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Influence of a high pulsed magnetic field on the tensile properties and phase transition of 7055 aluminum alloy

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Abstract
The effect of a high pulsed magnetic field on the tensile properties and microstructure of 7055 alloy were investigated. In the tensile properties test, the pulsed magnetic field was applied to improve the tensile strength and elongation via the magnetoplasticity effect. The results show that when the magnetic induction intensity (B) is 3 T, the tensile strength and elongation arrives at the maximum synchronously, which has been enhanced by 7.9% and 20% compared to the relevant 576.5 MPa (σ₀), 7.5% (δ) of the initial sample without magnetic field treatment. The high magnetic field takes effect by altering the spin state of free electrons stimulated between the dislocations and obstacles; afterwards, the structural state of the radical pair is converted from the singlet state with high bonding energy to the triplet state with low bonding energy. Under this condition, the dislocation mobility is enhanced and it becomes easier for a dislocation to surmount the obstacles. The residual stress in the sample is connected closely with the long distance stress generated from the dislocation behavior. At 3 T, the residual stress arrived at the minimum of 16 MPa. Moreover, in the presence of a magnetic field, the common η(MgZn₂) in the grain boundary dissolved and moved to internal grains because of the concentration difference, which helped to enhance the tensile strength and toughness of the materials. Finally, the fracture morphology was analyzed by scanning electronic microscopy. The fracture characteristic matches with the plasticity property.

1. Introduction
7xxx aluminum alloy exhibits the advanced characteristics of high strength, together with high toughness, good corrosion resistance and welding properties, which have been widely used in the fields of transport, weapons, aerospace and aviation, etc [1]. In the 1980s, based on the available 7150 aluminum alloy, the American Alcoa Company explored the 7055 aluminum alloy by increasing the Zn/Mg ratio and Cu component, and decreasing the impurities (Fe, Si). The microstructure, thereafter, has transformed from a saturated solid solution to an over-saturated solid solution [2]. Amounts of reinforced particulates precipitated at the grain boundary, which were beneficial to the enhancement of strength and intergranular corrosion resistance. As a result of the pinning effect of the precipitated particles, the dislocation mobility became worse and the plasticity of the materials was lowered [3]. It caused problems to the further plastic deformation processing in the form of rolling, squeezing or forging. In this paper, based on the magnetoplasticity effect (MPE) [4], the high pulsed magnetic field was applied in the plastic deformation processing, that is the tensile test, to investigate the influence of a magnetic field on the plasticity of 7055 alloy.

Recently, with the development of electromagnetic techniques, research on the electromagnetic processing of materials has been developed rapidly. As an important approach, the high magnetic field was introduced into materials processing by more and more scientists. In the 1970s, Kravchenko first found that the magnetic field influenced the viscosity of electronic clouds and the dislocation mobility and plasticity [5]. In 1987, Al’shits and coworkers reported that the B < 1 T static magnetic field would impel the dislocation movement in the NaCl single crystal [6]. In 1997, Li et al noted that the magnetic field would accelerate the launch, multiplication and
movement of dislocation in an Fe–Ni alloy [7]. It was argued that it was not the possible result of the magnetostriction effect. In the same year, Molotskii explained the MPE in the view of dislocation depinning due to an electronic spin conversion [8]. With the increase of the dislocation average length, the plasticity of materials would be enhanced thereafter. Until now, the available reports on the MPE mainly focused on the ionic crystal as NaCl, KBr and so on. The experimental results concerning the nonmagnetic metallic materials were rarely reported. In this paper, the 7055 aluminum alloy was selected to study the influence of a high pulsed magnetic field on plasticity in a tensile test.

2. Experimental and procedure

The commercial 7055 aluminum alloy was processed as a standard tensile sample whose size is shown in figure 1, illustrating the schematic of the experimental apparatus. The tensile tests were processed on a DNS10 electronic tension machine in the presence of a pulsed magnetic field with 0 T, 1 T, 3 T, 5 T, 7 T magnetic induction intensity (B). Under the condition of a fixed 0.5 mm min⁻¹ extension rate, the total time from beginning to fracture was about 13 min. In the first 10 min, the sample was exposed to a magnetic field. In total, 30 pulses were exerted on the sample with an interval of 20 s. In order to prevent the stress on the common steel-made clap by magnetic forces, a special nonmagnetic stainless steel clap was designed and utilized.

JEOL-JSM-7001F scanning electronic microscopy and JEM-2100 transmission electronic microscopy (TEM) were used separately to observe the fracture morphology, precipitates and dislocation characteristic, including morphology and distribution. The TEM sample was cut out from a position slightly above the fracture position and polished to 20 μm, followed by ionic reduction. Furthermore, in order to investigate the phase transition during the procedure, the ab initio method was introduced to study the bonding process of Mg and Zn atoms to form the MgZn₂ phase.

3. Results and discussion

3.1. Tensile properties

Figure 2 shows the stress–strain curves of the alloy tested at different B. It can be seen that the magnetic field influences the tensile properties of the material when B changes.

Figure 3 indicates the tendency of the tensile strength and elongation of the 7055 alloy tested at different B. The results reveal that when B is not more than 3 T, the tensile strength and elongation will increase with an increase in B. It demonstrates that the magnetic field has had a positive effect on the tensile properties. At B = 3 T, the tensile strength and elongation arrives at the maximum 610 MPa and 9.0% synchronously, which have been enhanced by 7.9% and 20% compared to the relevant 576.5 MPa, 7.5% of the initial sample without any magnetic field treatment. When B exceeds 3 T, the magnetic field takes a negative effect. In particular, when B is equal to 7 T, the tensile strength and elongation are 532 MPa and 6.5%, which are decreased by 5.8% and 13.3% individually compared to that of the untreated sample.
3.2. Mechanism of the magnetic field

The plasticity of the 7055 aluminum crystals with a large number of defects is controlled by the pinning and depinning of dislocations from the obstacles created by these defects. The defects may be roughly subdivided into two groups of strong and weak obstacles that are listed in Table 1, where \( \mu \) is the shear modulus, \( b \) is the Burgers vector, and \( l \) is the spacing. The parameter is the ‘thermal flow strength’ that means the shear strength in the absence of thermal energy. It reflects not only the strength but also the density and arrangement of the obstacles. Based on the definition standard in Table 1, in the 7055 alloy, the main obstacles belong to the medium ones because the sizes of the precipitates are small.

The start of a dislocation motion is controlled by its release from the obstacles. This becomes reasonable under the action of a strong enough mechanical stress. More importantly, the dislocation mobility is the key factor to determine its moving characteristic. When the tensile test is performed under a magnetic field, the
influence of the magnetic field on the dislocation, namely, the magnetoplasticity, will take effect. Figure 4 illustrates the dislocation movement in one period in the presence of a magnetic field and external stress, where the parallelogram plane represents the sliding plane and the shadow represents the region that has slid.

The whole process can be divided into four main steps, as per figures 4(a) ~ (d).

Figure 4(a) (step 1) shows the initial state of dislocation. Under the condition of external stress, the dislocation is free from the obstacles and moves forward along the sliding direction indicated by the arrow. The required characteristic time is $10^{-3} \sim 10^{-8}$ s. The accurate time is determined by the distance ($L$) between the adjacent obstacles.

Step 2 is the most complicated one during the whole process, which is displayed in figures 4(b), (c) and (f). During this period, the dislocation is close to the next obstacle. The relationship between the dislocation and obstacles are determined by $L$, which is relevant to the radical pair state. When the $L$ is larger than $L^*$ (about $10^{-9}$ m), which is the critical length to distinguish the state of the radical pair, the moving dislocation will pass the S, T resonance area where the electron spin directions are random (figure 4(e)). In the presence of a magnetic field, when the $L$ is smaller than $L^*$, the free electrons will be stimulated between the dislocation and obstacle (figure 4(b)). Two free electrons will generate some new radical pairs. The required time to form a radical pair is $10^{-14} \sim 10^{-6}$ s. The transitory time implies that the free electron stimulation and radical pair formation will be completed instantaneously. Under the Δg mechanism, the radical pairs were impelled to transform from the S state to T0 by an external magnetic field. Further, the magnetic field will influence the electron spin and induce the atomic rearrangement, which directly results in the spin lattice relaxation. Under this condition, the radical pair will transit from a T0 to T+, T- state (figure 4(f)) [9]. The characteristic time for the atomic arrangement is

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**Figure 4.** The schematic of the dislocation movement in the presence of a magnetic field. (a) shows surmounting the previous obstacle and moving forward; (b) displays the motions close to the obstacles, in the cases of $L > L^*$ (e) and $L < L^*$, (f), with $L$ being the distance between dislocation and obstacles, and $L^*$ the critical one; (c) refers to the stay at the obstacles; (d) indicates the surmounting the obstacles and moving forward.
$10^{-12} \sim 10^{-8}$ s. In view of the energy difference, the S state of the radical pair implies a higher bonding energy between the dislocation and obstacle when compared to the arbitrary T0, T+, T− state. Therefore, a high coverage of radical pairs at the T state will contribute to the enhancement of the plasticity of the material, which can be achieved in the presence of a magnetic field. The experimental phenomenon is referred to as the MPE. Apparently, the analysis and discussion about MPE is in the quantum scale.

As shown in figure 4(c) (step 3), the dislocation is hindered and stays at the obstacles. Though the radical pair is at the T state with lower bonding energy, the necessary impulsive energy is still needed. The possible energy resource comes from the stress or, sometimes, heat energy. The relevant characteristic time of this period is $10^{-5}$ s to $\infty$. The meaning of $\sim \infty$ means that if there is not enough energy to stimulate the dislocation movement, the dislocation will stay at the obstacles for long periods.

In figure 4(d), when the critical demanded energy is absorbed, the dislocation will depin from the obstacle and move forward. The required time is $10^{-5} \sim 10^{-10}$ s. One period of dislocation movement will be terminated.

Compared to the characteristic time of the four steps, it can be concluded that step 2, which includes the electron stimulation and atomic arrangement, can end up in momentary time, which meanwhile demonstrates the high efficiency of the magnetic field treatment. Nevertheless, the delay of the dislocations at the obstacle in step 3 is the most time-consuming period. It is regularly the rate-limiting step when performing the tensile test.

After making clear the details of the dislocation movement in the synchronous presence of stress and the magnetic field, we move onto the discussion of critical $B$ that is marked as $B^*$. Not only the experimental results but the theoretical derivation demonstrates that there is a $B^*$ when the MPE functions [10, 11]. The implication of $B^*$ means when $B$ is less than $B^*$ the MPE will have a positive effect with the increase of $B$; while in the range of $B > B^*$, the MPE will have a negative effect, which implies that with the enhancement of $B$, the MPE will be weakened. The derivation of $B^*$ is performed as follows:

The density of the radical pair ($D$) in the magnetic field is determined by the density matrix as:

$$\frac{\partial D}{\partial t} = \frac{1}{\hbar} i |H_{av}, D| - \frac{D}{t}$$  \hspace{1cm} (1)

where, $i$ denotes the vector unit of the matrix, $t$ denotes the dislocation movement, $\hbar$ is the Planck constant and $H_{av}$ is the Hamiltonian function. The energy transform ($E_M$) from the S to T0 state can be calculated by formula (2):

$$E_M = \frac{1}{2} \Delta g \mu_B H$$  \hspace{1cm} (2)

Among it, the $\Delta g$ stands for the g factor difference of the radical pair, which is often adopted as $10^{-3}$ [12]. $\mu_B$ and $H$ denote the Bohr magneton and magnetic field intensity respectively.

There are, in total, four kinds of irrelevant radical pair states between the dislocation and obstacles, so the initial condition of the density matrix ($D$) can be assumed to be:

$$D_S(0) = D_{T0}(0) = D_{T+}(0) = D_{T-}(0) = \frac{1}{4}$$  \hspace{1cm} (3)

The dislocation will move when driven by the unordered field of internal stress. The time function is randomly distributed when the radical pair passes through the S, T resonance region. According to the Possio time distribution, in the presence of the $H$ magnetic field, the average density matrix at the S state, that is $D_S(H)$, can be expressed by:

$$D_S(H) = \frac{1}{t_0} \int_0^\infty D_S(t) \exp\left(-\frac{t}{t_0}\right) dt$$  \hspace{1cm} (4)

Among it, $t_0$ stands for the average time for the radical pair to pass through the S, T resonance region. It is analyzed that $D_S(H)$ is the Laplace transition of $D_S(t)$ whose expression in linear style is:

$$D_S(H) = \frac{1}{4} \left(1 + \frac{T_1}{T_0}\right) \left(1 + \frac{T_2}{T_0}\right) + \frac{H^2}{\Delta g^2 \mu_B^2}$$  \hspace{1cm} (5)

where, the $T_1, T_2$ parameters stand for the spin relaxation time along the longitudinal and tangential directions respectively. As for the spin lattice relaxation in metallic material, the $T_1, T_2$ are often adopted as $2.4 \times 10^{-9}$ s [13]. Ulteriorly, the critical magnetic field intensity $H_m$, is calculated as:

$$H_m = \frac{\hbar}{\Delta g \mu_B \sqrt{T_1 + T_2}}$$  \hspace{1cm} (6)
\[ h = 6.63 \times 10^{-34} \text{ J} \cdot \text{s}, \Delta g \approx 10^{-3}, \mu_B = 9.27 \times 10^{-24} \text{ J} \cdot \text{T}^{-1}, T_1 = T_2 \approx 2.4 \times 10^{-9} \text{ s} \]

are substituted, the computed \( H_m \) is \( 2.4 \times 10^6 \text{ A m}^{-1} \) corresponding to \( B^* = 3 \text{ T} \). The derived result matches with the experimental phenomenon.

### 3.3. Phase transition under a magnetic field

Figure 5 shows the precipitates' characteristics at the grain boundary in the 0 T and 3 T samples. (a) 0 T sample; (b) the enlarged image of (a); (c) 3 T sample; (d) the enlarged image of (c).

Figure 5. Precipitates at the grain boundary in the 0 T and 3 T samples. (a) 0 T sample; (b) the enlarged image of (a); (c) 3 T sample; (d) the enlarged image of (c).

In figure 5(a) of the 0 T sample, the main precipitates at the grain boundary exhibit as particles with a size of 40 ~ 80 nm and a rod-like morphology. The high-resolution TEM and Fourier transformation analysis show that it is an \( \eta \) phase on the basis of an incoherent interface between the precipitate and matrix (figure 5(b)) [14]. The crystal grain is small and has a rod-like shape. According to particle phase analysis, it is considered as the \( \eta' \) phase. In the 3 T sample in figure 5(c), the precipitates exhibit a slender-rod morphology and there is a precipitate-free zone around the grain boundaries. The common \( \eta \) (MgZn2) in the grain boundary dissolves and moves to the internal grains because of the concentration difference, and the crystalline \( \eta' \) phase increases. The \( \eta' \) phase has a clear coherent relationship with the matrix. Because the components of the Mg, Zn elements in the \( \eta' \) phase are not exactly determined, the selected diffraction cannot be marked definitely. Moreover, the \( \eta' \) size falls in the range of 10 ~ 20 nm length and several nanometer thickness.

Figure 6 presents the precipitates in the grain internals in the 0 T and 3 T samples. The proportion of the slender-rod \( \eta' \) phase in figure 6(b) is obviously higher than that in figure 6(a).
Although the $\eta'$ (MgZn$_2$) and $\eta$ (MgZn$_2$) phases have the same elements of Mg and Zn, they have a different crystal structure. For the $\eta$ phase, the lattice parameters are $a = 0.515$ nm, $c = 0.86$ nm with the P6$_1$/mmc space group. It is an overaging phase produced in the heat treatment. It is beneficial to increase the stiffness and strength but detrimental to the plasticity of the material [15]. As for the metastable $\eta'$ phase, it is often regarded as the evolution product from the GP area and sometimes directly from the matrix. The $\eta'$ is half coherent with the matrix. The lattice parameters are $a = 0.496$ nm, $c = 1.403$ nm. According to the available reports, the higher volume fraction of $\eta'$ will benefit from improving the plasticity of the materials [16].

In order to discern the effect of a magnetic field on the $\eta'$ and $\eta$ characteristics, the first principle was used to investigate the bonding process of the Mg, Zn atom in the absence and presence of a magnetic field. Figure 7 presents the crystal structure of MgZn$_2$. According to the location diversity, the Zn atoms are distinguished as Zn$_1$ (at the boundary position) and Zn$_2$ (at the internal location).

Figure 8 shows the density of states (DOS) of the spin polarization, shortened to spin DOS, which implies the exposure of electrons to the magnetic field. According to the first principle theory, only in the areas where the spin DOS peak overlaps and the magnetic moment is similar can the bonding form completely [17].

As shown in figures 8(a) and (b), in the weak bonding region of $-10 \sim -6$ eV, the spin DOS of the Zn$_1$ 4p orbit overlaps with the Mg 3s, 3p orbits to a much greater extent compared with the other bonding regions, together with a tiny magnetic momentum difference. In the $2 \sim 6$ eV (antibonding region), the spin DOS of Zn$_1$
Figure 8. Spin DOS of Zn₁, Zn₂, and Mg atoms. (a) Summarized Zn₁ 4p and Mg 3s, 3p; (b) spin DOS of Zn₁ and Mg atoms; (c) spin DOS of Zn₂ and Mg atoms.

Figure 9. The dependence of residual stress on $B$. 

[Graph showing the dependence of residual stress on $B$.]
4p also overlaps with the Mg 3p orbit with a small magnetic momentum difference. It is regarded that after spin polarization, the bonding occurs mostly in the weak bonding regions. In the presence of a magnetic field, the Zn1 4p orbit can still form a covalent bond with the Mg 3s, 3p orbit, but the stability is lowered. In figure 8(c), in the strong bonding region of $-6 \sim 0$ eV, the spin DOS of the Zn$_2$ 4p orbit is different from that of the Mg 3s, 3p together with a big magnetic momentum difference. In the antibonding region of $0 \sim 8$ eV, the DOS curve of each Zn$_2$ orbit seldom overlaps with the Mg 3s orbit. So in the presence of a magnetic field, the covalent bond formation condition will be weakened between the Zn$_2$ 4p and Mg 3s orbits. Therefore, the newly generated MgZn$_2$ phase tends to be at a metastable state, which is the $\eta'$ phase.

3.4. Effect of a magnetic field on the residual stress

Figure 9 demonstrates the relationship between $B$ and residual stress, that is, with the enhancement of $B$, the residual stress decreases at first, and is then followed by the increasing tendency. At $B = 3$ T the minimum residual stress is 16 MPa. Compared with the initial sample, the residual stress is lowered as a whole after the magnetic field treatment.

In the sample, the dislocation distributes uniformly, which will induce a long distance internal stress. A higher internal stress can act as the driving force to impel the dislocation movement. The activated energy $U^*$ under the magnetic field can be calculated by the following formula:

![Figure 10. The classical distribution of dislocations in the crystal.](image)

![Figure 11. The possible mechanisms of dislocation density balance between the sparse and dense areas. (a) Initial state indicating the sparse and dense regions; (b) the three possible mechanisms.](image)
where $\sigma_M$ is the magnetic stress, $b$ is the Burgers vector, $l$ is the average dislocation length, $d$ is the dislocation width that is at $10^{-10}$ m order of magnitudes, and $\beta$ is a structural factor. In particular, $l$ denotes the momentum of dislocation in the magnetic field. The mass magnitude is identical to that of an atom of $10^{-27}$ kg and the moving velocity of dislocation in the magnetic field is $10^{-3}$ m s$^{-1}$ [18]. $H$ means the foreign magnetic field intensity, whose unit is A/m. Just as previously derived, the optimized value of $H$ is $2.4 \times 10^6$ A m$^{-1}$. Hence, the calculated $U^*$ is in the $10^5$ eV order of magnitude. It is the energy that comes from the internal stress in the presence of a magnetic field accompanied by the characteristic of long distance.

On the opposite aspect, the necessary energy for moving dislocations to surmount obstacles includes two parts. One is the activated energy for dislocation ($U_a$), which is at $0.1 \sim 1$ eV. The other is the interchange energy.

$$U^* = \sigma_M bld + \beta IH$$

(7)

Figure 12. The fracture morphology of the sample tested under different $B$. (a) $B = 0$T; (b) $B = 1$T; (c) $B = 3$T; (d) the enlarged image of $B = 3$T; (e) $B = 5$T; (f) $B = 7$T.

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between the dislocation and obstacles \((U_{\text{ex}})\) at 1 eV. So the long distance internal stress due to the magnetic field can increase the dislocation mobility. Meanwhile, the dislocation movement will relax the long distance stress. It can obviously be seen that there is an interaction between the long distance stress and the dislocation movement.

To clarify the influence of the dislocation distribution on the microstructure, a dislocation cell model is established to discuss the evolution. Based on the fact that dislocation distributes uniformly, the regions are defined as sparse and dense regions, as shown in figure 10. Moreover, the sparse areas can be thought of as the passage in the dislocation cell structure, while the dense areas are the cell walls [19].

When the sample is exposed to the foreign stresses and magnetic field, two possible situations will occur. If the sparse area received more power, the dislocation source in the sparse region will move and multiply (mechanism ①). The dislocation in the dense region will move towards the sparse region, which is attributed to increase the dislocation density in the sparse area. If the dense area received more power, a reaction between the dislocations will happen and induce some of the dislocations to disappear (mechanism ②). In other words, the foreign energetic resources take the effect of balancing the dislocation density in sparse and dense areas through three possible mechanisms. The process has been illustrated in figure 11.

Therefore the dislocation distribution tends to be uniform, which is beneficial in decreasing the internal stress that is just the residual stress in macroscopic view. The residual stress has been decreased gradually when \(B\) increases from 0–3 T. At \(B = 3\) T the residual stress is only 16 MPa, which implies that the internal stress has been released to a large extent [20].

When \(B\) increases further, more dislocations will depin from the weak obstacles and move along the slide slip until they meet the obstacles with higher energy, such as grain boundaries. The dislocation cluster will take place. The unbalanced sparse and dense area will appear again. Since the dislocation movement velocity is rapid, the clustering at the grain boundary is serious. The density difference between the sparse and dense regions will become enormous. Sometimes, it is more serious than in the initial sample. Obviously, it will do harm to the material properties.

3.5. Fracture analysis

Figure 12 presents the fracture morphology of the samples subjected to different \(B\). In the initial sample (figure 12(a)), it exhibits a combined fracture of cleavage and dimple, where the dark grey areas imply the fiber existence. Around the cleavage platform there are a number of small dimples [21]. In the 1 T sample (figure 12(b)), the tearing lip can be seen together with the disappearance of the cleavage. The characteristic ductile fracture becomes obvious: the coverage of big dimples is larger than that of the initial sample. The crack in the local regions reveals that the brittle fracture still exists in the sample, but the coverage of the brittle fracture is much lower than that in the initial sample. It should be noted that in the 3 T sample (figure 12(c)), both the cleavage and brittle cracks do not exist. There are several little peaks around the dimple, which is enlarged as figure 12(d). The ‘peak’ microstructure demonstrates that the ductile fracture proceeds sufficiently, which corresponds to a higher elongation of the 3 T sample. As for the \(B = 5\) T sample (figure 12(e)), the dimples are small. Compared with figures 12(b) and (c), the fracture surface is relatively flat. The plastic deformation does not prevail fully, which is relevant with a poorer plasticity. At \(B = 7\) T, as shown in figure 12(f), the cleavage and tearing lip appearance can be seen again. It implies that the plasticity of the materials worsens.

4. Conclusion

The tensile properties of the 7055 aluminum alloy were tested in the presence of a high pulsed magnetic field with different magnetic induction intensities of 1 T, 3 T, 5 T and 7 T. There is a threshold of magnetic field treatment. When \(B\) is less than 3 T, the effect of the magnetic field on the tensile properties of the 7055 alloy is positive. While \(B\) exceeds 3 T, the effect becomes negative. When \(B\) is equal to 3 T, tensile strength and elongation are 610 MPa and 9%, which have been enhanced by 7.9% and 20% separately compared to those of the initial sample. From the TEM photographs, it can be seen that the magnetic field has induced the phase transition from \(\eta' (\text{MgZn}_2)\) in the grain boundary to \(\eta (\text{MgZn}_2)\) of the internal grains. More coverage of the \(\eta'\) phase is beneficial to the improvement of tensile properties due to its coherency with the matrix. The residual stress in the 3 T sample is minimum, which is explained by the dislocation movement, mobility and distribution due to the MPE. Furthermore, the fracture morphology matches with the plasticity of the 7055 aluminium alloy subjected to magnetic field treatment.
Acknowledgments

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