Galvannealed (GA) boron steel sheets were subjected to direct hot stamping tests (V-bending) to investigate crack formation behavior. Specimens were heated at 1 173 K (900°C) in a combustion gas furnace and hot stamped on a cooled V-shaped die. The heating time was varied to alter the composition, phase, and microstructure of the coating layer. Generally, the reaction zone between the zinc coating layer and metal substrate could not be observed by scanning electron microscopy. For certain specimens, crack penetration occurred along primary austenitic grain boundaries of the substrate, with zinc enrichment observed at the crack sidewalls suggesting the occurrence of liquid metal embrittlement (LME) of the base metal. LME-induced cracking resulted when the coating layer corresponded to a biphasic Fe–Zn ferrite/liquid zinc structure of the binary Fe–Zn phase diagram during heating in the furnace. For cracked specimens, the maximum crack depth increased with increasing total X-ray diffraction intensity of θ and ε phases in the coating layer of hot-stamped samples, being positively correlated with the amount of “liquid zinc” in the coating layer.

KEY WORDS: hot stamping; galvannealed boron steel; crack; liquid metal embrittlement; liquid zinc; Fe–Zn ferrite; primary austenitic grain boundary.

1. Introduction

The use of automotive parts made from thin high-strength steel sheets has become prevalent due to achieving high collision safety and significant weight reduction of automobiles. Since boron-added steel exhibits good hardenability, it has been widely used for hot stamping. Direct hot stamping of boron steel sheets is characterized by simultaneous quenching and stamping, producing high-strength 1 500 MPa grade steel parts with good shape accuracy. Automotive steel parts require good surface quality for proper spot welding and paint adhesion. For uncoated boron steel sheets, heating in air or combustion gas before quenching produces iron oxide scale on their surface. Hence, descaling by shot blasting or other means is generally needed after hot stamping to achieve good surface quality. To shorten the descaling procedure, pre-coating solutions such as aluminized and galvannealed (GA) boron steels were developed, becoming widely used for automotive parts. This technology has succeeded in suppressing scale growth and securing the necessary surface quality of steel parts.

Liquid metal embrittlement (LME), a catastrophic brittle failure of ductile metals covered with a thin liquid metal film and subsequently subjected to tensile stress, is documented to occur in ferritic steels and austenitic stainless steels covered by liquid zinc. Embrittlement of ferritic steels by liquid zinc has been reported to occur between 673 K (400°C) and 893 K (620°C), with long-term exposure and intercrystalline attack needed to reduce the elongation to failure. A zinc-iron intermetallic layer functions as an inhibitor to suppress embrittlement. The occurrence of LME was recognized for high-strength structural steel products (e.g., steel pylons used for electric power lines) that were subjected to galvanization in a liquid zinc bath, and retained a certain tensile stress during galvanization.

Hot stamping involves heating the steel sheets to ~1 173 K (900°C), transferring them from the heating furnace to a die, and simultaneous hot forming and die quenching to obtain components with the desired shape and strength. Since the heating temperature generally lies well above the melting point of pure zinc (693 K (420°C)), galvanized
steel is expected to be exposed to liquid zinc and thus be embrittled. Drillet et al. conducted an omega-shape laboratory stamping test of galvannealed steel and showed that macro-cracks found in hot-stamped galvanized steel could be attributed to tensile-stress corrosion by liquid zinc that penetrated the grain boundaries of the substrate.

The authors suggested that a coating layer composition corresponding to the liquid phase of the binary Fe–Zn hot stamping phase diagram may increase the risk of cracking due to LME. Lee et al. examined LME in galvanized boron steel using a Gleeble thermomechanical process simulator, revealing that Fe–Zn intermetallic compounds transformed into a Zn-rich liquid phase above the peritectic temperature of the following transformation:

$$\Gamma_1 \rightarrow \text{Fe–Zn ferrite + Liquid Zn}$$

The above liquid phase penetrated the primary austenitic grain boundaries of the base metal and resulted in cracking failure under applied tensile stress. Takahashi et al. conducted a hot V-bend test of galvannealed boron steel to simulate hot stamping, pointing out the influence of heating time on the propensity to crack.11)

For hot stamping of zinc-coated boron steel, it is generally accepted that LME of the base metal occurs for the austenitic but not for ferritic structure.10) Lee et al. suggested that a zinc-rich liquid phase penetrates the subsurface austenitic grain boundaries of the base metal, weakening the cohesion of high-temperature grain boundaries and causing base steel cracking.10) Hot stamping requires heating steel sheets to temperatures at which their structure transforms from ferrite to austenite prior to quenching. Upon heating, the initial intermetallic phases and metallic zinc of the zinc coating layer transform to liquid zinc and Fe–Zn ferrite, with liquid zinc penetrating the primary base metal austenitic grain boundaries when tensile stress is applied. Hence, cracks generated along the primary austenitic grain boundaries of the base metal were examined in detail.

A major characteristic of LME in hot-stamped galvannealed and galvanized steels seems to be due to the fact that the composition of the zinc coating layer varies during heating at 1 173 K (900°C). In contrast to the galvanization of structural steel products, where abundant liquid zinc is present in the bath, hot stamping features the formation and annihilation of liquid zinc in the coating layer, which seems to be directly related to the occurrence/absence of LME of galvanized and galvannealed steel. Therefore, tracing the locus of the composition and temperature of the coating layer during heating and relating these to the occurrence/absence of LME is probably required for the correct understanding of this phenomenon. The present study aimed to examine the heating-time-dependent composition and phase changes of the coating layer of galvannealed boron steel during heating and subsequent laboratory hot stamping tests. The specimen temperature was monitored during heating and at the start of hot stamping, and the microstructures of the zinc coating layer and base metal were carefully examined. The depth and width of LME-induced cracks generated during hot stamping were measured, and the coating layer zinc concentration and phases of hot-stamped specimens were analyzed by energy-dispersive X-ray (EDX) spectroscopy and X-ray diffraction (XRD). The maximum specimen temperature at the onset of stamping, and the coating layer zinc concentration and phases were focused on and discussed from the perspective of LME.

2. Experimental Procedures

A 2.6-mm-thick GA steel sheet containing 0.21 wt.% C, 0.24 wt.% Si, 1.3 wt.% Mn, and 0.0018 wt.% B was used for the hot stamping test. The zinc coating weight per side equaled 61 g/m², and the Fe content equaled 12 wt.%. The coating layer thickness was around 10 microns; it consisted predominantly of the δ phase with some Γ₁ phase and slight η phase. Steel sheet used in this study had a ferritic-pearlitic microstructure. Specimens (60 × 40 mm) were cut from the sheet and heated at 1 173 K (900°C) in a combustion gas furnace, with a representative temperature history shown in Fig. 1. The specimens were taken out of the furnace after heating for 90–300 s, placed on a V-shaped die, and hot stamped using a punch operated by a hydraulic press machine. The forming speed equaled 100 mm/s, the load at the bottom equaled 40 kN, and the hold time at the bottom dead center corresponded to 30 s. The angle of the V-shaped die equaled 90°, the punch tip radius equaled 0.5 mm, and the die was continuously cooled by water, as schematically shown in Fig. 2. The true strain (0.45) of the hot-stamped specimen was estimated from the deformation of the indentation engraved by a Vickers hardness tester at the tip of the
V-bent specimen. The average time passed between taking the specimens out of the furnace and the start of stamping equaled 7–9 s. From the start of heating to stamping, the sample temperature was continuously monitored using a thermocouple. After hot stamping, the specimens were cut in half and mounted in resin to obtain a cross-section. Cross-sections at the external surface tip of V-bent specimens were examined by scanning electron microscopy (SEM) using secondary and backscattered electrons (BSE). SEM observations were conducted for a quarter circumference of the specimen front and an observation length of ~4 mm along the specimen surface. Some hot-stamped specimens were etched by picric acid to examine the microstructure and primary austenitic grain boundaries of the base metal. Zinc elemental mapping was performed for the cracked specimen cross-section by electron probe microanalysis (EPMA). In order to obtain the mean zinc concentration in the coating layer, five rectangular areas of approximately 8 × 10 μm in the middle of the zinc coating layer were characterized by EDX at an accelerating voltage of 15 kV, and the obtained data were averaged. For some coating layers, EDX point analyses were performed to obtain the zinc concentration in a selected microstructure. Hot stamped specimens taken from an unbent portion directly contacting the water-cooled die during stamping were also characterized by XRD (Co Kα radiation), since they were expected to have experienced the same temperatures as the hot V-bent portion. For XRD, diffraction counts per second (cps) were accumulated for a certain diffraction angle (2θ) range: 93–95° for the {721} plane of the Γ1 phase (JCPDS: 33-697), 48.5–49.2° for the {1412} plane of the δ phase,12 and 63.5–64.5° for the {1012} plane of the η (hcp-Zn) phase (JCPDS: 4-831), with background counts subtracted.

### 3. Results

The maximum specimen temperature upon heating in the furnace and the specimen temperature at the start of stamping were measured for respective experimental conditions and are listed in Table 1. The specimen temperature exhibited a maximum after heating for 180 s and remained at ~1 163 K (890°C) thereafter, whereas the specimen temperature at the start of stamping increased with increasing heating time up to 180 s, remaining in the range of 1 054 K (781°C) to 1 073 K (800°C).

Cross-sections of the front bend portion of specimens heated for 90–240 s are shown in Fig. 3. Irregular surfaces were observed for samples heated for 90 and 240 s due to cracking of the zinc coating layer. No cracking was observed for the metal substrate. Specimens heated for 120 and 180 s exhibited base metal cracking, which was much deeper than the thickness of the coating layer. Cross-sections of some typical large and deep cracks of specimens heated for 120, 150, and 180 s are presented in Fig. 4. These specimens were etched by picric acid to highlight the primary austenitic grain boundaries of the metal substrate. Cracks were also observed for the zinc coating layer, attributed to rapid forming and probably to the dissolution of the quenched liquid zinc by picric acid. Magnified cross-sections of the edges of the large cracks shown in Fig. 4 are presented in Fig. 5, with these areas considered to be the initiation sites of the above large cracks. The zinc coating layer of samples heated for 120, 150, and 180 s consisted of cracks and large grains (~10 μm and larger), as shown in Fig. 5. The cracks are considered to be introduced by hot stamping, presumably at sites where quenched liquid zinc was present.13 The quenched liquid zinc was expected to be dissolved in picric acid, with the remaining large grains composed of Fe–Zn ferrite.13 The subsurface metal exhibited martensite grains surrounded by primary austenitic grain boundaries (Fig. 5), with some crack fronts heading to these boundaries (see white arrows in Fig. 5(a)). Cracking possibly initiated at the bottom of liquid zinc phases (which were etched by

### Table 1. Maximum specimen temperatures upon heating and specimen temperatures at the start of hot stamping for respective tests.

<table>
<thead>
<tr>
<th>Heating time in the combustion gas furnace (s)</th>
<th>Maximum specimen temperature (°C)</th>
<th>Specimen temperature at start of stamping (°C)</th>
</tr>
</thead>
<tbody>
<tr>
<td>90</td>
<td>763</td>
<td>726</td>
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<tr>
<td>120</td>
<td>834</td>
<td>750</td>
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<tr>
<td>150</td>
<td>878</td>
<td>761</td>
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<td>210</td>
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<td>225</td>
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<td>799</td>
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<td>896</td>
<td>785</td>
</tr>
<tr>
<td>270</td>
<td>889</td>
<td>781</td>
</tr>
<tr>
<td>300</td>
<td>889</td>
<td>781</td>
</tr>
</tbody>
</table>

![Fig. 3. Cross-sectional BSE (backscattered electron) images of the surface portion at the tip of the bent specimens heated for (a) 90 s, (b) 120 s, (c) 180 s, and (d) 240 s.](image-url)
picric acid and appear “vacant” in Fig. 5(b)) and penetrated primary austenitic grain boundaries of the base metal (see black arrow in Fig. 5(b)). Therefore, cracking was attributed to liquid zinc, which was supplied from the coating layer and caused LME.\(^9,10\) In this study, cracks which are found in metal substrate and deeper than five microns are categorized as LME cracks.

EPMA mapping of zinc was conducted for a cross-section of a cracked specimen heated for 180 s, which was also examined by SEM (Fig. 6). The base metal crack sidewalls were enriched by zinc, suggesting that the observed large crack was induced by zinc intrusion from the zinc coating layer into the primary austenitic grain boundaries of the metal substrate. Hence, it is legitimate to state that liquid zinc formed in the coating layer penetrated the primary austenitic grain boundaries of the base metal, causing base steel cracking. Therefore, this cracking can be categorized as LME caused by liquid zinc.

It should be pointed out that a “reaction zone” expected to be generated between the zinc coating layer and the substrate metal upon heating\(^14\) was not visible, implying only minor diffusion of zinc into the \(\gamma\)-Fe substrate. Therefore, it is reasonable to accept that cracking of the metal substrate is initiated not at the “reaction zone” but at sites where the metal substrate is enriched with zinc.
metal substrate directly contacts liquid phases of the zinc coating layer. Therefore, we have focused on the chemistry and phases of the zinc coating layer.

LME-induced cracking was evaluated by measuring the depths and widths of the respective cracks penetrating the base metal, as schematically defined in Fig. 7. The crack depths and widths were plotted as a function of furnace heating time (Figs. 8 and 9). No LME-induced cracking was observed for the specimen heated for 90 s, while significant crack formation was detected for specimens heated for 120–195 s. Crack depth decreased with increasing heating time above 180 s. Cracks in specimens heated for more than 240 s were not significant, indicating the absence of LME for these samples. Crack width followed a similar heating time dependence. The deepest cracks were generally found near the center of the bend front.

The average zinc concentration in the coating layer (obtained by EDX) and the X-ray diffraction intensity of the respective phases are summarized in Table 2. The zinc concentration of the coating layer for a hot stamped specimen heated for 90 s equaled 75 wt.%, decreasing with increasing heating time to approach 29 wt.% after heating for 300 s. This decrease is considered to result from the inward diffusion of iron from the base metal to the zinc coating layer during heating in the furnace. It should be pointed out that for the current heating conditions, zincification of the underlying base metal is estimated to be virtually negligible. Data pertaining to the diffusion constant of zinc in γ-Fe are scarce, one example being the diffusion constant of zinc in Fe-20 wt.% Zn ferrite at 830–1 064 K (557–791°C)\(^{15}\). Extrapolating these data to 1 173 K (900°C) afforded a diffusion depth of ~0.5 μm for zinc in Fe–Zn ferrite heated to 1 173 K (900°C) for 210 s. Since the diffusion of zinc in γ-Fe is expected to be much slower than in α-Fe, zincification of γ-Fe is considered to be virtually negligible, and one does not necessarily need to consider the formation of Fe–Zn ferrite between the coating layer and base metal in the present experiments. Indeed, according to the microscopic observations shown in Fig. 5, the formation of an Fe–Zn ferrite layer between the zinc coating layer and the underlying base metal was not clear. The subsurface region of the base metal, contacting the zinc coating layer, exhibited a martensitic structure, with the transformation of martensite to Fe–Zn ferrite at these sites not clearly indicated. Therefore, one might accept that the γ-Fe metal substrate maintained “direct” contact with the zinc coating layer, and once a part of this layer melted, the liquid zinc intruded the primary austenitic grain boundaries of the γ-Fe substrate, causing cracking of the metal substrate upon hot stamping.

The as-received zinc coating layer comprised predominantly the δ phase with some Γ\(_1\) and η phases (Table 2). Most of the above δ and η phases transformed into the Γ\(_1\) phase during heating for 90 s, but a part of the Γ\(_1\) phase decomposed to re-generate δ and η phases when heated for 120 s. The content of Γ\(_1\) and δ phases decreased for

<table>
<thead>
<tr>
<th>Heating time in the combustion gas furnace (s)</th>
<th>Zinc concentration (wt.%)</th>
<th>X-ray diffraction intensity of phases (cps)</th>
<th>Γ(_1)</th>
<th>δ</th>
<th>η</th>
</tr>
</thead>
<tbody>
<tr>
<td>As received</td>
<td>88</td>
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<td>631</td>
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<td>125</td>
<td>62</td>
<td>12</td>
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<td>150</td>
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<td>57</td>
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<td>12</td>
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<td>35</td>
<td>19</td>
<td>29</td>
<td>8</td>
<td>&lt;5</td>
</tr>
<tr>
<td>195</td>
<td>36</td>
<td>9</td>
<td>20</td>
<td>&lt;5</td>
<td>&lt;5</td>
</tr>
<tr>
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<td>225</td>
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<tr>
<td>240</td>
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<tr>
<td>300</td>
<td>29</td>
<td>&lt;5</td>
<td>&lt;5</td>
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</tr>
</tbody>
</table>
specimens subjected to prolonged heating, and they could no longer be detected by XRD for samples heated for longer than 270 s. The pure zinc phase (η) was present only in the as-received specimens and those heated for 120–180 s.

4. Discussion

First, we would like to discuss the occurrence of LME during the hot stamping of zinc-coated boron steel from the perspective of liquid zinc formation in the coating layer, following the study of Drillet et al.9) Since the formation of liquid zinc is confined to the zinc coating layer, the average zinc concentration therein, along with the binary Fe–Zn phase diagram, can provide information on the presence/absence of liquid zinc in this layer. Hence, the average zinc concentration in the coating layer upon hot stamping and the specimen temperature of the stamping onset were plotted on the binary Fe–Zn phase diagram, as shown in Fig. 10. Heating time was designated for respective specimens. LME took place even for specimens where the temperature at the stamping onset was below the solidification temperature of liquid zinc (1 055 K (782°C)), suggesting that the liquid zinc formed in the furnace may have remained in the coating layer until hot stamping, causing LME. Therefore, the formation of liquid zinc in the coating layer upon heating was examined in detail.

The average zinc concentration in the coating layer and the maximum specimen temperature attained in the furnace were plotted on the binary Fe–Zn phase diagram (Fig. 11). For a 90-s heating time, the coating was suggested to primarily comprise the Γ1 phase, in agreement with XRD results. Since the composition of the as-received galvannealed coating layer corresponded to 88 wt.% Zn-Fe, dilution of zinc by iron diffusing from the base metal during heating would produce the Γ1 phase in the coating layer. The fraction of the δ phase, accounting for the majority of the as-received coating layer, significantly decreased for this specimen (Table 2). The maximum specimen temperature in the furnace, 1 036 K (763°C), was below the melting point of the Γ1 phase, 1 055 K (782°C). Since no liquid phase was present at hot stamping, the above specimen did not exhibit cracking.

For specimens heated for 120 and 150 s, the formation of liquid zinc and Fe–Zn ferrite is suggested to have taken place in the coating layer during heating in the furnace at 1 107 K (834°C) and 1 151 K (878°C), respectively, and cracking is considered to have occurred due to LME upon hot stamping. Although the specimen temperatures at the start of stamping equaled 1 023 K (750°C) and 1 034 K (761°C), respectively, being below the solidification temperature of liquid zinc (1 055 K (782°C)), some liquid zinc is suggested to have remained in the coating layer upon hot stamping. XRD characterization of the coating layer identified δ and η phases, supporting the presence of liquid zinc at hot stamping. Since the time required for specimen transfer from the heating furnace to the hot stamping die was in the range of 7–9 s, the following solidification reaction seems to be incomplete upon hot stamping:

\[
\text{Fe} - \text{Zn ferrite} + \text{Liquid Zn} \rightarrow \text{Fe} - \text{Zn ferrite} + \Gamma_1
\]

Specimens heated for 180 and 195 s are suggested to still exhibit the average biphasic liquid zinc/Fe–Zn ferrite coating composition in the furnace, as shown in Fig. 11. For these specimens, hot stamping was performed at 1 058 K (785°C) and 1 073 K (800°C), respectively, i.e., above the melting point of the Γ1 phase (1 055 K (782°C)). Although the corresponding amount of liquid zinc at hot stamping is expected to be less than those of specimens heated for 120 and 150 s, LME-induced cracking still occurred. XRD indicated the presence of both δ and η phases in these specimens, suggesting that liquid zinc remained in their coatings during hot stamping. Hence, LME is considered to have also taken place for these specimens.

Figure 11 indicates that the coating compositions of specimens heated for 210 and 225 s were located near the boundary between the Fe–Zn ferrite phase and the biphasic region containing liquid zinc, resulting in slight cracking of these specimens. Since the zinc concentration was analyzed for hot-stamped specimens and not for those just taken out...
from the furnace, the coatings heated in the furnace for 210 and 225 s might have been slightly enriched with zinc compared to the analyzed specimen. Hence, a small amount of liquid zinc might have existed in the coating layer upon heating, causing minor LME in the base metal.

No LME occurred for specimens heated for longer than 240 s. The corresponding coating layers are suggested to comprise a single Fe–Zn ferrite phase (see Fig. 11), with liquid zinc considered not to be present in the coating layer upon heating in the furnace. For these specimens, diffusion of iron from the base metal during heating resulted in the formation of a solid phase in the coating layer. The XRD results, indicating the presence of a small amount of the δ phase and no Γ₁ or η phases, support this conclusion. However, it should be emphasized that liquid zinc must have formed for these specimens at least during the initial heating stage of 120–225 s, since LME-induced cracking took place in specimens subjected to a nearly identical temperature treatment. Specimens heated for longer than 240 s exhibited the formation and annihilation of liquid zinc during heating in the furnace.

To confirm the liquid phase formation in the coating layer of cracked specimens, cross-sections of non-deformed areas of hot-stamped specimens heated for 120 and 150 s were investigated by SEM (Fig. 12), revealing significant LME at front bends (see Fig. 3). Figure 12 indicates that despite some cracks noticed in the coating layer, the underlying base metal was crack-free, suggesting that the applied tensile stress was lower than that needed to initiate LME in the base metal. The coating layer comprised two phases, colored white and light gray, particularly pronounced for the 120-s specimen. EDX analysis revealed that the white phase was enriched in zinc, exhibiting a Zn content of 76–85 wt.% for the 120-s specimen and that of 85–90 wt.% for the 150-s specimen (Fig. 12). The white Zn-rich phase must originate from the liquid phase formed in the furnace, since its zinc concentration is much higher than average coating layer levels (44 wt.% for the 120 s specimen and 39 wt.% for the 150 s specimen). It should be pointed out that the zinc content of the white phase was even higher than the peritectic zinc composition of the Γ₁ phase (75 wt.%), suggesting the formation of a Zn-rich liquid phase in the coating layer of these specimens during heating in the furnace.

According to the Fe–Zn phase diagram, the most of the δ phase identified in the cracked specimen coating layer by XRD (Table 2) was formed via solidification of the liquid phase at 945 K (672°C). Similarly, the η phase can be considered to originate from the liquid phase. Therefore, the X-ray diffraction intensities of δ and η phases (Table 2) were summed and used as a quantitative index to represent the amount of the liquid phase existing in the coating layer upon heating in the furnace. The maximum crack depth was plotted as a function of the total X-ray diffraction intensity (Fig. 13), increasing with increasing total XRD intensity and

![Fig. 13. Maximum crack depth as a function of the total X-ray diffraction intensity of δ and η phases (cps).](image)

**Fig. 12.** BSE images and spot EDX zinc concentrations for coating layer of non-deformed as-polished specimens heated for 120 and 150 s.
subsequently exhibiting saturation. Assuming that the above XRD intensity reflects the amount of liquid phase present in the coating layer formed in the furnace, the maximum depth of LME-induced cracks increased with increasing liquid phase amount, equaling 127 μm for the current test geometry and conditions.

5. Conclusions

Direct hot stamping tests were conducted for a galvannealed boron steel sheet to investigate crack formation. The depths and widths of the obtained cracks were measured, and the zinc concentration, phases, and microstructure of the coating layer and the subsurface metal were analyzed in detail.

- Cracks penetrated the base metal in specimens heated for 120–225 s. For specimens heated for 90 s or longer than 240 s, the cracks remained inside the coating layer and did not affect the base metal.
- For cracked specimens etched by picric acid, the zinc coating layer comprised predominantly Fe–Zn ferrite grains of ~10 μm and larger, while the subsurface metal exhibited a martensitic structure with primary austenitic grain boundaries. Under the current experimental conditions, the Fe–Zn ferrite reaction zone between the coating layer and the metal substrate, caused by diffusion of zinc into the base metal, was not prominent.
- For cracked specimens, some crack fronts headed to the primary austenitic grain boundaries of the base metal, resulting in the crack sidewalls being enriched in zinc. Cracking of the metal substrate was categorized as LME caused by liquid zinc generated in the coating layer.
- LME-induced cracking occurred when the coating layer formed during heating in the furnace was indicated to have a biphasic Fe–Zn ferrite/liquid zinc structure based on the binary Fe–Zn phase diagram. LME occurred during hot stamping when liquid zinc was present in the coating layer. We believe that the liquid zinc, once formed in the furnace, remained in the coating layer during hot stamping.

- At prolonged heating times, dilution of zinc occurred in the coating layer, probably due to the diffusion of iron from the base metal to the coating layer. On the contrary, the diffusion of zinc from the coating layer to the subsurface matrix was virtually negligible.
- The maximum crack depth increased with increasing total XRD intensity of δ and η phases of the coating layer, reaching 127 μm in the present test. Assuming that the XRD intensity of δ and η phases represents the amount of liquid zinc formed upon heating in the furnace, the crack depth increased with increasing amount of the liquid phase in the coating layer.

REFERENCES