Crystal structure, Curie temperature, and electromagnetic absorption properties in Fe$_x$Co$_{30}$Ni$_{60-x}$Si$_5$Al$_5$ ($x = 30, 35, 40, 45$) high entropy alloys

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Abstract

In order to meet the requirements of future national defense for high temperature electromagnetic (EM) absorbing performance, a series of Fe$_x$Co$_{30}$Ni$_{60-x}$Si$_5$Al$_5$ ($x = 30, 35, 40, 45$) high-entropy alloys (HEAs) powders was prepared and their Curie temperatures ($T_C$) were measured by a self-made Wheatstone bridge. According to the results, varying the Fe/Ni ratio affected the crystal structure, Curie temperature, oxidation resistance, and electromagnetic absorbing properties of the above compounds. Since Fe has a BCC structure and is thus easier to form the solid solutions with Si and Al, the crystal structure of the alloy has changed from FCC toward BCC with increasing Fe dopant content. In turn, the Curie temperature ($T_C$) decreased from 473.68 °C to 358.07 °C, being lower than their initial oxidation temperature ($>800$ °C). The reflection losses (RL) of powders at room temperature and high temperatures ($\leq$500 °C) were calculated as well. It was found that the flake powders after ball milling gained a larger aspect ratio, resulting in the better absorption effect, which was due to high toughness and low strength characteristics of the initial FCC structure. Furthermore, the permittivity and permeability of alloys upon heating reached impedance matching at a certain temperature, thus achieving the greater $R_{L_{\text{max}}}$ value. Finally, the high-temperature EM absorption characteristics of HEAs were shown to merit a thorough study.

1. Introduction

In recent years, electromagnetic (EM) pollution has attracted increasing attention due to the rapid development of electronic information technology. In that regard, the microwave absorbing properties of composites, metals, nanomaterials, and silicon carbides have been extensively studied to elucidate their potential in real applications [1–4] as well as to ensure that the materials meet the stringent requirements for social production, human life, and national security. In particular, state-of-the-art EM absorbing materials for national defense are forced to operate under extreme environments such as high temperature and corrosion. However, most of them are unable to withstand harsh environmental conditions. Therefore, the development of new absorbing materials that would fulfill the formulated demands has become a hot topic [5, 6]. Traditional EM absorbents have shortcomings such as low Curie temperature ($T_C$) and poor oxidation resistance at high temperature [7]. When the temperature rises to a near-$T_C$ value, the EM properties will deteriorate rapidly and the structure will change from ferromagnetic to paramagnetic. Moreover, long-term high-temperature exposure of the material without any dense protective oxide layer will cause its complete oxidation until failure [8].

Over the past years, high-entropy alloys (HEAs) have attracted much attention because of their outstanding mechanical properties, magnetic characteristics, and thermal stability, especially high temperature resistance. Since the HEAs consist of five or more elements with the same or close atomic ratio, they possess very high mixing entropy with randomly disordered atoms [9–11]. This allows one to produce the disordered FCC-BCC...
solid-solution phases [12–14]. Therefore, the HEAs exhibit a unique ‘cocktail structure’ because of various characteristics of alloy elements and the interactions between them, which endows the material with high TC and oxidation resistance simultaneously [15].

High entropy alloys are usually designed by doping with other elements at maintaining equal proportions of Fe, Co, and Ni. For instance, Zhang et al [16, 17] prepared FeCoNiSi1-x, FeCoNiAlx, and FeCoNi(SiAl)x compounds to explore the influence of Al and Si. The experimental results revealed that an increase in Al content led to changes in the crystal structure of alloys from FCC to BCC. In turn, adding Si enabled one to form a new compound Ni3Si with deteriorated magnetic properties. On the other hand, the best balance between magnetic, electrical, and mechanical properties was found in a FeCoNi(AlSi)x system. As another example, Duan et al [18] produced a FeCoNiSi1-xAl alloy and showed that Si dissolved in the FCC and BCC phases due to its small atomic radius. This led to the upshift in the main angular peak positions in XRD of both phases as the Si content increased. Furthermore, the phase composition of the alloys was reported to be closely related to a Si dopant concentration, which could improve their microwave absorbing properties via controlling phase transitions.

Besides, a comprehensive analysis of the microwave absorption properties of (FeCo)xNi(100−x) by Peng et al [19] revealed that introducing Ni into a FeCo structure allowed one to significantly enhance the degree of crystallinity and flake-like particles in the sample. Moreover, the higher was the Ni content, the better were the complex permeability and permittivity of the alloys.

Even though the room-temperature magnetic and EM properties of HEAs have been systematically studied to date, there are still few reports about TC and EM absorbing properties of these materials at high temperatures. So in this work, the effects of different Fe/Ni ratios on these parameters of FeCoNiSi1-x, Si5Al5 alloys were systematically investigated. In particular, varying the proportion of Fe and Ni elements in these systems was shown to effectively tune their TC and EM characteristics through the formation of new phases.
2. Methods

2.1. Materials and fabrication
Fe$_x$Co$_{30}$Ni$_{60-x}$Si$_5$Al$_5$ (x = 30, 35, 40, 45) alloys were prepared via induction melting the constituent elements of 99.9% purity in a high vacuum environment. The ingots were melted three to four times for homogeneity. Ribbons were then produced from the ingots using a single wheel melt spinning technique. For this, the ingots were cut into small pieces and put in a quartz tube with an orifice diameter of 0.8 mm for melt spinning under argon pressure (≤10 psi); the rotational speed of the copper wheel was 2500 rpm. The ribbons were denoted as R30, R35, R40, R45, respectively. In order to reduce the hardness of the alloy and release the internal strain generated during the melting process, ribbons were annealed at 750 °C for an hour in an argon atmosphere with 5% hydrogen additive. The ribbons were then milled into alloy powders (powder-to-ball-mass ratios were 1: 40) for 2 h in a Mixer Mill unit. The powders were labeled P30, P35, P40, P45, respectively.

2.2. Characterization and measurements
The morphology of the powders was characterized using a field-emission scanning electron microscope (FESEM, JSM-7600F, JEOL) equipped with an energy-dispersive x-ray spectrometer (EDS, NORAN SYSTEM 7, Thermo Scientific). The microstructure of the specimens was probed via x-ray diffraction (XRD); for this the measurements were conducted on a XRD-7000 Shimadzu diffractometer in Co K$_\alpha$ radiation ($\lambda = 1.5406$ Å) within a 2θ angular range from 20° to 90°. A vibrating-sample magnetometer (VSM, Riken, Japan) was employed to investigate the magnetic properties such as saturation magnetization ($M_s$) and coercive force ($H_c$) under a maximum applied field of 10 000 Oe at room temperature. The EM parameters ($\varepsilon_\prime$, $\varepsilon_\prime\prime$, $\mu_\prime$, and $\mu_\prime\prime$) of the specimens were assessed by an Agilent vector network analyzer (Agilent N5230) in the 2–12 GHz range. Prior to trials, the powders were dispersed in paraffins with a mass ratio of 6: 10, and then pressed into rings with outer diameters of 7.0 mm, inner diameters of 3.04 mm and thicknesses of 3–4 mm. The thermal stability of alloys was tested via thermogravimetric analysis (TGA) and differential scanning calorimetry (DSC, STA 449F3, NETZSCH). The Curie temperatures were measured using a self-assembled Wheatstone bridge.

3. Results and discussion

3.1. Microstructure analysis
Figure 1 shows the XRD patterns of the ribbons and powders of Fe$_x$Co$_{30}$Ni$_{60-x}$Si$_5$Al$_5$ composition. In both cases, solid solution structures were formed instead of intermetallic compounds. Hence the intensities of the strongest peak of (110) for BCC and (111) for FCC are denoted as $I_{110}$BCC and $I_{111}$FCC, respectively. According to figure 1(a), the ratio of $I_{110}$BCC/$I_{111}$FCC increased, suggesting that the FCC-to-BCC phase transition took place with the increase of Fe content. The experimental results are basically consistent with the conclusion that the structure of FeCoNi alloy is directly related to its electron concentration (c/a) [20]. The ability of metal atoms to form a solid solution mainly depends on their crystal structure and electronic balance, especially on their atomic radius. This is due to the fact that different atomic radii may lead to disordered atomic occupation. Since Fe has a BCC structure and Ni is characterized by a FCC configuration, the crystal structure of the alloy will change from FCC to BCC with the increase of Fe concentration. Meanwhile, Al is also an element which is easy to form a BCC phase as reported by Tan et al [21]. As seen in figure 1(a), there is a high peak at 50.973° for R30 and R40, suggesting that excessive cooling rate during the strip processing results in the preferred orientation which means the crystal growth in a specific direction. As shown in figure 1(b), after one hour of heat treatment at 750 °C and ball milling, this phenomenon disappeared because the crystal faces became averaged. While R30 and R35 specimens underwent the transition from FCC + BCC to FCC phase, R40 changed its structure from FCC + BCC to BCC. After annealing at 750 °C, the elements were fully diffused and the phase transition occurred towards the lowest Gibbs automatic energy. Figure 1(c) displays the peak height, main diffraction peak position, grain size, and internal stress for four kinds of powders. The grain size and internal stress are calculated by the Williamson-Hall equations: $\frac{\beta \cos \theta}{\lambda} = \frac{K}{D} + \frac{4 \varepsilon \sin \theta}{\lambda}$, where $\lambda$ is the FWHM of the peaks, $K = 0.9$, D is the grain size, and $\varepsilon$ is the internal stress. Then, $\frac{\beta \cos \theta}{\lambda}$ and $\frac{4 \varepsilon \sin \theta}{\lambda}$ were defined as the y-axis and x-axis, respectively. The average and grain size and the internal stress can be obtained from the slope of the line and the intercept of the Y axis [22]. The left half in the graph corresponds to FCC phase with a lower Fe content and the right half is associated with BCC phase with elevated Fe concentration. The results show that, even though the difference between these two crystal phases, the dopant-induced changes observed in the above parameters are quite similar. With the increase of Fe content, the peak height decreases, the main diffraction peaks of FCC and BCC move toward the lower angles, the grain size increases, and the internal stress decreases. Since a BCC phase has a smaller density than FCC, each interstitial void in the BCC structure is larger than that in FCC. Moreover, while entering into a solid solution structure, the Al elements with larger atomic radius will cause noticeable
lattice distortions [16]. As a result, the main XRD peaks of both phases downshift and the internal stress of the crystal increases.

3.2. Morphological analysis

From the study of the microwave magnetic properties of the lamellar particles with low magnetocrystalline anisotropy and high shape factor, it can be seen that the increase of the lamellar size is the most important factor. In particular, the aspect ratio of particles can improve the effective complex permeability and microwave absorbing properties of composites [23]. Figures 2(a)–(d), (f)–(i) show the back-scattered images of the Fe$_{30}$Co$_{30}$Ni$_{30-x}$Si$_5$Al$_x$ powders, figure 2(e) Schematic diagram of ball milling and figure 2(j) EDS analysis of ball milling residue. As shown in figures 2(a)–(d), the ribbons after heat treatment and ball milling become flake particles with large aspect ratio, whose thicknesses can be estimated from figures 2(f)–(i). As the increase of Fe dopant concentration promotes the development of BBC phase and makes its plasticity worse, flake particles instead of micropowders are observed. The formation of flakes in the course of ball milling is illustrated in figure 2(e), showing how the small ball with a huge kinetic energy makes the ribbon fold and crack (as seen in figures 2(b) and (d)), thus turning into flake particles. The milling process runs until the ball reaches its

Figure 2. (a)–(d) Backscattered electron SEM images of Fe$_{30}$Co$_{30}$Ni$_{30-x}$Si$_5$Al$_x$ powders. (e) Ball milling diagram. (f)–(i) Backscattered electron SEM images of Fe$_{30}$Co$_{30}$Ni$_{30-x}$Si$_5$Al$_x$ powder particles; (j) EDS spectrum acquired on the red-highlighted particle in image.
Due to the milling time, milling ratio and un-screening, all the four composite powders are composed of large and uneven particles. Since FCC phase has better toughness and worse hardness than BCC, therefore, the particles become thicker and smaller. Among the above samples, Fe30Co30Ni30Si5Al5 is found to have the largest aspect ratio that can enhance the surface polarization. It is worth noting that flake powder particles have raised particles as marked with red in figure 2(i). After the EDS analysis, this is left by the surface wear of ZrO2 ball during ball milling. Table 1 summarizes the EDS results acquired on four composites. It is seen that the prepared powders are similar to the designed compositions. At the same time, the presence of acceptable amounts of oxygen is due to the influence of the non-inert gas environment during melting, ball milling, and drying.

| EDS results for alloy powders (at%) | Fe  | Co  | Ni  | Si  | Al  | O  |
|-----------------------------------|-----|-----|-----|-----|-----|----|
| Fe30Co30Ni30Si5Al5                | 27.16 | 28.64 | 28.78 | 5.79 | 6.14 | 3.50 |
| Fe35Co30Ni25Si5Al5                | 33.94 | 31.08 | 22.64 | 4.29 | 3.25 | 4.80 |
| Fe40Co30Ni20Si5Al5                | 39.38 | 30.73 | 18.01 | 4.81 | 5.45 | 1.61 |
| Fe45Co30Ni15Si5Al5                | 44.48 | 28.25 | 12.08 | 5.50 | 5.56 | 4.14 |

Figure 3. The diagram of an experimental circuit principle.

Figure 4. The curie temperature (TC) of Fe_xCo_{30}Ni_{60-x}Si_5Al_5 samples.
3.3. Curie temperature ($T_C$) and oxidation temperature

$T_C$ is measured by self-made Wheatstone bridge and the principle of an experimental circuit is shown in figure 3. In such a construction, a signal generator sends out the excitation power and a pair of self-made coil inductances is connected with a pair of precision resistances according to the Wheatstone principle. The change of bridge voltage is measured by a precision AC voltmeter. A self-made inductor is a pair of a 30 $\times$ 20 $\times$ 5 mm sample and a ceramic. The metal wires are sheathed in the high-temperature bushing to prevent short circuit. When the temperature rises above $T_C$, the inductance produced by the magnetic properties of the sample disappears and the AC bridge is out of balance and the output voltage signal $V_0$ decreases rapidly which is measured by a precision AC ammeter. In this experiment, the real-time voltage and temperature are respectively connected with the computer through a precision AC ammeter (DMM7510, KEITHLEY) and thermometer (HWP-C804, Green Instrument) with communication protocol, and a visual window is established by LabVIEW2018 to record the change of voltage with temperature in real time. Figure 4 displays the Curie temperatures of the four components. According to the XRD results in figure 1(a), the Curie temperature of Fe-doped alloys decreased from 473.68 °C to 358.07 °C as the crystal structure transformed from FCC to a BCC phase. This is due to the fact that Al and Si are non-magnetic elements and their solid solubility in the BCC state is higher than that in the FCC phase.

![Figure 5](https://example.com/figure5.png)

**Figure 5.** TG/DSC curves of Fe$_x$Co$_{30}$Ni$_{60}$-$x$Si$_5$Al$_5$ ribbons.

![Figure 6](https://example.com/figure6.png)

**Figure 6.** Hysteresis loops for (a) different alloy ribbons and (b) related powders. The changes of MS and HC are shown in the insets.
Figure 5 shows the TG/DSC curves of the Fe$_{x}$Co$_{30}$Ni$_{60-x}$Si$_{5}$Al$_{5}$ ribbons, where the exothermic peak (>1200 °C) represents the oxidation reaction and the endothermic peak (>1300 °C) reflects the melting temperature ($T_m$) in the DSC curve. According to the TG curves, the initial oxidation temperature of the samples is greater than 800 °C, which is much higher than the $T_c$ value of the samples. A lack of weight gain after 1200 °C indicates that the samples are completely oxidized. This is because the large specific surface area of the ribbons makes it impossible for Al and Si to form a dense oxide film on the sample surface.

3.4. Magnetic properties

Static magnetic properties play an important role in the relevant dynamic magnetic characteristics. In this study, the static magnetic properties of the Fe$_{60-x}$Co$_{30}$Ni$_{x}$Si$_{5}$Al$_{5}$ powders are investigated at room temperature under an applied field of $-10000 \text{Oe} < H < 10000 \text{Oe}$. Figure 6 displays the magnetic hysteresis loops of the ribbons and powders, as well as saturation magnetization ($M_s$) and magnetic coercivity ($H_c$) are presented in the inset. As reported by Zhang et al [21], the $M_s$ value of the BCC phase is higher than that in the FCC phase. Given the above mentioned, $M_s$ rises with the increase of Fe content in both cases. In turn, the $H_c$ values of the powders are larger to those of the ribbons. It is noteworthy $H_c$ is related to the crystal structure, grain size, and defects [22, 24]. Moreover, this parameter in the FCC state is lower than that in the BCC phase. The $H_c$ can be evaluated according to the formula [18]:

$$H_c^{\sigma} = \frac{(1 - V_{am})^2 K_1 D^6}{\mu_0 M_s A^3}$$

where $V_{am}$ is the volume of the amorphous phase, $K_1$ is the magnetocrystalline anisotropy constant, $D$ is the grain size, and $A$ is the exchange coupling constant. Although the change of $H_c$ in figure 6(b) is not completely consistent with that in figure 1(c), it is related to the internal stress. The larger ball milling ratio and vibration frequency make the ball have the greater kinetic energy, which leads to the more pronounced internal stress of the powder in the course of high-energy ball milling. In turn, the increase of internal stress hinders the magnetic domain motion, increasing the $H_c$ value.

3.5. Dynamic magnetic properties

EM wave absorption performance is usually evaluated through relative complex permittivity ($\varepsilon_r = \varepsilon' - j\varepsilon''$) and relative complex permeability ($\mu_r = \mu' - j\mu''$). The storage and loss of electric and magnetic energies are
therefore represented by real and imaginary parts, respectively. In turn, the EM wave loss is divided into conduction loss, dielectric loss, and electromagnetic loss.

Figure 7 shows the complex permittivity and permeability of the four samples. As seen in figure 7(a), the alloys have the resonance peaks at 6.98, 6.97, 5.49, and 8.58 GHz, respectively, and their \( \varepsilon' \) values decrease with increasing frequency. At the same time (figure 7(b)), the \( \varepsilon'' \) parameter increases with frequency, evidencing the resonance peaks at 8.32, 8.15, 6.73, and 9.33 GHz with a hysteresis behavior. The permittivity properties of P30, P35, P40 and P45 can be explained via Debye relaxation:

\[
\varepsilon' = \frac{\sigma}{\omega^2 \varepsilon_0} + \varepsilon_\infty \tag{2}
\]

\[
\varepsilon'' = \sigma / \omega \varepsilon_0 \omega = 2 \pi f \tag{3}
\]

According to the formula 2 and 3, the \( \varepsilon' \) and \( \varepsilon'' \) have a positive correlation with the conductivity (\( \sigma \)). Combined with the EDS results in table 1, P30, P35 and P45 have more oxygen content and form more oxides than P40, which leads to their lower conductivity. The low permittivity of P30, P35 and P45 specimens is due to the formation of a resistance layer on the surface of the particles, which weakens the initially high permittivity. The lowest permittivity of P45 alloy is owing to the fact that Si and Al are more easily dissolved in the BCC phase, thus decreasing the conductivity and, respectively, the \( \varepsilon' \) value [9].

According to figure 7(c), the \( \mu' \) value increases with the increase of Fe content due to the rising amount of the BCC phase, but decreases with rising frequency. Besides, the easy dissolution of Si and Al elements in the BCC phase will reduce the magnetocrystalline anisotropy constant (\( K_1 \)) and magnetostrictive coefficient constant (\( \lambda_3 \)), as can be concluded from the formula below [25]:

\[
\mu_1 \approx \frac{\mu_0 M_S^2}{K_1 + \frac{3}{2} \lambda_3 \sigma \delta / d \beta^2 / \beta} \tag{4}
\]

where \( \sigma \), \( \beta \), \( \delta \), and \( d \) are the internal strain, the volume concentration of impurity, the domain wall thickness, and the impurity diameter, respectively. As found from figure 7(d), the \( \mu' \) parameter exhibits the same change trend as \( \mu'' \), decreasing with the increase of frequency. The correlation between \( \mu'' \), \( \mu' \) and \( \sigma \) can be established through the eddy current loss as follows:
The reflection loss (RL) of Fe₃₀Co₃₀Ni₃₀Si₅Al₅ powders can be calculated from the measured relative complex permittivity (\(\varepsilon_r = \varepsilon' - j\varepsilon''\)) and permeability (\(\mu_r = \mu' - j\mu''\)) values at a given frequency according to transmission line theory expressions below:

\[
\mu'' = 3\pi\mu_0(\mu')^2d^2f
\]  \(\text{(5)}\)

The reflection loss (RL) of Fe₃₀Co₃₀Ni₆₀Si₅Al₅ powders can be calculated from the measured relative complex permittivity (\(\varepsilon_r = \varepsilon' - j\varepsilon''\)) and permeability (\(\mu_r = \mu' - j\mu''\)) values at a given frequency according to transmission line theory expressions below:

\[
RL(dB) = 20 \log \left( \frac{Z_{in} - Z_0}{Z_{in} + Z_0} \right)
\]  \(\text{(6)}\)

\[
Z_{in} = Z_0\left(\frac{\mu_r}{\varepsilon_r}\right)^{1/2}\tanh\left(\frac{2\pi fd}{c}\right)^{1/2}(\mu_r\varepsilon_r)^{1/2}
\]  \(\text{(7)}\)

Here, \(Z_{in}\) is the impedance of the composites, \(Z_0\) represents the intrinsic impedance of free space, \(d\) is the thickness of absorber, \(f\) is the incident electromagnetic wave frequency, \(c\) is the velocity of light, and \(\mu_r\) and \(\varepsilon_r\) are the complex permeability and complex permittivity, respectively. As can be seen from the above formula 6 and 7, the absorption of electromagnetic waves depends on their electromagnetic parameters, thickness, and frequency. Figure 8 displays the RL calculated at the thickness \(d\) ranging from 1 to 4 mm. According to the plots, when the crystal structure is FCC, the overall absorption performance is better than that of BCC. P₃₀ alloy has the maximum RL with a thickness of 2.3 mm can reach \(-47.03\,\text{dB}\) at \(9.60\,\text{GHz}\). The worst absorption effect, following from \(RL_{\text{Max}}\) of \(-18.65\,\text{dB}\) at \(1.13\,\text{GHz}\), is achieved in P₄₀ sample with a thickness of 4 mm because of the mismatch between permittivity and permeability.

It is worth mentioning that formula 4 can be simplified as \(\mu \propto M_r^2\) if only the influence of temperature on \(M_r\) is considered. In this respect, one can calculate the \(\mu^2-T\) curve by using the curve obtained in figure 4. Similarly, formula 5 can also be regarded as \(\varepsilon'' \propto \mu'^2\).

\[
\varepsilon'_r = A(1) + A(2)T + A(3)T^2
\]  \(\text{(8)}\)

\[
\varepsilon''_r = B(1) + B(2)T^2 + B(3)T^3 + B(4)T^4
\]  \(\text{(9)}\)

The coefficients in formulae 8 and 9 are determined by Li's report [26] in order to calculate the high temperature properties of dielectric constants. Finally, the RL values of the four materials with thicknesses of 3 mm within a temperature range of 150°C to 500°C are found using formulae 6 and 7. The results are displayed in figure 9. Obviously, comparing with figure 8 that the RL_{\text{Max}} values of P₃₀, P₃₅, and P₄₀ at 150°C are lower than those obtained at the same frequency at room temperature. Because of the increase of dielectric with rising
temperature, the frequency of the RL\textsubscript{Max} for P45 alloy decreases from 7.5 to 7.15 GHz and the value of RL\textsubscript{Max} decreases from $-16.62$ to $-21.52$ dB. As a result, the permeability-to-dielectric constant ratio becomes more suitable. It is noteworthy that the RL\textsubscript{Max} of the four materials decreases with the increase of temperature until the T\textsubscript{C} value is reached, moving toward higher frequencies.

4. Conclusions

In this study, a series of Fe\textsubscript{x}Co\textsubscript{30}Ni\textsubscript{60–x}Si\textsubscript{5}Al\textsubscript{5} HEA powders was prepared by induction melting and a single wheel melt spinning technique. Their crystal structure transformation, T\textsubscript{C}, magnetic properties, and EM absorbing properties at high temperature and room temperature were systematically studies. HEAs tended to form disordered solid solutions, hence the change of their crystal state mainly depended on the original crystal structure of Fe and Ni atoms. An increase in Fe content was found to induce the FCC-BCC structural transition in the alloys, exerting influence on the non-magnetic Si and Al elements as well. Since Si and Al tended to exist in BCC rather than in FCC phase, the T\textsubscript{C} value of alloys decreased with increasing Fe dopant concentration. Furthermore, the coercivity H\textsubscript{C} and saturation magnetization M\textsubscript{s} were showed to be related to internal stress and crystal structure of the compounds, increasing with rising amount of BCC phase, respectively. At the same time, the FCC structure exhibits better tensile strength and ductility. In this respect, flaky particles with higher aspect ratio could be obtained via ball milling, resulting in the maximum absorption peak in a Fe\textsubscript{30}Co\textsubscript{30}Ni\textsubscript{30}Si\textsubscript{5}Al\textsubscript{5} system at room temperature. The microwave absorbing performance of the alloys at high temperature was calculated. In particular, one could successfully achieve the better impedance matching between the permittivity and permeability at a certain temperature. The RL\textsubscript{Max} of Fe\textsubscript{45}Co\textsubscript{30}Ni\textsubscript{15}Si\textsubscript{5}Al\textsubscript{5} at 7.15 GHz and 150 °C was found to be $-21.52$ dB, exceeding that at 7.5 GHz at room temperature ($-16.62$ dB). The best EM absorbing ability and the highest T\textsubscript{C} value have been enriched in a Fe\textsubscript{30}Co\textsubscript{30}Ni\textsubscript{30}Si\textsubscript{5}Al\textsubscript{5} compound, which undoubtedly merits a further study. To conclude, the adequate study of microwave absorbing properties at high temperature needs to take into consideration that the dielectric constant increases with temperature and the matching between the dielectric constant and the permeability is dynamic.

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Data availability statement

All data that support the findings of this study are included within the article (and any supplementary files).

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