1. Introduction

Recently, the development of advanced materials with high specific strength (strength-to-density ratio) is strongly demanded in the automotive industry because the reduction of the weight of transport vehicles leads to the improvement of fuel efficiency and the reduction of CO2 exhaust. The demand for a reduction of weight has been met by the thinning of steels made possible by their high strengthening. Advanced high strength steels such as ausforming steels,1) maraging steels,2) transformation-induced plasticity (TRIP)3) steels, low alloyed TRIP steels 4) and high manganese austenitic steels showing the TRIP effect and the twin-induced plasticity (TWIP) effect5,6) have been developed. Another method of increasing the specific strength of steels is to lower the density by adding light elements such as Al and Si to Fe-based alloys is mentioned. Although Fe–Al alloys have attracted interest as low density steels, Al is a strong α(bcc)-stabilizer element of Fe and the maximum solubility of Al in the γ(fcc) phase of Fe is less than 1.0 mass%. Moreover, in the α phase of Fe–Al alloy, the ordering from bcc(A2) to B2 and DO 3 structures occurs with increasing Al content, which causes brittleness of an α Fe–Al alloy with a high Al content. Widening the γ phase region to the Al-rich side by the addition of Mn and C to Fe–Al alloys is very useful to improve their ductility. It is known that Fe–Mn–Al–C alloys with the γ phase are the basic system for austenitic stainless steels without Ni and Cr,7,17) cryogenic materials8–21) and magnetic alloys.22) Fe–Mn–Al–C-based alloys show high strength and excellent ductility at room temperature. In addition, it has been reported that Fe–Mn–Al–C-based alloys also have excellent high-temperature oxidation resistance28) and that the addition of Cr to Fe–Mn–Al–C alloys is effective for enhancing oxidation resistance.29) Although there have been extensive studies regarding Fe–Mn–Al–C-based alloys, the degree to which low density can be obtained in Fe–Mn–Al–C-based alloys while maintaining high strength is not yet clear. In this study, in order to clarify the compositional and microstructural dependences of the specific strength of Fe–20Mn–Al–C(–Cr) alloys, the density of Fe–20Mn–Al–C(–Cr) alloys showed a higher specific strength than conventional steels.

2. Experimental Procedures

Fe–20Mn–(10–14)Al–(0–1.8)C (mass%) quaternary and Fe–20Mn–(10–14)Al–(0.75–1.8)C–5Cr (mass%) quinary alloys were prepared by melting electrolytic Fe (99.95%), electrolytic Mn (99.9%), Al (99.7%), electrolytic Cr (99.2%) and graphite in a high frequency induction furnace under an argon atmosphere, the addition of Cr being expected to enhance the corrosion resistance and oxidation resistance. The obtained ingots were hot-rolled into sheets with a thickness of about 1.5 mm at 1 200°C. The hot-rolled sheets were cut into small pieces for heat treatment at various temperatures. Dumbbell-shaped specimens for tensile testing were prepared by wire-cut electro-discharge machining. The obtained samples were heat treated in the temperature range of 900–1 100°C for 15 min, and then quenched in water or air cooled to room temperature. The cold-workability was evaluated using the cold-
workability parameter $W$ which was defined as $W = (t_0 - t_{\text{min}})/t_0 \times 100$, where $t_0$ is the initial thickness and $t_{\text{min}}$ is the minimum thickness before a crack appeared during cold-rolling at room temperature. The microstructures were observed by optical microscopy (OM) and transmission electron microscopy (TEM). The hardness of the specimens was determined by Vickers hardness measurement at room temperature. The mechanical properties were investigated by tensile testing at room temperature at a strain rate of 0.5 mm/min using physically polished dumbbell-shaped specimens with gauge dimensions of 25 mm × 5 mm × 1 mm.

3. **Estimation of Density**

In this study, a simple method to estimate the density of Fe–Mn–Al–C(–Cr) γ(fcc) alloys was proposed. By defining the density of the fcc structure of each pure element as a density parameter, the density of Fe–Mn–Al–C(–Cr) alloy with fcc phase $D_{\text{fcc}}$ was estimated using the following equation:

$$D_{\text{fcc}} \text{(g/cm}^3\text{)} = \sum x_i D_i \text{fcc},$$

where $D_{\text{fcc}}$ is the density parameter of pure element $i$ with the fcc structure and $x_i$ is the atomic fraction of pure element $i$. The elements Fe, Mn, Cr and C do not have the fcc structure at room temperature. Therefore, the density parameters in Fe, Mn and Cr with the fcc structure at room temperature were estimated from the lattice constants of fcc Fe, fcc Mn and fcc Cr evaluated by extrapolating from the lattice constants in Fe–Mn\textsuperscript{20,31} and Cr–Pt\textsuperscript{32} binary fcc alloys. The density parameter of C was evaluated from the lattice constant of pure C estimated by extrapolating from the lattice constant in Fe–C\textsuperscript{33} binary fcc alloy, although C atoms are located at interstitial sites in the Fe–C binary alloy. The density parameters used in this study were $D_{\text{Fe fcc}} = 8.19$, $D_{\text{Mn fcc}} = 7.19$, $D_{\text{Al fcc}} = 2.70$, $D_{\text{Cr fcc}} = 6.36$ and $D_{\text{C fcc}} = 2.85$. It was confirmed that the calculated value well agrees with the experimental value measured by Archimedes method, e.g., in Fe–20Mn–10Al–1.5Cr–1.5C (mass%) alloy, the calculated and observed values were 6.64 and 6.67 g/cm\textsuperscript{3}, respectively.

**Figure 1** shows iso-$D_{\text{fcc}}$ contour lines calculated from Eq. (1) on the experimentally determined phase diagram of the Fe–20Mn–Al–C quaternary system at 1100°C\textsuperscript{34}, where the κ phase is the ternary carbide Fe\textsubscript{3}AlC\textsubscript{x} with perovskite structure and $x$ is round 0.66\textsuperscript{35}. According to Fig. 1, γ single phase alloys with a low density below 7.0 g/cm\textsuperscript{3} can be obtained by controlling Al and C contents. The estimated iso-$D_{\text{fcc}}$ lines were linearly extended to the other phases as indicated in Fig. 1. The density of Fe–40Mn–14Al–1C alloy with the γ + α two-phase structure calculated from Eq. (1) was 6.31 g/cm\textsuperscript{3}, which is slightly different from the observed value of 6.13 g/cm\textsuperscript{3}. Therefore, the density of Fe–Mn–Al–C(–Cr) alloys other than γ single phase alloys can also be estimated from Eq. (1).

4. **Results and Discussion**

4.1. **Compositional Dependence of Mechanical Properties in Water-quenched Alloys**

**Figures 2(a) and 2(b)** show the dependence of Al and C contents on hardness and cold-workability in Fe–20Mn–Al–C and Fe–20Mn–Al–C–5Cr alloys, respectively, the specimens being annealed at 1100°C for 15 min, followed...
by water quenching. The iso-

A

D

A

fcc contour lines calculated

with Eq. (1) are depicted in the both figures, where the con-

stituent phases of each alloy were estimated from OM ob-

servation.

Figures 3(a)–3(d) shows typical OM microstruc-

tures at room temperature for (a) Fe–20Mn–10Al–0.5C quaternary alloy, (b) Fe–20Mn–11Al–1.8C–5Cr, (c) Fe–

20Mn–13Al–1.8C–5Cr and (d) Fe–20Mn–13Al–1C–5Cr quinary alloys. Hereafter, the alloys identified as

g single

phase and

\( g / H \) \( a \) two phases from the OM observation were

specified as

g alloys and

\( g / H \) \( a \) alloys, respectively. The ex-

perimentally determined equilibrium phase diagram of

Fe–20Mn–Al–C quaternary alloy at 1 100°C34) and the cal-

culated equilibrium phase diagram of Fe–20Mn–Al–C–5Cr quinary alloy at 1 100°C by the CALPHAD method36) are

also depicted in Figs. 2(a) and 2(b), respectively. It is sug-

gested from the phase diagrams that the

\( g / H \) \( a \) and the

\( g / H \) \( a \) phase boundaries shift toward the high-C side and

the \( g / H \) \( a \) two-phase region widens to the high-Al side by

the addition of Cr. The results of the OM observation in the

Fe–20Mn–Al–C–5Cr quinary alloys are in a good agree-

ment with the calculated phase diagram.

The tensile properties such as yield strength defined as

0.2% proof strength, ultimate tensile strength and elonga-

tion in all specimens are listed in Table 1, where the calcu-

lated \( D^A_{\text{fcc}} \) of all specimens are also indicated. Figures 4(a) and 4(b) show the typical stress–strain curves in Fe–Mn–

Al–C quaternary and Fe–Mn–Al–C–

Cr alloys annealed at 1 100°C for 15 min followed by quenching in water.

Table 1. Hardness, density \( D^A_{\text{fcc}} \), yield strength, tensile strength, elongation and specific strength in Fe–Mn–Al–C and Fe–Mn–Al–C–

Cr alloys annealed at 1 100°C for 15 min followed by quenching in water.

<table>
<thead>
<tr>
<th>Alloy (mass %)</th>
<th>Hardness (HV)</th>
<th>Density ( D^A_{\text{fcc}} ) (g/cm³)</th>
<th>Yield strength (MPa)</th>
<th>Tensile strength (MPa)</th>
<th>Elongation (%)</th>
<th>Specific strength (MPa·cm³/g)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Fe20Mn10Al0.25C</td>
<td>325</td>
<td>6.94</td>
<td>600</td>
<td>771</td>
<td>10.1</td>
<td>111</td>
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<tr>
<td>Fe20Mn10Al0.5C</td>
<td>270</td>
<td>6.89</td>
<td>552</td>
<td>808</td>
<td>29.4</td>
<td>117</td>
</tr>
<tr>
<td>Fe20Mn10Al1.0C</td>
<td>295</td>
<td>6.80</td>
<td>530</td>
<td>843</td>
<td>59.1</td>
<td>124</td>
</tr>
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<td>Fe20Mn10Al1.2C</td>
<td>302</td>
<td>6.77</td>
<td>636</td>
<td>881</td>
<td>38.1</td>
<td>130</td>
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<tr>
<td>Fe20Mn10Al1.5C</td>
<td>466</td>
<td>6.72</td>
<td>900</td>
<td>1092</td>
<td>17.1</td>
<td>163</td>
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<tr>
<td>Fe20Mn11Al1.0C</td>
<td>300</td>
<td>6.72</td>
<td>840</td>
<td>946</td>
<td>35.7</td>
<td>141</td>
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<tr>
<td>Fe20Mn11Al1.8C</td>
<td>413</td>
<td>6.59</td>
<td>1240</td>
<td>1251</td>
<td>0.2</td>
<td>190</td>
</tr>
<tr>
<td>Fe20Mn13Al0.5C</td>
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<td>6.63</td>
<td>720</td>
<td>926</td>
<td>1.1</td>
<td>140</td>
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<tr>
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<td>6.59</td>
<td>860</td>
<td>1054</td>
<td>5.2</td>
<td>160</td>
</tr>
<tr>
<td>Fe20Mn13Al1.5C</td>
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<td>6.64</td>
<td>710</td>
<td>934</td>
<td>65.6</td>
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</tr>
<tr>
<td>Fe20Mn13Al1.8C</td>
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<td>6.51</td>
<td>1136</td>
<td>1192</td>
<td>29.0</td>
<td>183</td>
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<tr>
<td>Fe20Mn12Al1.0C</td>
<td>351</td>
<td>6.56</td>
<td>880</td>
<td>1000</td>
<td>32.2</td>
<td>152</td>
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<tr>
<td>Fe20Mn12Al1.8C</td>
<td>442</td>
<td>6.43</td>
<td>1000</td>
<td>1148</td>
<td>1.6</td>
<td>179</td>
</tr>
<tr>
<td>Fe20Mn13Al1.0C</td>
<td>384</td>
<td>6.47</td>
<td>880</td>
<td>1136</td>
<td>28.6</td>
<td>176</td>
</tr>
<tr>
<td>Fe20Mn13Al1.3C</td>
<td>425</td>
<td>6.43</td>
<td>915</td>
<td>1140</td>
<td>22.2</td>
<td>177</td>
</tr>
<tr>
<td>Fe20Mn13Al1.8C</td>
<td>457</td>
<td>6.35</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>Fe20Mn14Al1.0C</td>
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<td>6.39</td>
<td>-</td>
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<td>-</td>
<td>-</td>
</tr>
<tr>
<td>Fe20Mn14Al1.8C</td>
<td>500</td>
<td>6.28</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
</tr>
</tbody>
</table>

It is seen from the results of the Fe–20Mn–10Al–

(0.25–1.0)C alloys that cold-workability and tensile elonga-

tion drastically increase with increasing C content because of the decrease in the volume fraction of the \( \alpha \) phase. The quaternary \( g \) alloy showed a hardness of over 295 Hv with

excellent cold-workability. It is noteworthy that the Fe–

20Mn–10Al–1.5C \( g \) alloy with a hardness of 466 Hv showed a high yield strength of 850 MPa and an upper yield point followed by almost constant flow stress. The Fe–

20Mn–11Al–1.8C alloy showed a high hardness of 413 Hv but was very brittle. This is due to the formation of a cer-

tain amount of \( \gamma \) phase, which stably exists at the annealing

temperature. The hardness and the yield and tensile strengths of \( \gamma + \alpha \) alloys increased with increasing Al and C

contents. The quaternary \( \gamma + \alpha \) alloys with a high Al con-

tent of 13 mass% showed high strength but had very low

tensile elongation because of a large \( \alpha \) volume fraction.

In the Fe–20Mn–Al–C–5Cr quinary alloys, \( g \) alloys with

both high hardness and good cold-workability can be ob-

tained in alloys with higher Al and C contents compared to the quaternary alloys without Cr. The Fe–20Mn–11Al–

1.8C–5Cr \( g \) alloy possessing the microstructure shown in
Fig. 3(b) exhibited a high hardness of 431 Hv with a good cold-workability of 71%. This means that \(\kappa\)-carbide is suppressed by the addition of Cr. In addition, the Fe–20Mn–11Al–1.8C–5Cr alloy with \(D^{\kappa} = 6.51\) g/cm\(^3\) exhibited a high specific strength of over 180 MPa·cm\(^3\)/g with good tensile elongation of around 30%. It was confirmed by X-ray measurement that there was no stress-induced martensite phase in the tensile deformed specimen.\(^{39}\) The reason why the Fe–20Mn–11Al–1.8C–5Cr \(\gamma\) alloy with the microstructure shown in Fig. 3(c) showed high hardness, yield strength and tensile strength without the large loss in elongation should be due to the precipitation of a very fine \(\kappa\) phase during quenching. Ishida \textit{et al.} have reported that in Fe–20Mn–10.5Al–1.2C alloy, the precipitation of a very fine \(\kappa\) phase occurs during quenching from temperatures above 900°C.\(^{34}\) It has also been reported that aging of Fe–Mn–Al–C alloys with a \(\gamma\) phase produces the precipitation of the \(\kappa\) phase, which leads to age hardening.\(^{16,37–39}\) Therefore, in this study, the dependence of the cooling rate after the annealing treatment on the microstructure and the mechanical properties in the Fe–20Mn–10Al–1.5C–5Cr \(\gamma\) alloy was investigated.

### 4.2. Dependence of Cooling Rate on Tensile Properties

Figure 5(a) shows the stress–strain curves obtained in the Fe–20Mn–10Al–1.5C–5Cr \(\gamma\) alloy quenched in water and air cooled to room temperature from the annealing temperature of 1100°C, where the cooling rates of the water quenching and the air cooling were about 1100°C/s and about 15°C/s, respectively. The air-cooled specimen had a much higher yield strength than the water-quenched specimen and showed an upper yield point followed by almost constant flow stress. Such stress–strain behaviors are quite similar to those of the Fe–20Mn–10Al–1.5C quaternary and the Fe–20Mn–11Al–1.8C–5Cr quinary \(\gamma\) alloys quenched in water, as shown in Fig. 4. Figure 5(b) shows the true stress–strain curves of Fig. 5(a). The water-quenched specimen showed an almost linear work-hardening, while the air-cooled specimen showed an almost zero work-hardening rate at a small plastic strain region below 5% and then work-hardening with increasing plastic strain.

Figures 6(a) and 6(b) show the diffraction patterns taken from the water-quenched and the air-cooled Fe–20Mn–10Al–1.5C–5Cr alloys, respectively. Although both patterns reveal extra spots between the fundamental reflections, which should indicate the formation of the \(\kappa\) phase with the perovskite structure, the intensity of the superlattice reflections in the air-cooled specimen is much higher than that in the water-quenched specimen. Figures 6(c) and 6(d) show dark field images of the (100) \(\kappa\) superlattice reflection in the water-quenched and air-cooled Fe–20Mn–10Al–1.5C–5Cr alloys, respectively. In the water-quenched specimen, very fine \(\kappa\) phase precipitates are slightly confirmed, while the cuboidal \(\kappa\) phase precipitates aligned in the \(001\) directions are clearly observed in the air-cooled specimen. The microstructure observed in the air-cooled Fe–20Mn–10Al–1.5C–5Cr alloy is very similar to the morphology observed in Ni-based superalloys with the \(\gamma + \gamma'\) two-phase. Therefore, it is concluded that high hardness and strength, as well as a small work-hardening rate obtained in the water-quenched Fe–Mn–Al–C(–Cr) alloys with a high C content are due to the nano-size pre-
cipitation of $\kappa$ phase during cooling from the annealing temperature. The formation mechanism of very fine $\kappa$ precipitates during cooling is not clear at present, even though one could possibly attribute it to precipitation through spinodal decomposition from the point of view of the crystal structure similarity between the $\gamma$ (fcc) and the $\kappa$ (perovskite) phases. It has been reported that composition fluctuation due to spinodal decomposition precedes the precipitation of the $\kappa$-carbide in aged Fe–(30–33)Mn–(8–9)Al–(0.9–1.3)C alloys.\textsuperscript{16,40}

4.3. Tensile Properties by Precipitation Hardening and Work Hardening

As mentioned above, the hardness and strength of the present alloys can be enhanced by precipitation hardening due to the formation of a very fine $\kappa$ phase during cooling. Moreover, work hardening is commonly used to harden and strengthen a ductile material. Therefore, in this study, the effect of cooling rate from several annealing temperatures and cold rolling after heat treatment on the hardness and tensile properties was investigated.

The hardness and tensile properties of various specimens annealed at 1100°C followed by air cooling to room temperature are listed in Table 2. In all specimens, the hardness and strength increased by air cooling from the annealing temperature, which is due to the precipitation of the $\kappa$ phase during cooling. In the Fe–20Mn–13Al–1.3C–5Cr $\gamma$+$\alpha$ alloy, a high specific strength of about 200 MPa·cm$^3$/g was obtained by air cooling. The hardness and tensile properties of Fe–20Mn–10Al–1.5C–5Cr $\gamma$ alloys annealed

![Stress-strain curves](image1)

**Fig. 5.** (a) Stress–strain curves in Fe–20Mn–10Al–5Cr–1.5C alloys annealed at 1100°C for 15 min followed by water quenching (WQ) and air cooling (AC) to room temperature. (b) True stress–strain curves of (a).

![Diffraction patterns](image2)

**Fig. 6.** Diffraction patterns taken from (a) water-quenched and (b) air-cooled Fe–20Mn–10Al–1.5C–5Cr alloys, and dark field images of the (100)$_\kappa$ superlattice reflection in (c) water-quenched and (d) air-cooled specimens.

<table>
<thead>
<tr>
<th>Alloy (mass %)</th>
<th>Hardness (Hv)</th>
<th>Density $D_A^{fcc}$ (g/cm$^3$)</th>
<th>Yield strength (MPa)</th>
<th>Tensile strength (MPa)</th>
<th>Elongation (%)</th>
<th>Specific strength (MPa·cm$^3$/g)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Fe20Mn-10Al-1.0C</td>
<td>425</td>
<td>6.80</td>
<td>630</td>
<td>794</td>
<td>42.0</td>
<td>117</td>
</tr>
<tr>
<td>Fe20Mn-10Al-1.2C</td>
<td>438</td>
<td>6.77</td>
<td>730</td>
<td>873</td>
<td>18.2</td>
<td>129</td>
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<tr>
<td>Fe20Mn-10Al-1.5C</td>
<td>517</td>
<td>6.72</td>
<td>1120</td>
<td>1216</td>
<td>18.9</td>
<td>181</td>
</tr>
<tr>
<td>Fe20Mn-10Al-1.5C-5Cr</td>
<td>382</td>
<td>6.64</td>
<td>1005</td>
<td>1060</td>
<td>40.4</td>
<td>160</td>
</tr>
<tr>
<td>Fe20Mn-11Al-1.8C-5Cr</td>
<td>427</td>
<td>6.51</td>
<td>1040</td>
<td>1223</td>
<td>41.0</td>
<td>188</td>
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<tr>
<td>Fe20Mn-13Al-1.0C-5Cr</td>
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<td>6.47</td>
<td>760</td>
<td>1197</td>
<td>7.3</td>
<td>185</td>
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<tr>
<td>Fe20Mn-13Al-1.3C-5Cr</td>
<td>454</td>
<td>6.43</td>
<td>920</td>
<td>1277</td>
<td>12.4</td>
<td>199</td>
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</table>

Table 2. Hardness, density $D_A^{fcc}$, yield strength, tensile strength, elongation and specific strength in Fe–Mn–Al–C and Fe–Mn–Al–C–Cr alloys annealed at 1100°C for 15 min followed by air cooling to room temperature.
at 1000°C and 900°C followed by water quenching (WQ) and air cooling (AC) are shown in Table 3. It is seen that the yield and tensile strengths increased and the elongation decreased with decreasing annealing temperature, which is considered to be due to the formation of \(\alpha\), \(\kappa\) and \(\text{Cr-carbide}\) phases with decreasing annealing temperature. Figure 7(a) shows the plots of the tensile strength as a function of the elongation in the water-quenched and air-cooled Fe–20Mn–10Al–1.5C–5Cr \(\gamma\) alloys. As expected, the yield and tensile strengths increased and the elongation decreased with decreasing annealing temperature, which is considered to be due to the formation of \(\alpha\), \(\kappa\) and \(\text{Cr-carbide}\) phases with decreasing annealing temperature. Figure 7(a) shows the plots of the tensile strength as a function of the elongation in the water-quenched and air-cooled Fe–20Mn–10Al–1.5C–5Cr \(\gamma\) alloys. As expected, the

### Table 3. Hardness, density \(D_{\text{fcc}}\), yield strength, tensile strength, elongation and specific strength in Fe–20Mn–10Al–1.5C–5Cr alloys.

<table>
<thead>
<tr>
<th>Annealing temperature (°C)</th>
<th>Cooling condition</th>
<th>Hardness (HV)</th>
<th>Density (D_{\text{fcc}}) (g/cm³)</th>
<th>Yield strength (MPa)</th>
<th>Tensile strength (MPa)</th>
<th>Elongation (%)</th>
<th>Specific strength (MPa·cm³/g)</th>
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</thead>
<tbody>
<tr>
<td>1000</td>
<td>WQ</td>
<td>315</td>
<td>6.64</td>
<td>840</td>
<td>1060</td>
<td>52.4</td>
<td>160</td>
</tr>
<tr>
<td></td>
<td>AC</td>
<td>366</td>
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<td>1080</td>
<td>1136</td>
<td>33.8</td>
<td>171</td>
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</table>

Table 4. Hardness, density \(D_{\text{fcc}}\), yield strength, tensile strength, elongation and specific strength in Fe–Mn–Al–C and Fe–Mn–Al–C–Cr alloys cold rolled at various reduction ratios after annealing at 1100°C for 15 min followed by quenching in water.

<table>
<thead>
<tr>
<th>Alloy (mass %)</th>
<th>Cold rolling ratio</th>
<th>Hardness (HV)</th>
<th>Density (D_{\text{fcc}}) (g/cm³)</th>
<th>Yield strength (MPa)</th>
<th>Tensile strength (MPa)</th>
<th>Elongation (%)</th>
<th>Specific strength (MPa·cm³/g)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Fe–20Mn–10Al–1.2C</td>
<td>94%</td>
<td>556</td>
<td>6.59</td>
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<td>1812</td>
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<td>275</td>
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<td>6.54</td>
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<td>1500</td>
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</tbody>
</table>

Figure 7. Plots of tensile strength vs. elongation in (a) water-quenched and air-cooled Fe–20Mn–10Al–1.5C–5Cr \(\gamma\) alloys and (b) as-quenched and cold-rolled Fe–20Mn–Al–C–(5Cr) alloys.

Figure 8 shows the relationship between specific strength and tensile elongation in various steels.

4.4. Comparison of Various Steels

Figure 8 shows the relationship between specific strength and tensile elongation in various steels, where the density of steels not previously reported was estimated by the same method as Eq. (1). In all steels, a tendency for strength increased by air cooling. Table 4 shows the hardness and tensile properties of specimens cold rolled at various reduction ratios after annealing at 1100°C followed by water quenching. Figure 7(b) shows the plots of the tensile strength as a function of the elongation in the as-quenched and cold-rolled Fe–20Mn–Al–C–(5Cr) alloys. In all specimens, the tensile strength as well as the hardness and yield strength drastically increased with cold rolling at room temperature. The 30% cold-rolled Fe–20Mn–11Al–1.8C–5Cr \(\gamma\) alloy showed specific strength over 230 MPa·cm³/g with elongation over 10%. Furthermore, the Fe–20Mn–10Al–1.5C and Fe–20Mn–10Al–1.5C–5Cr alloys cold rolled at reduction ratios more than 90% possessed a high hardness of over 650 Hv and a high specific strength of over 300 MPa·cm³/g, although elongation drastically decreased.
the tensile elongation to decrease with increasing specific strength can be observed. It is seen that the present Fe–Mn–Al–C(–Cr) alloys showed a higher specific strength than the conventional steels such as austenitic,\textsuperscript{11} ferritic,\textsuperscript{12} martensitic,\textsuperscript{13} precipitation hardened,\textsuperscript{14} dual-phase\textsuperscript{15} and super toughness steels.\textsuperscript{16} Moreover, the specific strength of the present alloys is higher than that of Fe–Mn–Si–Al TWIP steels.\textsuperscript{17} In addition, the present alloys also showed much higher yield strength than the TWIP steels.\textsuperscript{18} TRIP steels reported by Zakay et al.\textsuperscript{19} possess a much higher specific strength than the present alloys. However, the high strength of the TRIP steels can be obtained through a complicated thermomechanical process.\textsuperscript{20} On the other hand, the present alloys have an advantage in that both high specific strength and large tensile elongation can be obtained through simple processes controlling the annealing temperature and cooling rate.

5. Conclusion

The mechanical properties and microstructure of Fe–20Mn–Al–C quaternary and Fe–20Mn–Al–C–5Cr quinary alloys with a low density of less than 7.0 g/cm\textsuperscript{3} were investigated. The results obtained were as follows.

(1) The hardness, yield and tensile strengths of the γ alloys increased with increasing Al and C contents, while the tensile elongation decreased. The γ alloys with high Al and C contents showed a high yield strength and an upper yield point followed by almost constant flow stress. It was found that γ + α two-phase alloys also exhibited high strength, together with good ductility, by controlling the α volume fraction.

(2) Microstructural observation showed that the γ/γ′ + κ phase boundary shifted toward the high-C side suppressing the precipitation of κ carbide by the addition of Cr. It was also found that γ/γ′ + α phase boundary shifted toward the high-C side and that the γ + α two-phase region widened to the high-Al side by the addition of Cr.

(3) The Fe–20Mn–10Al–1.5C–5Cr γ alloy air cooled from the annealing temperature showed much higher hardness and strength than that quenched in water from the annealing temperature and exhibited an upper yield point followed by almost constant flow stress. TEM observation indicated that a very fine cuboidal κ phase was precipitated during cooling, which produced high hardness and strength of Fe–Mn–Al–C(–Cr) alloys.

(4) The Fe–20Mn–11Al–1.8C–5Cr alloy with a density of 6.51 g/cm\textsuperscript{3} showed a high specific strength of more than 180 MPa·cm\textsuperscript{3}/g with good tensile elongation of 40%. The present Fe–20Mn–Al–C(–Cr) alloys with low density showed a higher specific strength than the conventional steels.

(5) High specific strength of over 300 MPa·cm\textsuperscript{3}/g was obtained in the Fe–20Mn–10Al–1.5C(–5Cr) alloys by high reduction of cold rolling.

Acknowledgment

This study was supported by an ISIJ Research Promotion Grant and the Global COE Project.

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