Friction Stir Welding of High Phosphorus Weathering Steel—Weldabillities, Microstructural Evolution and Mechanical Properties

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Phosphorus (P) addition is expected to simultaneously increase the strength and corrosion resistance of weathering steels. However, P causes solidification cracking in the fusion welding process and reduces the toughness of the steel. To avoid these problems, the P content of weldable SMA490AW weathering steel is currently limited to below 0.035 mass%. High P steels which are impossible to be joined by the fusion welding process, can be joined by a solid-state joining process, i.e., friction stir welding (FSW). Because the stir zone obtained by FSW contained very fine grains, its toughness was expected to improve. This study applies FSW to high-P weathering steels and examines the weldability of the product. The microstructural evolution and mechanical properties of the stir zone were investigated at different welding temperatures. The macroscopic cross-sectional observations of the FSW joints revealed crack-free structures even in steel containing 0.3 mass% P. Moreover, FSW significantly refined the grain structure in the stir zone. Consequently, the ductile-to-brittle transition temperature of the stir zone was approximately 150°C lower in the steel containing 0.3 mass% P and welded below Ar (average grain size = 2.5 μm) than in the base material (average grain size = 23 μm). It appears that the grain refinement by FSW overcomes the embrittlement caused by excessive P content.

KEY WORDS: friction stir welding; weathering steel; phosphorus; microstructure; mechanical property; ductile-to-brittle transition temperature; stress-strain curve.

1. Introduction

Weathering steels containing alloying elements, such as Cu, P, Cr and Ni, exhibit excellent corrosion resistance under atmospheric conditions by forming a protective rust layer on their surface. These steels provide effective long-term anticorrosion performance without the need of painting and thus reduce the life-cycle cost of structures. Hence, they are widely used, mainly for bridges.

In weathering steels, P is a suitable alloying element because it improves weatherability and strength at minimal cost. Previous studies have reported that the weatherability of steel improves as the P content increases to at least 0.5 mass%.1–6) Moreover, the addition of a small amount of P effectively increases strength.7,8) However, P reduces the toughness of the base material (BM)9–16) as well as its weldability.17–21)

The negative effect of P on weldability is caused by solidification cracking, which is a form of high-temperature embrittlement that occurs due to lowering of the solidification temperature.21) Therefore, SMA490AW,22) a commercial weathering steel, contains less than 0.035 mass% P to ensure that it can be fusion-welded. SPAH23) is another example of weathering steel, which contains 0.1 mass% P; it exhibits excellent weatherability but poor performance in fusion welding.

To overcome the problems related to weldability, friction stir welding (FSW) can be employed. FSW is a solid-state joining process, in which materials are joined at a temperature below their melting point. This process is primarily used for aluminum alloys.24–27) Many studies on practical application of FSW for steels have been conducted.28–34) FSW does not involve solidification; hence, joining of steels with large amounts of P is possible.

Regarding the effect of P on toughness, many studies have reported that P segregation at the grain boundaries reduces
toughness.\textsuperscript{9-15} Yamaguchi\textsuperscript{15} explained that an increase in P segregation at the grain boundary reduces the cohesive energy of the grain boundary, thereby reducing the fracture stress at the boundary. In contrast, Suzuki et al.\textsuperscript{9} showed that in steels with approximately $\geq 0.01$ mass\% C, C first segregates at the grain boundaries, resulting in site competition between C and P; consequently, the segregation of P at the grain boundary is inhibited, which leads to suppression of grain boundary fractures. Weathering steels contain approximately 0.1 mass\% C; thus, grain boundary fracture is unlikely to occur if P content is suitably controlled and an appropriate fabrication process is employed. Nevertheless, the solute P within the grains generally hinders the gliding of dislocations due to a significant solid solution strengthening effect; thus, an increase in P content may causes cleavage fracture and reduce the toughness.

Grain refinement is effective for improving the toughness.\textsuperscript{35-39} For example, in ferritic steel, the ductile-to-brittle transition temperature (DBTT) decrease to 282\textdegree C when the grain diameter is refined from 20 $\mu$m to 1 $\mu$m.\textsuperscript{35} For grain refinement, in addition to conventional rolling technology,\textsuperscript{36,37} combinations of severe plastic deformation, such as equal-channel angular pressing (ECAP)\textsuperscript{38} and accumulative roll-bonding (ARB),\textsuperscript{14,39} and heat treatment have been proposed. During FSW, an extremely large shear strain is introduced together with heat generation, which induces dynamic recrystallization and leads to a significantly fine microstructure.\textsuperscript{31,32} Therefore, in high-P steels, FSW is expected to improve joint toughness through grain refinement.

In this study, weathering steels with 0.1–0.3 mass\% P, which cannot be fusion-welded, were produced, and the weldability of these steels using FSW was investigated. Furthermore, the microstructures and mechanical properties of the BMs and the stir zones (SZs) of FSW were evaluated, and the effects of P content and grain refinement on the mechanical properties of the steels were investigated.

## 2. Experimental Procedure

Table 1 lists the chemical compositions of the steels used in this study. SMA490AW is a commercial weathering steel with only 0.01 mass\% P and can be fusion-welded. The samples of 0.1P and 0.3P are the high P weathering steels fabricated in the laboratory. These samples were prepared using the following method. The ingots were produced using electrolytic iron (99.9\%), pure carbon (99.9\%), pure silicon (99.9\%), electrolytic manganese (99.9\%), ferrophosphorus (Fe with 20 mass\% P), pure copper (99.99\%), and electrolytic chromium (99\%). These metals were placed in an alumina crucible and melted in a high-frequency induction melting furnace. This melting process was performed in a 0.5-atm argon atmosphere. The final solidified sections of the ingots were cut and shaped into $\phi 30 \times 40$ mm pieces. Subsequently, hot rolling was performed nine times at 1 000\textdegree C until the plate thickness became 3.5–4.0 mm. This was followed by heat treatment at 1 000\textdegree C for 10 min and then air cooling. The hot rolling temperature of 1 000\textdegree C was higher than $A_3$ point of the 0.1P and 0.3P samples, which implies that the hot rolling was conducted in the austenite phase. Finally, the plates were ground into a 3 mm thickness to remove the surface oxide layer.

FSW was conducted on the fabricated plates. The welding tool was made of a WC-based material and equipped with a columnar probe without a thread. The shoulder diameter, probe diameter, and probe length were 15.0 mm, 6.0 mm and, 2.9 mm, respectively. Two types of welding conditions were employed: welding at a temperature above $A_1$ (rotation and welding speeds were 400 rpm and 150 mm/min, respectively) and below $A_1$ (rotation and welding speeds were 80 rpm and 150 mm/min, respectively). K-type thermocouples were placed on the bottom surface and at the centerline of the materials to measure the changes in temperature during FSW.

Microstructure observations and Vickers hardness tests were performed on the cross-section of the joints perpendicular to the welding direction. The microstructures were observed using a field-emission scanning electron microscope (FE–SEM) equipped with an electron backscattered diffractometer (EBSD). The specimens for the SEM observations were mechanically polished and etched using a solution containing 4% nitric acid and 96% ethanol. The samples for the EBSD measurement were mechanically polished and etched using a solution containing 4\% nitric acid and 96\% glacial acetic acid. The Vickers hardness profiles of the joints were measured at a load of 2.94 N for 15 s. Tensile and impact tests were performed to investigate the mechanical properties of the BMs and the SZs. The tensile tests were performed at a strain rate of $5 \times 10^{-3}$ s$^{-1}$. The length, width, and thickness of the gauge section were 5, 2, and 2 mm, respectively. The direction of the gauge section was perpendicular to the welding direction, as shown in Fig. 1(a). During the tensile test, the strain was measured using the digital image correlation (DIC) method, and the nominal strain was evaluated with an initial gauge length of 4.5 mm. The impact test was performed using a miniaturized Charpy impact machine (Tanaka, MI20T-D05KJ type) for a specimen with a length of 20 mm, width of 1 mm, and thickness of 1 mm (Fig. 1(b)). This specimen had a V-shaped notch with a depth of 0.2 mm and a root radius of 0.08 mm. The specimens were picked up such that the notch was at the joint center for the SZ, and at a position of at least 10 mm away from the joint center for the BM. The DBTT in this study was defined as

| Table 1. Chemical composition of base materials investigated in this study. |
|---------------------------------|-----|-----|-----|-----|-----|-----|-----|-----|-----|
| Material          | C    | Si   | Mn  | P   | S   | Cu  | Ni  | Cr  | Al  |
| SMA490AW         | 0.12 | 0.20 | 1.14 | 0.01 | 0.002 | 0.32 | 0.10 | 0.48 | –   |
| 0.1P             | 0.10 | 0.51 | 0.09 | 0.10 | <0.001 | 0.49 | 0.01 | 0.39 | 0.03 |
| 0.3P             | 0.08 | 0.50 | 0.09 | 0.27 | <0.001 | 0.49 | 0.01 | 0.40 | 0.02 |
the temperature at the boundary between the completely brittle and brittle–ductile transition because the influence of specimen size was negligible.

3. Results

3.1. Welding Temperature and Joint Appearance

Figure 2(a) shows the heat cycles during FSW under each welding condition, and Fig. 2(b) shows the pseudo-binary phase diagram of the Fe(0.1C-0.5Si-0.1Mn-0.5Cu-0.4Cr)-P steel. The maximum temperatures during the welding above the \( A_3 \) and below the \( A_1 \) temperatures were 1018°C (higher than the \( A_{e3} \) temperatures of the 0.1P (907°C) and 0.3P (985°C) samples) and 695°C (less than the \( A_{e1} \) temperatures of the 0.1P (739°C) and 0.3P (744°C) samples), respectively. These values indicate that each condition has been applied as intended.

Figure 3 shows cross-sectional photographs of the friction-stir-welded joints for all samples. All samples exhibited defect-free joints.

| Condition | Above \( A_3 \) | Below \( A_1 \) |
|-----------|----------------|----------------|
| 0.1P      | (a)            | (b)            |
| 0.3P      | (c)            | (d)            |

Fig. 1. Schematic of (a) sampling position of tensile test specimen and impact test specimen and (b) shape of impact test specimen.

Fig. 2. (a) Heat cycles obtained by thermocouples located on the bottom surface at the centerline for SMA490AW; (b) Fe(0.1C-0.5Si-0.1Mn-0.5Cu-0.4Cr)-P pseudo-binary phase diagram evaluated by thermodynamic calculation, using the Thermo-Calc. TCFE9.

Fig. 3. Photographs showing the cross section of FSW joints for 0.1P and 0.3P samples under the two different FSW conditions.
3.2. Microstructures of Base Materials and Stir Zones

Figure 4 shows the SEM microstructures of the BMs and SZs. The BM for each sample consisted of ferrite and pearlite structures. The ferrite grain diameters of the 0.1P and 0.3P samples are 49 μm and 23 μm, respectively. The SZ obtained at a welding temperature above $A_3$ (SZ above $A_3$) consisted of low-temperature transformation products, such as bainite and martensite, as well as ferrite and pearlite. These products were formed during the transformation from austenite. The amount of low-temperature transformation products in the SZ decreased with increasing the P content. Furthermore, after FSW, the grain diameters of the 0.1P and 0.3P samples decreased to 12 and 7.3 μm after FSW, respectively. The SZ obtained at a welding temperature below $A_1$ (SZ below $A_1$) consisted of ferrite and fine spherical cementite. The ferrite grain diameters of the 0.1P and 0.3P samples were 1.4 and 2.5 μm, respectively. Contrary to the expectation, the 0.3P sample had a larger ferrite diameter. The SZs of both materials were extremely refined via FSW below the $A_1$ temperature. Moreover, the sizes of fine cementite for the SZs obtained below $A_1$ were approximately 0.5 μm or less.

Figure 5 shows ferrite microstructures consisting of high-angle grain boundaries with a misorientation of >15°; these microstructures were obtained through inverse pole figure (IPF) maps acquired using the EBSD. The grains of the SZs of all samples were finer than those of the BMs due to dynamic recrystallization during the FSW. In particular, the grains of the SZ obtained below $A_1$ were considerably finer than those of the SZ obtained above $A_1$. The SZ obtained below $A_1$ for the 0.1P sample showed relatively homogeneous fine grains with an average grain diameter of approximately 1.4 μm; in contrast, one of the 0.3P samples showed a bimodal structure with an average grain diameter of approximately 2.6 μm.

Fig. 4. SEM microstructures of base material and stir zones under above $A_3$ and below $A_1$ conditions for 0.1P and 0.3P samples.

Fig. 5. EBSD map showing grain boundaries of base materials and stir zones under above $A_3$ and below $A_1$ conditions for 0.1P and 0.3P samples.
Figure 6 shows the grain diameter distributions in the SZs obtained below $A_1$ for the 0.1P and 0.3P samples. The dashed lines represent the average grain diameter. For the 0.1P sample, all grains had a diameter of $<4 \mu m$. Contrarily, for the 0.3P sample, several grains had a diameter of $>4 \mu m$, and a second peak existed at a grain diameter of approximately $4 \mu m$ (arrow in Fig. 6(b)). This result indicates a bimodal-like grain size distribution.

3.3. Mechanical Properties of Base Materials and Stir Zones

Figure 7 shows the engineering stress–strain curves of all samples. The curve for the BM of SMA490AW is also shown for comparison. In each sample, the tensile strength increased in the order of BM $<\frac{\text{SZ}}{A_3} < \frac{\text{SZ}}{A_1}$. For both BMs and SZs, the 0.3P sample had a tensile strength higher than that of the 0.1P sample. In the stress–strain curves of the SZs obtained below $A_1$, large yield point elongations were observed regardless of the P content. Furthermore, the 0.1P sample with relatively uniform fine ferrite and cementite exhibited an excellent balance between tensile strength and total elongation.

Figure 8 shows the temperature dependence of the absorbed impact energy for all samples. For the 0.1P sample (Fig. 8(a)), the DBTTs decreased in the order of $\text{SZ below } A_1 < \frac{\text{SZ}}{A_3} < \text{BM}$. In particular, the SZ obtained below $A_1$ exhibited ductile fracture even at a boiling point of liquid nitrogen, as well as high absorbed energy. Similar to those for the 0.1P sample, the DBTTs for the 0.3P sample (Fig. 8(b)) also decreased in the order of $\text{SZ below } A_1 < \frac{\text{SZ}}{A_3} < \text{BM}$. However, compared to those for the 0.1P sample, the DBTTs for the 0.3P sample increased due to the increase in P content in both the BM and the SZs. The SZ obtained below $A_1$ for the 0.3P sample exhibited brittle fracture at about $-180^\circ C$, which deviated from the behavior of the 0.1P sample.

Figure 9 shows the fracture surfaces of an entirely brittle-fractured specimen (arrows in Fig. 8) for all samples. The BMs in each sample exhibited a cleavage fracture surface, where a river pattern was observed. Furthermore, the SZs obtained above $A_3$ exhibited a cleavage fracture surface regardless of the P content. However, the unit size of the fracture surface of the SZ obtained above $A_3$ was smaller than that of the BM. In contrast, the appearance of the fracture surface of the SZ obtained below $A_1$ for the 0.3P sample was significantly different from that of the SZ obtained above $A_3$. Intergranular fracture surfaces with a...
size of approximately 1 μm (arrow in Fig. 9(e)), corresponding to the ferrite grain size, were observed in some areas. No brittle fracture surface was observed on the SZ obtained below $A_1$ for the 0.1P sample, because ductile fracture was exhibited at all temperatures.

4. Discussion

4.1. P Content and Weldability

The cross-sectional macro-photographs (Fig. 3) show that the joints were defect-free for all samples under all conditions. FSW can join weathering steels containing 0.1 mass% C even with at least 0.3 mass% P.

P impairs fusion-weldability primarily because it promotes solidification cracking. The solid–liquid coexistence region during solidification has extremely low ductility; such a coexistence region is called the brittleness temperature range (BTR). Therefore, when the strain due to solidification shrinkage exceeds the critical strain, solidification cracking occurs.\(^{40}\) The P accumulates in the liquid phase during solidification and expands the BTR to cover lower temperatures. This expansion increases the risk of solidification cracking. The phase diagram in Fig. 2(b) shows that as the temperature drops from the stable temperature range of the liquid phase, the δ phase is formed first, followed by the γ phase. Subsequently, the three-phase ($\delta+\gamma+L$) coexistence, and finally, the two-phase ($\delta+\gamma$) coexistence is reached. However, at the final solidification stage, the concentration of P increases remarkably. As shown in Fig. 2(b), the increase in P concentration expands the temperature range of the $\delta+\gamma+L$ coexistence state to cover lower temperatures, resulting in high-temperature cracking. In contrast, the solid-state joining method FSW can join steels without the solidification process, resulting in defect-free joints.

P influences the microstructure and reduces the joint toughness in the fusion welding process,\(^{41-43}\) for example, by delaying the precipitation of cementite from the untransformed austenite and by promoting the formation of the martensite–austenite (M–A) component, which reduces
toughness.\(^{41}\) In contrast, during FSW at the maximum temperature below \(A_\text{f}\), the transformation from ferrite to austenite does not occur, which prevents M–A formation. Moreover, as discussed in Section 4.2.3, significant grain refinement could improve the toughness of the SZ owing to FSW. Therefore, FSW of high-P steel prevents defects in the joints, as well as improves the mechanical properties of the joints.

### 4.2. Mechanism of Microstructure Formation

Ferrite grains were refined due to the increase in the P content in the BM and SZ obtained above \(A_\text{f}\) (Figs. 4, 5). This result could be attributed to the solute drag effect of P. P segregates at both the austenite and the ferrite grain boundaries.\(^{9-15,44}\) The solid solubility of P in austenite is lower than that in ferrite; hence, the grain boundary segregation is assumed to be substantial in austenite. Therefore, the austenite grains formed during the fabrication process of the BM and during the FSW above \(A_\text{f}\) are presumably refined as the P content increases, owing to the solute drag effect. Accordingly, the grain structures obtained through subsequent cooling also tend to be refined as the P content increases.

In the SZ obtained above \(A_\text{f}\), the amount of low-temperature transformation products, such as bainite and martensite, decreased due to the increase in the P content (Fig. 4). In general, P is a ferrite-forming element, so which raises the \(A_\text{f}\) temperature and promotes ferrite formation.\(^{45,46}\) Furthermore, as already mentioned, P leads to austenite grain refinement during FSW, promoting ferrite transformation. However, P also enhances hardenability.\(^{47}\) When the P content was increased in this study, the increase in the \(A_\text{f}\) temperature and the refinement of the austenite grains were presumably more significant than the increase in the hardenability. Consequently, the amount of low-temperature transformation products in the SZ was reduced with an increase in P content.

By contrast, in the SZ obtained below \(A_\text{f}\), microstructures composed of fine ferrite and fine cementite were obtained regardless of the P content (Figs. 4 and 5). During FSW below \(A_\text{f}\), the ferrite was refined through dynamic recrystallization, and the pearlite present in the BM was crushed and finely dispersed. However, the ferrite in the 0.3P sample was larger than that in the 0.1P sample and exhibited a bimodal structure. As already discussed, the amount of pearlite than the 0.1P sample did (Figs. 4(a) and 4(d)). When the BM composed of ferrite and pearlite was friction-stir-welded at a temperature below \(A_\text{f}\), the nucleation frequency of the dynamically recrystallized grains in the region containing pearlite was presumably higher than that in the ferrite region; in addition, more significant grain refinement during FSW occurred in the region of prior pearlite in the BM. Therefore, the ferrite size in the 0.3P sample after FSW could be larger than that in the 0.1P sample, because the 0.3P sample had a smaller amount of pearlite in the BM. Moreover, the amount of pearlite in the BM of the 0.3P sample was small, and the distribution of the pearlite was considerably heterogeneous; hence, the 0.3P sample was assumed to have a prominent bimodal structure.

### 4.3. Mechanical Properties

#### 4.3.1. Strength and Elongation

In the nominal stress–strain curves (Fig. 7) of the SZ obtained above \(A_\text{f}\), the 0.1P and 0.3P samples behaved differently; the 0.1P sample showed continuous yielding, whereas the 0.3P sample showed discontinuous yielding. This result could be due to the difference in the amount of low-temperature transformation products, such as bainite and martensite, between the 0.1P and 0.3P samples. Steels consisting of ferrite and low-temperature transformation products exhibit continuous yielding owing to the presence of mobile dislocations at the interface between soft and hard structures.\(^{48}\) Therefore, the continuous yielding of the 0.1P sample could be due to the larger amount of low-temperature transformation products. In contrast, the discontinuous yielding of the 0.3P sample could be because of the considerably small amount of martensite. Moreover, the large amounts of bainite and martensite resulted in poor elongation of the SZ obtained above \(A_\text{f}\) for the 0.1P sample compared with that for the 0.3P sample.

The strength of the SZ obtained below \(A_\text{f}\) was higher than that of the BM and the SZ obtained above \(A_\text{f}\) in both the 0.1P and 0.3P samples. Furthermore, the 0.1P sample had an excellent balance between strength and ductility. Compared with those of the BM, the yield point strength of the SZ obtained below \(A_\text{f}\) for the 0.1P and 0.3P samples were improved by 394 MPa and 353 MPa respectively. The Hall–Petch relationship holds between the yield strength and the grain diameter,\(^{52-55}\) thus, the improvement in the yield strength in the SZ obtained below \(A_\text{f}\) was mainly due to ferrite grain refinement.

The SZs obtained below \(A_\text{f}\) for both the 0.1P and 0.3P samples showed remarkable yield point elongation. Yield point elongation develops and increases with grain refinement.\(^{52-55}\) Therefore, the fine grains in the SZ obtained below \(A_\text{f}\) must lead to yield point elongation. Furthermore, the yield point elongation of the 0.1P sample was larger than that of the 0.3P sample; this phenomenon can be attributed to the finer grains of the 0.1P sample. In addition, the local elongation of the 0.1P sample was superior to that of the 0.3P sample. Grain refinement is reported to improve the reduction of area in a tensile test.\(^{56}\) Therefore, the SZ obtained below \(A_\text{f}\) for the 0.1P sample with grains finer than those of the 0.3P sample was assumed to exhibit improved local elongation. In contrast, the relatively small local elongation of the 0.3P sample might be due to the larger grains as well as the bimodal structure. In the bimodal structure of the 0.3P sample, the strain should be concentrated in the weak part of the structure with larger grains, resulting in reduced local elongation.

#### 4.3.2. Toughness

Compared with those of the BM, the DBTTs of the SZs decreased both for the 0.1P and 0.3P samples. In particular, the SZ obtained below \(A_\text{f}\) for the 0.1P sample did not cause brittle fracture even at a boiling point of liquid nitrogen. Furthermore, the SZ obtained below \(A_\text{f}\) for the 0.3P sample exhibited excellent toughness with a DBTT as low as −175°C. The excellent toughness in the SZ obtained
below \( A_1 \) is assumed to be primarily controlled by grain refinement. Figure 10 shows the relationship between \( (d)^{-1/2} \) and DBTT for all samples, where \( d \) is the grain diameter. For comparison, previously reported data for C–Mn,\(^{57} \) C–Mn–Nb,\(^{58} \) and IF steels\(^{54} \) are also included. The slopes of the lines for the 0.1P and 0.3P samples were assumed to be equal to that of the C–Mn steel (0.15C–1.5Mn [mass%]), which has approximately the same amount of C. For the 0.1P sample, the DBTT of the SZ obtained above \( A_1 \) was higher than the line of the BM. This result could be due to the presence of coarse martensite (>1 \( \mu \)m) in the SZ obtained above \( A_1 \) (Fig. 4). Meanwhile, for the 0.3P sample, the DBTT of the SZ obtained below \( A_1 \) was slightly higher than that of the BM. This phenomenon could be due to grain boundary embrittlement because the sample showed grain boundary fracture (Fig. 9). DBTT was slightly affected by the microstructure and fracture morphology. However, for the 0.1P and 0.3P samples, FSW both above \( A_1 \) and below \( A_1 \) significantly improved the toughness owing to grain refinement. In particular, the SZ obtained below \( A_1 \) for the 0.3P sample showed a DBTT lower than that of a commercial SMA490AW (Fig. 8) specimen in which the amount of P was reduced to 0.01 mass%. This result indicates that grain refinement can overcome embrittlement due to high P content.

The difference between the DBTTs of the 0.1P and 0.3P samples with identical grain diameters \( d \) (Fig. 10) could be attributed to the effect of P on DBTT. Increasing the amount of P by 0.2 mass% increased the DBTT to 115°C, attributed to the effect of P on DBTT. Increasing the amount of P by 0.2 mass% increased the DBTT to 115°C. This result indicates that grain refinement can over-

5. Conclusions

The following findings were obtained by investigating the weldabilities, microstructure evolution, and mechanical properties of the FSW joints of high-P (0.1–0.3 mass%) weathering steel, which cannot be joined through fusion welding.

(1) FSW provided defect-free joints at both welding temperatures below \( A_1 \) and above \( A_1 \). This occurred because FSW does not involve solidification, which causes defects in fusion welding. A weathering steel containing 0.1 mass% C can be joined via FSW even with at least 0.3 mass% P.

(2) The grains of the SZs were remarkably finer than those of the BMs, particularly during FSW at temperatures below \( A_1 \). The DBTTs decreased with grain refinement. In the 0.3P samples, the DBTT of the SZ obtained below \( A_1 \) (~2 \( \mu \)m) was reduced to approximately 150°C from that of the BM (~23 \( \mu \)m). This result indicates that grain refinement can overcome embrittlement caused by a high P content.

(3) As the P content increased, the grains in the BMs and SZs obtained above \( A_1 \) were refined. This phenomenon could be due to the solute drag effect of P. Meanwhile, in the SZs obtained below \( A_1 \), the 0.3P sample had large grains and showed a bimodal structure. This finding can be
attributed to the small amount of pearlite in the BM of the sample, resulting in a low nucleation rate of the dynamically recrystallized ferrite during FSW and increased heterogeneity of the nucleation site.

(4) In the BMs and SZs obtained above $A_1$, the fracture surface of an entirely brittle-fractured specimen showed a cleavage fracture. Conversely, in the SZ obtained below the surface of an entirely brittle-fractured specimen showed a low nucleation rate of the dynamically recrystallized ferrite during FSW and increased heterogeneous grain boundary fracture. 

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