probably due to the presence of reversible domain wall translation as the dominant process for magnetization change. It has been observed in steel [9] that the size of domains increases with grain size. We assume that this domain size – grain size correlation is also true for nickel. As shown by Khan et al. [15], reversible domain wall motion (bowing out) will be enhanced as the domain size and thus, the grain size increase. Thus, the gap between the two peaks which is a representation of the extent of reversible translation during magnetization could be expected to increase with increasing grain size.

Comparing the MB and AB signals in fully annealed nickel, shown in Figures 3 and 5, indicates that 90° domain walls have similar reversible properties as the 180° domain walls. The results indicate that during magnetization of fully annealed nickel, both 180° and non-180° domain walls move together, as was also observed for decarburized steel [13, 14]. This similarity ends, however, as soon as there are other defects, such as dislocations, present. As shown in our earlier studies [16, 17], non-180° domain walls in deformed nickel interact stronger with individual dislocations than 180° domain walls. Thus, as a result of deformation, reversible non-180° domain wall motion decreases and the gap in the AB disappears, unlike in the MB signal.
Comparison of MB and AB Signals in Nickel and Decarburized Steels

As shown in Figures 7(a) and 7(b), the decarburized steels contain both precipitates and grain boundaries unlike the nickel samples with only grain boundaries. The micrographs and hardness of decarburized steel (Table 3) and nickel samples (Table 1) indicate that all the samples were fully annealed. Therefore, it is quite reasonable to assume that the AB and MB signals were equally affected by dislocations and the difference in these signals from one sample to the other was independent of dislocation effects. Thus, the change in MB and AB signals in nickel samples can be attributed to the change in grain size. On the other hand, the change in MB and AB signals in decarburized steel samples is due to a variety of metallurgical variables, namely, in grain size, phosphorous segregation at grain boundaries and the density of MnS precipitates.

The total number of counts of AB signals for decarburized steel and nickel is plotted in Figures 9 and 10, respectively.

The AB is generated due to the sudden change in the localized magnetostriction when domain walls pinned at defects are released. Since the change in the magnetostrictive strain is caused mainly by the translation of non-180° type domain walls, AB is mostly caused by unpinning of non-180° domain walls during magnetization. Assuming that for a low magnetization frequency the individual AB pulses are independent of each other and of the same amplitude, the total number of counts of AB is proportional to the critical velocity and density
Figure 9. AB signal of decarburized steel as a function of grain size
Figure 10. AB signal of nickel as a function of $1/d$, where $d$ is the grain diameter.
of domain walls, as well as the magnetostrictive strain induced by the motion of non-180° domain walls. Thus, the AB signal is given by

$$N_{AB} = c \int \rho_{n-180^\circ}(H) \cdot \bar{v}_{n-180^\circ}(H) \varepsilon_m dH$$  \hspace{1cm} (1)$$

where $c$ is a constant, $\rho_{n-180^\circ}(H)$ is the density (Bloch wall area per unit domain) of non-180° domain walls at a field $H$ and $\bar{v}_{n-180^\circ}(H)$ is the average critical velocity at the center of a bulged non-180° domain wall when it is released from the pinning points. If $\bar{S}$ is the average displacement at the center of a bulged domain wall, the average critical velocity, $\bar{V}(H)$, is proportional to $d\bar{S}/dH$. $\varepsilon_m$ is the average magnetostrictive strain change due to a unit displacement per unit area of domain walls.

The total number of counts of MB signals for decarburized steel and nickel is plotted in Figures 11 and 12, respectively.

The MB signal is generated due to a sudden change in the magnetic flux when 180° domain walls are released from pinning sites [1]. In similarity to Eq. (1), the MB signal can, therefore, be written as

$$N_{MB} = c' \int \rho_{180^\circ}(H) \bar{v}_{180^\circ}(H) \bar{AB} dH$$  \hspace{1cm} (2)$$

where $c'$ is a constant which depends on the time constant of the pickup coil, the permeability and conductivity of the sample; $\rho_{180^\circ}(H)$ is the density of 180°-domain walls at a field $H$; $\bar{v}_{180^\circ}(H)$ is the average critical velocity of a 180° domain wall when it is released from
Figure 11. MB signal of decarburized steel as a function of grain diameter.
Figure 12. MB signal of nickel as a function of $1/d$, where $d$ is the grain diameter.
pinning sites; and \( \overline{\Delta B} \) is the average change in the local magnetic induction due to unit displacement per unit area of domain walls. This change is much larger for \( 180^\circ \) domain walls than for non-\( 180^\circ \) domain walls [18].

Having discussed the source mechanisms of \( \Delta B \) and \( \Delta M \) signals, one is now in a position to qualitatively explain the various features observed in the experimental results. As the grain size increases in a perfectly annealed ferromagnetic material, the density of both \( 180^\circ \) and non-\( 180^\circ \) domain walls decreases [9]. Thus, \( \rho_{180^\circ}(H) \) and \( \rho_{n-180^\circ}(H) \) decrease as the grain size increases. Moreover, with increasing grain size, the mean free path of domain wall motion increases. Consequently, the incremental field (\( \Delta H \)) required for bulging the domain walls before they are unpinned also increases. Thus, \( \overline{V}(H) \) decreases with increasing grain size. As a result of this, it is expected that both \( \Delta B \), as well as \( \Delta M \), signals decrease with grain size in perfectly annealed nickel which is in agreement with the observations shown in Figures 10 and 12.

Comparing the steel with nickel data, it becomes apparent that the above picture is changed in the presence of precipitate, e.g., it has been shown by Kameda and Ranjan [19] that the grain boundary segregation increases the net magnetostatic energy and thereby, alters the domain structure. Thus, phosphorous segregation should affect \( \Delta B \) and \( \Delta M \) signals, but a correlation has not yet been clearly established. The precipitates seem to affect \( \Delta B \) and \( \Delta M \) signals in two ways. They can act as nucleation sites when both non-\( 180^\circ \) and \( 180^\circ \) domains are nucleated during magnetization [20]. As the density of these
precipitates increases, both $\rho_{180^\circ}(H)$ and $\rho_{n-180^\circ}(H)$ should increase too. They also could act as pinning sites for the growing domain walls. As the density of precipitates increases, the spacing between neighboring precipitates will decrease, which causes $\bar{V}_{180^\circ}(H)$ and $\bar{V}_{n-180^\circ}(H)$ to increase. As mentioned earlier, the decarburized steel samples contained precipitates and grain boundary phosphorous segregation. Therefore, it appears to us that in decarburized steels, the changes in MB and AB signals are altered due to the presence of MnS precipitates and grain boundary segregation of phosphorous, as shown in Figures 9 and 11. Further studies are necessary to resolve these problems.
CONCLUSION

Both AB and MB signals are sensitive to grain size and can be used for nondestructively measuring grain size under certain conditions. In nickel, both AB and MB signals decrease with increasing grain size since the grain size is the dominant microstructural factor affecting the magnetic properties in nickel. In decarburized steel, however, AB and MB signals increase with increasing grain size. This is due to the presence of precipitates and grain boundary phosphorous segregation which act as nucleation sites for domains and thus, the density of domain walls increases. Moreover, the presence of precipitates increases the average velocity of domain walls after unpinning. Precipitates, thus, are more important than grain size in determining the magnetic properties of steels.
REFERENCES

18. R. Ranjan, O. Buck and R. B. Thompson, A study of grain size effect on the magnetic properties of ferromagnetic materials using magnetic NDE methods (to be published in J. Appl. Phys.).


SECTION III. MICROSTRUCTURAL CHARACTERIZATION OF STEELS USING ACOUSTIC AND MAGNETIC BARKHAUSEN NOISES
INTRODUCTION

Microstructure evolution and solute segregation induced in a thermal or radiation environment often produce the degradation of the mechanical properties of steels. It is, therefore, necessary to develop a nondestructive evaluation (NDE) technique for examining the microstructure and local composition changes in order to control the safety operation of steel structures such as turbine rotors and pressure vessels. It has been recently recognized [1] that a magnetic NDE method has a great advantage over the NDE (e.g., ultrasonic and x-ray) methods because of its high sensitivity to the change of several metallurgical features.

Titto [2] studied the effect of aging of 0.014% C steel using magnetic Barkhausen (MB) signals. The Fe$_3$C carbide precipitation seemed to affect the MB signals and it was attributed to the stress field interaction between carbides and domain walls which causes domain wall pinning. Deimel et al. [3] have observed a remarkable change in MB, as well as acoustic Barkhausen (AB) signals during tempering of 22 Ni Mo Cr 37 and 15 Mn Mo NiV 53. However, no significant influence on the magnetic domain structure through the variation of the alloy carbide density, shape and size was observed. Several other research groups [4-6] have demonstrated that the characteristics of the AB and MB signals in steels are affected by an applied stress and heat treatment or compositions. However, the correlation of the microstructure (w.r.t carbide size, shape and density) with the AB and
MB signals has not been clearly established. Thus, a detailed study on the effect of several metallurgical features on the AB and MB signals is further required in order to make the magnetic NDE technique a viable tool.

This study is aimed at investigating the effect of the change in carbide morphology and hardness on the characteristics of the AB and MB signals. The observed results are discussed in terms of the nucleation and growth processes of domains in a magnetic field which are related to the source mechanisms of the two types of the Barkhausen signals.
EXPERIMENTAL PROCEDURE

This study employed three commercial carbon steels (1020, 1045 and 1095). The chemical compositions of the steels are listed in Table 1. The 1020 steel was austenitized at 900°C for 1 hr. and quenched into ice water at -10°C, and then, tempered at various temperatures ranging from 100°C to 600°C. All the steels were spheroidized at 700°C for 40 hrs. The hardness of the quenched and tempered (QT) and spheroidized steels was measured using a Rockwell hardness C and B ($R_C$ and $R_B$), respectively. In order to reveal the carbide morphology, specimens of the QT steel were etched using solution ($5 \text{ g CuCl}_2$, $100 \text{ ml HCl}$, $100 \text{ ml CH}_3\text{OH}$ and $100 \text{ ml H}_2\text{O}$). In addition, specimens of the spheroidized steels were pre-etched in 4% picral and followed by etching in Beraha's #1 solution ($3 \text{ g KHSO}_4$, $10 \text{ g Na}_2\text{S}_2\text{O}_3$ and $100 \text{ ml H}_2\text{O}$). The carbide precipitate morphology of the QT and spheroidized steels was examined using scanning electron and optical microscopes.

An alternating magnetic field, $H$, was applied to cylindrical specimens of 8 mm diameter and 82 mm length inside a solenoid coil powered by a bipolar power supply. The magnetic hysteresis loop (B-H) of each sample was measured by a commercial hysteresisgraph using a Hall probe and a pick-up coil of 400 turns. AB signals induced by a magnetic field were detected using a piezoelectric sensor with 175 KHz resonance frequency attached to the end surface of the specimen. A band pass filter ranging from 150 to 200 KHz and a gain of 80 dB were used for the AB signal measurement. MB signals were measured by a
Table 1. Chemical compositions of steels (wt.%)

<table>
<thead>
<tr>
<th></th>
<th>C</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
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<td>0.82</td>
<td>0.016</td>
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<td>1045</td>
<td>0.46</td>
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<td>0.009</td>
<td>0.023</td>
</tr>
<tr>
<td>1095</td>
<td>0.96</td>
<td>0.45</td>
<td>0.019</td>
<td>0.031</td>
</tr>
</tbody>
</table>

pick-up coil of 4000 turns with the gauge length of 63 mm. This optimum number of turns of the pick-up coil was determined so as to maximize the signal/noise ratio [7]. A band pass filter ranging from 1 to 300 KHz and a gain of 60 dB were used for the MB signal measurement. Both the AB and MB signal data were recorded using a digital memory oscilloscope with time intervals ranging from 5 to 200 μsec. The characteristics of the AB and MB signals vs. applied H were plotted on an X-Y recorder using an analog/digital converter. The average peak voltage of the AB and MB signals was determined from 50 data. The standard deviation was within ±10% of the average peak value. In order to examine the magnetization frequency dependence of the AB and MB signals, the magnetic sweep rate, dH/dt, was varied.
RESULTS

Quenched and Tempered 1020 Steel

Figure 1 shows an optical micrograph of the lathe martensite microstructure of QT 1020 steel. The prior austenite grain size was about 100 μm. Tempering treatment caused the precipitation of fine carbides and the softening of the martensite. It is well known that epsilon carbides appear at tempering temperatures between 100 and 200°C and cementites (Fe₃C) start precipitating between 200 and 300°C. Figures 2(a)–2(f) show scanning electron micrographs of fine carbide precipitates in the QT steel tempered at various temperatures. The morphology of the carbide precipitation changed from the needle-like to spheroidal shape, and the size and spacing of carbides increased with increasing tempering temperature. The carbide size varied from ~0.08 to ~0.2 μm and the spacing from ~0.2 to ~0.8 μm in the 300 to 600°C tempering range. Note that in the sample tempered at relatively low temperatures, some island regions did not reveal carbide precipitates (Figures 2(a) and 2(b)).

The magnetic hysteresis loop of the as-quenched and the specimens tempered between 350 and 600°C are shown in Figure 3. The magnetic hysteresis behavior strongly depends on the tempering temperature which produces different carbide morphology and hardness. In order to understand the effect of the tempering treatment on the magnetic properties, the variations of the saturation magnetic induction, $B_s$ (defined at $H = 1.43 \times 10^4$ A/m), the coercive force, $H_c$, and the
Figure 1. Lathe martensite microstructure of QT 1020 steel
Figure 2. Carbide precipitation morphology: (a) and (b) tempered at 300°C, (c) and (d) tempered at 400°C, and (e) and (f) tempered at 600°C
Figure 3. Magnetic hysteresis loop of as-quenched, and 350 and 600°C tempered 1020 steels, indicating peak position of acoustic and magnetic Barkhausen signals (major peak with * and minor peak with **)
hysteresis loss, $W_h$, as well as Rockwell hardness, $R_C$, with tempering temperature are shown in Figure 4. The carbide morphology change is also depicted in this figure. The values of $H_C$ and $W_h$ started decreasing at lower tempering temperatures than that of $R_C$. The hardness change is strongly related to the coarsening process of cementites. A marked change of $H_C$, $W_h$ and $B_s$ was found to occur if specimens were tempered between 200 and 400°C. It appears from the micrographs shown in Figures 2(a)-2(d) that the change in needle-like cementite structure gives rise to the change in the magnetic properties.

Typical AB signal characteristics of the as-quenched and tempered steels (350° and 600°) are shown in Figures 5(a)-5(c). The AB voltage increased substantially with increasing tempering temperature. A double peak was observed in all of the QT samples except the as-quenched and the ones tempered at 100°C. The AB signal became more prominent with increasing tempering temperature. As shown in Figure 3, when $H$ is changed so as to induce the magnetization path indicated by an arrow, a major AB peak (AB*) occurs in the lower knee region and a minor AB peak (AB**) does in the upper knee region. While the position of AB* remains the same, independent of the tempering temperature. In order to characterize the AB behavior, the value of the major AB peak signal, $(V_{AB})^{max}$, was selected. It should be noted that the major AB peak voltage was about proportional to the minor peak voltage. Figure 6(a) shows the relationship of $(V_{AB})^{max}$ to $dH/dt$ for various tempering conditions. Figure 6(b) shows the variation of $(V_{AB})^{max}$ with the tempering temperature for different $dH/dt$. A linear dependence
Figure 4. Variations of hardness, saturation magnetic induction, coercive force, and hysteresis loss as well as carbide morphology with tempering temperature in QT 1020 steel.
Figure 5. Characteristics of AB signal vs. applied magnetic field: (a) as quenched, and (b) 350 and (c) 600°C tempered 1020 steels (dH/dt=5.73x10^4 A/msec). An arrow indicates the magnetic sweep direction. ▲ and ▼ indicate major and minor AB peak signals, respectively.
Figure 6. Relationships of $(V_{AB})_{max}$ to (a) $dH/dt$ for various tempering conditions and (b) tempering temperature for several values of $dH/dt$ in QT 1020 steel.
Figure 6. Continued
of \((V_{AB})_{\text{max}}\) on \(\frac{dH}{dt}\) was found in the QT 1020 steel tempered at various temperatures. The slope drastically increased with increasing tempering temperature. It is clear that \((V_{AB})_{\text{max}}\) changes greatly at tempering temperatures ranging from 200 to 400°C.

We now evaluate the characteristics of the MB signal in the QT steels. Figures 7(a)–7(c) show the characteristics of the MB signal emitted during the magnetization of the as-quenched, and 350 and 600°C tempered steels. Compared to the characteristics of the AB signal, several different features of the MB signal were observed. As shown in Figures 3 and 7, a single MB signal peak was observed when \(H\) was decreased from the saturation state to zero (an arrow with MB indicates the position of the MB signal peak in Figure 3). The critical magnetic field which produces the MB signal peak greatly increased with increasing tempering temperature. Similarly, we used the MB peak voltage, \((V_{MB})_{\text{max}}\), to represent the effect of microstructure. The relationships of \((V_{MB})_{\text{max}}\) to \(\frac{dH}{dt}\) for various tempering conditions are shown in Figure 8(a). Figure 8(b) also shows the variation of \((V_{MB})_{\text{max}}\) with tempering temperature for different \(\frac{dH}{dt}\). Unlike indicated by the results of the AB signal, a nonlinear dependence of \((V_{MB})_{\text{max}}\) on \(\frac{dH}{dt}\) was found. The value of \((V_{MB})_{\text{max}}\) asymptotically increased as \(\frac{dH}{dt}\) was increased. \((V_{MB})_{\text{max}}\) increased more gradually with increasing tempering temperature than \((V_{AB})_{\text{max}}\). A small change in \((V_{MB})_{\text{max}}\) occurred in the temperature range between 200 and 400°C (Regime 3). This is the regime in which the needle-like carbides precipitate out. Moreover, two relatively large transitions of \((V_{MB})_{\text{max}}\) occurred
Figure 7. Characteristics of MB signal vs. applied magnetic field: (a) as quenched, and (b) 350°C and (c) 600°C tempered 1020 steels (dH/dt=5.73x10⁴ A/m·sec). An arrow indicates the magnetic sweep direction. ▲ indicates MB peak signals.
Figure 8. Relationships of $(V_{MB})^{max}$ to (a) $dH/dt$ for various tempering conditions and (b) tempering temperature for several values of $dH/dt$ in QT 1020 steel.
Figure 8. Continued
between 100°C and 200°C where epsilon carbides precipitate (Regime 2) and between 400°C and 500°C where carbides start spheroidizing (Regime 4).

Spheroidized Steels

Microstructures of the spheroidized 1020, 1045 and 1095 steels are shown in Figures 9(a)-9(c), respectively. The spheroidized 1020 steel showed coarser ferrite grain than the other steels. The cluster of spheroidized carbides was formed in the 1020 steel, while carbides were more uniformly distributed in the 1045 and 1095 steels. The density of spheroidized carbides increased with increasing carbon content. The carbide size increased from ~0.7 to ~1.5 μm and the spacing decreased from ~8 to ~3 μm as the carbon content increased. The carbide size and spacing in the spheroidized steels were an order of magnitude greater than in the QT steel.

The magnetic hysteresis behavior in the spheroidized steels was found to be similar to that in the 1020 steel tempered at 600°C. Figure 10 shows the variations of \( R_B \), \( B_s \), \( H_c \) and \( W_h \) with carbon content. The values of \( H_c \) and \( W_h \), as well as \( R_B \), slightly increased and that of \( B_s \) decreased with increasing carbon content. The decrease in \( B_s \) is due to an increase in the volume fraction of carbides which are very weak ferromagnetic substances.

As shown in Figures 11(a)-11(c), double peaks of the AB signal were also observed in the spheroidized steels. However, the minor peak in the spheroidized steels was not as clear as in the QT 1020 steel (see
Figure 9. Microstructures of spheroidized steels: (a) 1020, (b) 1045, and (c) 1095.
Figure 10. Variations of hardness, saturation magnetic induction, coercive force, and hysteresis loss with carbon content in spheroidized steels.
Figure 11. Characteristics of AB signal vs. applied magnetic field: (a) 1020, (b) 1045, and (c) 1095 \((dH/dt=5.73 \times 10^4 A/m \cdot \text{sec})\). An arrow indicates the magnetic sweep direction. \(^{\uparrow}\) and \(\dagger\) indicate major and minor AB peak signals, respectively.
also Figures 5(b) and 5(c)). The major AB peaks were found to occur at the same part of the B-H loop as in the case of the QT steel. A linear dependence of \((V_{AB})_{\text{max}}^2\) on \(dH/dt\) was also found in the spheroidized steels, as shown in Figure 12(a). \((V_{AB})_{\text{max}}^2\) decreased with increasing carbon content (Figure 12(b)). It should be noted that the magnitude of \((V_{AB})_{\text{max}}^2\) in the spheroidized 1020 steel is markedly lower than in the 600°C tempered steel (compare between Figures 6(b) and 12(b)).

All the spheroidized steels exhibited a distinct single MB signal (Figures 13(a)-13(c)). The critical magnetic field which produces the MB signal peak was nearly independent of the carbon content and was the same as in the case of 1020 steel tempered at 600°C (see Figure 3). Figure 14(a) shows a nonlinear dependence of \((V_{MB})_{\text{max}}^2\) on \(dH/dt\) for all the spheroidized steels. It is apparent from Figure 14(b) that \((V_{MB})_{\text{max}}^2\) is greatest for the 1045 steel. However, as shown in Figure 13, the average MB voltage decreased with increasing carbon content.
Figure 12. Relationship of \((V_{AB})_{\text{max}}\) to (a) \(\frac{dH}{dt}\) for three spheroidized steels and (b) carbon content for several values of \(\frac{dH}{dt}\)
Figure 12. Continued

Peaks Value of Acoustic Barkhausen Signal, (V(AB)max, μV)

Sweep Rate (10^4 A/m sec)
- 15.18
- 9.17
- 5.73
Figure 13. Characteristics of MB signal vs. applied magnetic field: (a) 1020, (b) 1045, and (c) 1095 (dH/dt=5.73x10^4 A/m·sec). An arrow indicates the magnetic sweep direction.  Indicates MB peak signals.
Figure 14. Relationship of \( (V_{MB})^{max} \) to (a) \( dH/dt \) for three spheroidized steels and (b) carbon content for several values of \( dH/dt \).
Figure 14. Continued
DISCUSSION

The effect of microstructure (i.e., carbide precipitation and hardness) on the AB and MB signals induced during magnetization was investigated in QT and spheroidized steels. Many different features in the AB and MB signals were observed in the present study as summarized in the following.

1. In general, AB signals exhibit a double peak, whereas MB signals show a single peak.

2. In all the steels, \((V_{AB})^{\text{max}}\) changes linearly, whereas \((V_{MB})^{\text{max}}\) changes nonlinearly with respect to the magnetic sweep rate.

3. An abrupt increase in \((V_{AB})^{\text{max}}\) and a gradual increase in \((V_{MB})^{\text{max}}\) occur at tempering temperatures between 200 and 400°C where needle-like carbides precipitate in QT steel.

4. In the spheroidized steels, \((V_{AB})^{\text{max}}\) decreases with increasing carbon content, i.e., carbide density, while \((V_{MB})^{\text{max}}\) is at a maximum when the carbon content is 0.46 wt%.

It is clear that the aforementioned characteristics are strongly related to the different source mechanisms of the two types of the Barkhausen signals.

Dynamics of Domain Walls

In order to discuss the source mechanisms of the AB and MB signals, we must first consider the change of the domain structure induced during magnetization. The change of the domain structure with
H is due to several processes such as the nucleation, annihilation and growth of domains during magnetization. These changes are affected by defects such as precipitates, grain boundaries and dislocations and are, therefore, governed by the probability of their occurrence. Since the cumulative probability of the domain wall motion may be described by Boltzmann statistics, an energy consideration can be used as a guideline for model calculations. A schematical illustration (which will be explained in detail later) of the variation of the probability of the nucleation, annihilation and growth of domains, and the domain wall density as a function of H is shown in Figure 15. This illustration is used to discuss the occurrence of AB and MB signals [8].

The indicated magnetization path of the material's B-H curve has been divided into two parts, namely, Regions I and II (Figure 15(a)). In the following, we discuss these regions separately.

When a magnetic field is reduced from the field that induces saturation magnetization (Region I), new domains are nucleated close to precipitates and grain or lathe boundaries where magnetic poles are accumulated during the spin rotation [2]. The domain nucleation can proceed when the reduction in the magnetostatic energy associated with the poles during the formation of new domains is greater than the work required to form domains [9]. The probability of the domain nucleation, \( P_n \), increases with decreasing H so as to reduce the magnetostatic energy associated with the poles. Thus, the density of domain walls, \( F_d \), increases with decreasing H (Figure 15(c)). The driving force for the growth of the domains arises from the difference
Figure 15. Schematic illustration of (a) magnetization path (Regions I and II) in hysteresis loop, and variation of (b) probability of nucleation, $P_n$, annihilation, $P_a$, and growth, $P_g$, of domains, (c) domain wall density, $F_d$, (d) density of propagating domain walls, $N_g$, and (e) $H$ derivatives of density of growing and nucleated domains, $dN_g/dH$ and $dN_n/dH$, with magnetic field.
between the applied field energy and the domain wall energy. Thus, the probability for the growth of the domains, \( P_g \), decreases with decreasing \( H \) as the wall energy of the domain preferentially oriented to the \( H \) direction and the field energy decrease, as shown in Figure 15(b).

When a magnetic field in the reverse direction increases beyond \( H_c \) (Region II), domains formed in Region I grow and the spin in the domains rotates at the expense of unfavorably oriented domains. The annihilation of the unfavorably oriented domains associated with the growth of the favorably orientated ones and the spin rotation results in a decrease in \( F_d \) (i.e., the domain wall energy). Thus, as schematically shown in Figure 15(c), the value of \( F_d \) becomes greatest at about \( H=0 \). As the driving force for the domain wall propagation increases with increasing \( H \), \( P_g \) as well as the probability for domain annihilation, \( P_a \), increases (Figure 15(b)).

Source Mechanism for Acoustic Barkhausen Signals

The source of the AB signal is the emission of elastic waves due to a sudden change in stress fields when domain walls pinned at defects are released. The rapid motion of the domain walls causes a change in the localized magnetostrictive strain.\(^1\) Therefore, it is reasonable to assume that the source of the AB signal is mainly related to the

---

\(^1\)The motion of non-180° domain walls produces magnetostrictive strains, while that of 180° domain walls does not [8]. In the present steels with high density of dislocations or precipitates, it appears that the dynamic behavior of both non-180° and 180° domain walls is identical during magnetization. Therefore, the type of domain walls is not specified in the discussion.
acceleration of domain walls released from pinning sites. On the basis of seismology theory [10], the acoustic signal caused by a single source is proportional to the accelerating velocity of a source and the strain change induced by the motion of the source. Assuming that the interaction between many sources (i.e., propagating domain walls) is negligible, the intensity of acoustic signals induced by many sources may be given by the product of the number of acoustic sources and the individual acoustic source strength. Thus, the AB signal intensity, $V_{AB}(H)$, can be approximately described by

$$V_{AB}(H) = C \cdot \left( f_g \cdot P_g(H) \cdot \frac{ds}{dt} \cdot \varepsilon_m \right)$$  \hspace{1cm} (1)$$

where $C$ is the numerical constant which depends on the specimen volume and geometry, the elastic moduli and the characteristics of the instrument; $f_g$ is the volume fraction of defects such as carbides and grain boundaries; $ds/dt$ is the average accelerating velocity of domain walls released from the pinning sites; and $\varepsilon_m$ the local magnetostrictive strain change associated with the jump of a single domain wall. The first bracket indicates the density of domain walls which jump from pinned to unpinned states and the second one indicates the average individual source strength. It should be noted that $P_g$ reflects a macroscopic drag effect and $ds/dt$ represents the microscopic pinning effect.

In Eq. (1), $C$, $\varepsilon_m$, $f_g$ and $ds/dt$ are thought to be independent of $H$. For a given $dH/dt$, the distribution of $V_{AB}(H)$ is assumed to be
proportional to the density of jumping domain walls, \( N_\text{g}(H) = F_d(H) \cdot P_d(H) \), as schematically depicted in Figure 15(d). This explains the double peak structure which was consistently observed in the AB signals from most of the steel samples. However, in all cases, \((V_{AB})^{\text{max}}\) was observed at the same \(H\) (Figure 3).

From Figures 6(b) and 12(b), we deduce the relationship between \((V_{AB})^{\text{max}}\) and \(dH/dt\) to be

\[
(V_{AB})^{\text{max}} = \alpha \frac{dH}{dt}
\]

where \(\alpha\) is the proportionality constant. Comparing Eqs. (1) and (2) and using Figures 6(b) and 12(b) indicate that \(\alpha\) depends on the microstructure as discussed in the following. \(ds/dt\), as well as \(P_g\), depends on two types of interactions of domain walls with defects: the surface tension and magnetic pole effects [11-14]. The action of a magnetic field counterbalances either the domain wall tension or the magnetostatic interaction force between the domain wall and the poles around defects. As a first order approximation, the displacement of pinned domain walls, \(s\), is assumed to be proportional to the local applied field, as well as the spacing of pinning sites [11-14]. It is then apparent that the accelerating motion of domain walls released from the pinning sites is related to the magnitude of \(s\) just before unpinning. The time derivative of \(s\) can be justified because of the fast time decay during the AB signal measurement. Thus, \(ds/dt\) should be proportional to \(dH/dt\) for small \(dH/dt\), as is true for the present
While $H_c$ gives a macroscopic measure of some average resistance to the irreversible domain growth, the AB measurement reflects the jump frequency of domain walls pinned at defects on a microscopic scale. It is expected that there exists a correlation between the two quantities. As shown in Figure 16, $\alpha$ decreases with increasing $H_c$. It should be noted that the results of the QT and spheroidized steels fall on different curves. This is attributed to the different carbide morphologies and hardesses between the two types of the steels. The microstructure effect on the AB signal will be discussed later.

Source Mechanism for the Magnetic Barkhausen Signals

We now consider the source mechanism for the MB peak signals induced during the sweeping of $H$. The magnetic induction is attenuated due to the eddy current loss. Thus, the characteristics of the MB signal measured by a pick-up coil are related to the magnetic flux change localized in the specimen surface region. The intensity of the MB signal is assumed to be proportional to the product of the time derivative of the magnetic flux density, $dB/dt$, and the effective surface skin volume, $A$.

The magnetic flux density change is induced by the magnetic moment change during the nucleation, as well as growth processes of domains. Thus, we assume that the magnetic flux density, $B$, can be described by

$$B = \lambda \left( \beta_n N_n + \left( \frac{S}{w} \right) N_g \right) \tag{3}$$
Figure 16. Relationship of the proportionality constant, $\alpha$, with respect to relation between $(V_{AB})_{\text{max}}$ and $dH/dt$ to $H_c$ for QT and spheroidized steels.
where \( \lambda \) is the numerical coefficient related to the atomic magnetic moment, \( \beta_n \) is the coefficient related to the spike shape of nucleated domains, \( N_n \) is the density of nucleated domain walls which is proportional to the probability of nucleation, \( P_n \), \( \bar{s} \) is the average displacement of the domain wall, \( \delta_w \) is the domain wall thickness and \( N_g \) as defined above. The first and second terms indicate the effects of the nucleation and growth of domains on the magnetic flux change, respectively. Since the time constant of a pick-up coil used in this study (the order of msec) is much greater than the individual event duration of the domain nucleation and growth processes (the order of \( \mu \)sec), the time derivative can be converted into the \( H \) derivative multiplied by \( dH/dt \). Then, \( dB/dt \) is given by

\[
\frac{dB}{dt} = \lambda \left( \beta_n \frac{dN_n}{dH} + \left( \frac{\bar{s}}{\delta_w} \right) \frac{dN_g}{dH} + \left( \frac{\bar{\delta}^2}{\delta_w^2} \right) \frac{d\bar{s}}{dH} \right) \frac{dH}{dt}.
\] (4)

The variations of \( dN_n/dH \) and \( dN_g/dH \) with \( H \) are schematically shown in Figure 15(e). Experimental results shown in Figure 3 indicate that the MB signal peak appears in the region where nucleation of domains dominate. Thus, in order to discuss the origin of MB peak signal, we assume that the contribution from \( dN_g/dH \) is small. Therefore, the variation of \( dB/dt \) is mainly attributed to the terms of \( dN_n/dH \) and \( N_g \cdot d\bar{s}/dH \). \( \bar{s} \), the mean free path of domain walls, is considered to be a continuous function of \( H \). When the materials contain only a small density of defects, the growth terms (i.e., \( d\bar{s}/dH \)) should become significant and the double MB peak signal can be observed due to the
double peak structure of \( N \) \( g \) [13]. However, the mean free path of propagating domain walls is short in the materials with many defects. Thus, \( ds/dH \) may be neglected in the present case and the major contribution to the MB peak signal in steels is from the domain nucleation \( (dN_n/dH) \). This is supported by the fact that a MB peak (MB\(*\)) always appears in Region I where the nucleation of domains dominates (Figures 3 and 15(a)). Therefore, \( (V_{MB})^{\text{max}} \) can be given by

\[
(V_{MB})^{\text{max}} = \lambda \beta \frac{dN_n}{dH} \frac{dH}{dt} A .
\]  

It should be noted that the value of \( A \) varies with \( dH/dt \) because of the eddy current losses. From electrodynamics, it is likely that \( A \) decreases exponentially with increasing \( dH/dt \). Consequently, the dependence of the MB signal on \( dH/dt \) becomes nonlinear, as can be seen in Figures 8(a) and 14(a).

Microstructural Effect

As mentioned earlier, the source mechanisms of the AB and MB signals in steels are related to the growth and nucleation processes of domains, respectively. In order to discuss the effect of microstructure on the source mechanisms of the AB and MB signals, the variation of hardness, carbide size, \( a \), and spacing, \( \lambda \), which are related to \( f_g \), the volume fraction of defects, with tempering temperature and carbon content in the QT and spheroidized steels are first summarized in part A of Table 2. In this table, upward and downward arrows indicate the increasing and decreasing trends of the
Table 2. Summary of microstructure and domain behavior during magnetization in QT and spheroidized steels

<table>
<thead>
<tr>
<th>PART A</th>
<th>PART B</th>
<th>PART C</th>
</tr>
</thead>
<tbody>
<tr>
<td>Microstructural parameter</td>
<td>Domain behavior</td>
<td>Prediction</td>
</tr>
<tr>
<td>Hardness</td>
<td>Size</td>
<td>Carbide</td>
</tr>
<tr>
<td>QT steel:</td>
<td></td>
<td></td>
</tr>
<tr>
<td>$(a&lt;\delta_w$ and $\ell&lt;5\delta_w)$ tempering</td>
<td>$\uparrow$</td>
<td>$\uparrow$</td>
</tr>
<tr>
<td>temperature</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Spheroidized steel:</td>
<td></td>
<td></td>
</tr>
<tr>
<td>$(a&gt;\delta_w$ and $\ell&gt;5\delta_w)$ Carbon content</td>
<td>$\uparrow$</td>
<td>$\uparrow$</td>
</tr>
</tbody>
</table>
metallurgical factors with increasing tempering temperature and carbon content, respectively. The size of the arrows indicates the magnitude of the change. Secondly, the effect of hardness, carbide size and spacing on several magnetic domain parameters such as $P_g$, $ds/dt$, $F_d$ and $dN_n$ (in Eqs. 1 and 5) have been discussed earlier with the trends being summarized in Table 3. In the following, we try to predict the characteristics of the AB and MB peak signals in the QT and spheroidized steels (summarized in parts B and C of Table 2) with the aid of Table 3 and part A of Table 2.

The increase in hardness which arises from the local stress surrounding dislocations increases the magnetoelastic interaction between the domain walls and the dislocation [15] and thereby, increases the resistance to the growing domains (i.e., a decrease in $P_g$ and $ds/dt$). Since the magnetoelastic energy caused by the local stress does not contribute to the domain nucleation, the hardness change does not affect $dN_n/dH$ and $F_d$. On the other hand, the carbide size and spacing affect these quantities strongly, as will be discussed qualitatively in the following. The carbide size affects the domain growth and nucleation differently, depending on the carbide size and $\delta_w$ [12]. For a carbide size smaller than $\delta_w$, the pinning strength of domain walls at carbides increases with increasing carbide size. Therefore, $P_g$ and $ds/dt$ decrease as the carbide size increases. However, the small carbides which cause a small change in the magnetostatic energy act as a weak nucleation site and, therefore, do not greatly affect $dN_n/dH$ and $F_d$. When the carbide size is greater than $\delta_w$ ($a > \delta_w$), the carbide starts
Table 3. Summary of effect of the various microstructural factors on parameters describing dynamic domain behavior

<table>
<thead>
<tr>
<th>Microstructural factors</th>
<th>Domain characteristic parameters</th>
<th>$P_g$</th>
<th>$\frac{ds}{dt}$</th>
<th>$F_d$</th>
<th>$\frac{dN}{dH}$</th>
</tr>
</thead>
<tbody>
<tr>
<td>Hardness</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Carbide $a &lt; \delta_w^a$</td>
<td></td>
<td>\downarrow</td>
<td>\downarrow</td>
<td>\uparrow</td>
<td>\uparrow</td>
</tr>
<tr>
<td>Size, $a &gt; \delta_w$</td>
<td></td>
<td>\downarrow</td>
<td>\uparrow</td>
<td>\uparrow</td>
<td>\uparrow</td>
</tr>
<tr>
<td>Carbide $\lambda &lt; 5\delta_w^b$</td>
<td></td>
<td>\uparrow</td>
<td>\uparrow</td>
<td>\uparrow</td>
<td>\uparrow</td>
</tr>
<tr>
<td>Spacing, $\lambda &gt; 5\delta_w$</td>
<td></td>
<td>\uparrow</td>
<td>\uparrow</td>
<td>\downarrow</td>
<td>\downarrow</td>
</tr>
</tbody>
</table>

$^{a} \delta_w$ is the domain wall thickness (1000Å).

$^{b}$ We estimate roughly the critical spacing of $5 \delta_w$ which corresponds to the critical length of spike domain based on the geometric considerations of domain size.
acting as a domain nucleation site rather than a pinning site since the magnetostatic energy will increase. An increase in carbide size, therefore, increases \( \frac{dN_n}{dH} \) and \( F_d \), and the size effect on \( P_g \) and \( \frac{ds}{dt} \) becomes relatively small. In general, an increase in spacing, \( \ell \), results in an increase in \( P_g \) and \( \frac{ds}{dt} \) since the growth resistance against growth of domains decreases. The spacing of coarse carbides \((a > \delta_w)\), therefore, strongly affects the nucleation process. If the spacing is small \((\ell < \sim 5 \delta_w)\), the strong magnetostatic interaction between the neighboring precipitates makes nucleation more difficult. As a result, \( \frac{dN_n}{dH} \) and \( F_d \) increase with increasing spacing. On the contrary, if the carbide spacing is relatively large \((\ell > \sim 5 \delta_w)\), the density of the noninteracting nucleation sites (i.e., on the carbide) decreases as the spacing increases. This decreases \( \frac{dN_n}{dH} \), as well as \( F_d \).

Applying the aforementioned qualitative relationships in conjunction with part A of Table 2, the effect of tempering temperature and carbon content on the characteristic domain parameters is summarized in part B of Table 2. Using Eqs. (1) and (5), the trends of \( (V_{AB})_{\text{max}} \) and \( (V_{MB})_{\text{max}} \) are given in part C of Table 2. The predicted results appear to be consistent with our observations (see Figure 8(b), e.g.). We must note the following interesting features of AB and MB signals in the two types of the steels. In the QT steel, the change of carbide shape from needle-like to spheroidal with increasing tempering temperature should affect the nucleation process of domains and therefore, \( (V_{MB})_{\text{max}} \), as shown in Figure 8(b). Since the carbide
size and spacing have an opposite effect on $\frac{dN_n}{dH}$ in the spheroidized steels, the relationship of $(V_{MB})^{max}$ to the carbon content shows a peak in Figure 14(b). The different characteristics of $(V_{AB})^{max}$ between the QT and spheroidized steels shown in Figures 6(a) and 12(a) are due to the large difference in the microstructure.
CONCLUSION

Microstructural parameters (such as hardness and carbide morphology) of a QT and spheroidized steels were nondestructively evaluated by measuring the AB and MB signals induced during magnetization. In all the steels investigated here, a linear dependence of the AB peak signal on $dH/dt$ was observed; on the other hand, the relationship of the MB peak signal and $dH/dt$ was found to be nonlinear. The QT steel showed an abrupt increase in the AB peak voltage and a gradual increase in the MB peak voltage if tempered between 200 and 1°C. The AB peak voltage in the spheroidized steels continuously increased with increasing carbon content; in contrast, the MB peak voltage became greatest when the carbon content was 0.46 wt%. The different characteristics of the AB and MB peak signals with respect to microstructures were rationalized in terms of the source mechanisms for the two types of Barkhausen signals which are controlled by the growth and nucleation processes of domains, respectively.
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Magnetic NDE techniques, namely, the acoustic Barkhausen noise (AB), the magnetic Barkhausen noise (MB) and the magnetic hysteresis curves were used for microstructural characterization of ferromagnetic materials. Simultaneous use of aforementioned techniques was conducted in order to be able to understand the interaction of various types of domain walls, viz., 180° and non-180° domain walls with various microstructural features.

It was observed that in nickel, non-180° domain walls interact more strongly with dislocations than non-180° domain walls. This supports the theoretical calculations of Theiner et al. [17]. In addition, the MB signals exhibited a "valley" in the recrystallization temperature range. This is supported to be a combined effect of the change in grain size, as well as a change of the dislocation density.

MB and AB signals showed a great potential as a NDE tool for the grain size measurement in nickel as well as decarburized steels. In nickel, however, MB signals are more sensitive to grain size changes than AB signals and thus suggests that the 180° domain walls interact more strongly with grain boundaries than non-180° domain walls.

In steel, both AB and MB signals were sensitive to the change in carbide morphology induced during tempering at various temperatures. It is suggested that the MB signal peak could be attributed to the domain nucleation. The AB signal peak, on the other hand, is attributed to the domain growth during magnetization.
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APPENDIX: DESIGN OF A MAGNETIC BARKHAUSEN SENSOR

The MB noise, as stated earlier, is caused by abrupt changes in magnetization with a changing magnetic field $H$, principally by the motion of $180^\circ$ domain walls [1-3]. These are detected by having a search coil around the sample. As depicted in Figure 1, the discontinuous motion of $180^\circ$ domain walls results in a small induction change, generating a $B$ field given by Eq. (1), which propagates within the material before arriving at a surface and, thereby, inducing a time-varying voltage, $V(t)$, in the search coil.

$$\nabla^2 B(r, \theta) = 4\pi \mu \sigma \frac{dB}{dt}.$$  \hspace{1cm} (1)

In Figure A1, $M(t)$ and $S(t)$ are material and microstructure dependent functions. Thus, the reliability of the technique depends strongly on the function $G(t)$, i.e., sensor design. Figure A2 shows the schematic diagram of the essential features of the arrangement of a magnetic Barkhausen system.

The main problem with the design of a MB sensor is that the amplitude of the induced voltage pulse during a Barkhausen jump is small and of the order of a few microvolts. If any small induced voltages are to be detected, it is essential that the sensor and the amplifier should be designed to give the lowest possible signal-to-noise ratio. It is also essential to carry out a thorough examination of factors affecting the relation between the primary changes and measured
Induced Voltage: \( V(t) \propto G(t) \times M(t) \times S(t) \)

Figure A1. Depiction of magnetic Barkhausen noise generation
Figure A2. Schematic diagram of the essential features of a magnetic Barkhausen system.
effects. This involves theoretical investigations into the rate of
decay of the induction in a cylindrical specimen following a
discontinuous change in magnetization, and into the effects of the
sensor on the form of the induced voltage. As stated earlier, a small
induction change due to the discontinuous motion of a 180° domain wall,
propagates within the material following Eq. (1). As a result of this,
the induced voltage on the surface of the sample can be solved with
appropriate boundary conditions [4] and is given by:

\[
e = \frac{\phi_0}{2\pi \mu \sigma a^2} \sum_{n=1}^{\infty} \frac{\lambda_n J_n(\lambda_n b/a)}{J_0(\lambda_n)} \exp \left( -\frac{\lambda_n^2 \tau t}{4\pi \mu \sigma a^2} \right)
\]  

(2)

where \( a \) is the radius of the cylinder, \( b \) is the location of the
induction change, \( \lambda_n \) is the \( n \)th root of the Bessel function, \( J_n(\lambda) = 0 \).

The design approach of the sensor is based on considering the
Barkhausen jump and the sensor as two coupled inductors with individual
time constants and with a coupling constant which depends on the
relative geometry of sensor and the specimen. When the time constant of
the sensor coil, \( \tau_s \), is much longer than the time constant of the
Barkhausen jump, \( \tau_s \), the induced voltage as a result of \( \Delta \phi \) flux change
from a Barkhausen jump is

\[
v(t) = \frac{n \Delta \phi}{\tau_c} \left[ e^{-t/\tau_c} - e^{-t/\tau_s (1-q^2)} \right] \tau_s \ll \tau_c
\]

(3)

where \( q \) is the coupling constant. The maximum voltage is obtained by a
differentiation and yields
\[ V_{\text{max}} = \frac{n\Delta \phi}{\tau_c} \]  

\( \tau_s \) can be derived from the exponential term in Eq. (2). Thus, a proper choice of \( \tau_c \) with appropriate impedance matching and subsequent signal conditioning can yield the best sensor with subsequent signal conditioning for detection of magnetic Barkhausen noises. For example, \( \tau_s \) for iron has been found experimentally to be \( \sim 2-30 \mu\text{sec} \) and

\[ \tau_c = \frac{L_c}{R_c} \]

where \( L_c \) and \( R_c \) are the inductance and resistance of the sensor coil, respectively. In order to satisfy Eq. (3), \( \tau_c \) should be much larger than \( 2-30 \mu\text{sec} \). Moreover, for the desired frequency range of the signal, \( \tau_c \) is chosen such that the experimental conditions fall in the low noise region in the typical noise figure contours of the preamplifier (e.g., Model 113, EG&G low-noise amplifier).

Conclusion

For the particular material used in the experiments, \( \tau_s \) can be obtained from Eq. (2), which contains materials properties. A proper choice of \( \tau_c \) optimizes the sensor with respect to the low-noise preamplifier used (high signal-to-noise ratio) leading to an excellent Barkhausen signal detector.
References


