MODELING CRACK INITIATION IN Al-Si COATING DURING HEATING/QUENCHING PHASE OF HOT STAMPING PROCESS

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ABSTRACT: In hot-stamping processes, Al-Si coating is generally applied on the steel substrate to avoid decarburization and to enhance corrosion resistance of the hot-stamped parts. However, during hot stamping, the AlSi coating fractures due to thermal and mechanical loads. This deteriorates the surface quality of the stamped parts, increasing tool wear and friction between the stamping tool and coated sheet metal. These cracks are generally initiated during the heating and/or quenching phase due to phase transformations and thermal loads. The initiation of the cracks in the coating can be largely influenced by the evolution of coating microstructure, i.e., intermetallic compounds- Fe\textsubscript{3}Al\textsubscript{5}, each of which has different thermal and mechanical properties. These intermetallic compounds are formed during the heating phase and grow in a natural order of increasing iron content in the layers close to the substrate-coating interface.

The goal of this study is to investigate the initiation of cracks in the coating during quenching stage due to thermal loads only. Heat treatment experiments are conducted on the Al-Si coated hot-stamping steel at different austenitization temperatures, dwell times and cooling rates. The distribution of voids/micro-cracks and intermetallic compounds in the coating are examined via digital microscopy and SEM/EDX measurements, respectively. A thermal-structural finite-element model is built to predict the crack initiation in Al-Si coating during quenching: the model accounts for the spatial distribution and mechanical properties of different intermetallic compounds. The results show large strain localization around the voids due to thermal loads during quenching, leading to micro-cracks towards the surface.

KEYWORDS: austenitization temperature, dwell time, coating fracture, voids, micro-cracks

1 INTRODUCTION

For high-temperature stamping operations of ultra-high strength steel, Al-Si coatings are preferred for their resistance to oxidation and to decarburization of the substrate at austenitization temperatures. With the substrate, the coating forms a continuous metallurgical bonding, which ensures proper sealing and efficient heat transfer \cite{1}. Unlike Zn coatings, Al-Si does not suffer from liquid-metal embrittlement during the heating process \cite{2}. However, since Al-Si has lower forming limits than the substrate steel \cite{3}, the coating fractures under thermo-mechanical stress during hot-forming and quenching. The coating fracture results in debris, which causes tool wear and high friction coefficient between the tool and the blank. The initiation of damage (i.e. voids) in the coating is triggered at the initial heating stage, after which those voids lead to cracks during the quenching and forming processes. In this article, we shall focus on the influence of thermal loading (heating followed by quenching) on Al-Si coating under hot-stamping conditions.

The heating stage in hot-stamping process is a crucial juncture because austenitization temperature and dwell time define the thermo-mechanical properties of the steel substrate and coating. Phase transformations in steel occur depending upon the level of heating followed by the rate of cooling. Likewise, the Al-Si coating layer evolves during the heating stage producing several intermetallic compounds as a function of austenitization temperature (T\textsubscript{AUS}) and dwell time (t\textsubscript{AUS}). During the heating step, Fe from the substrate diffuses into the coating, creating various Al\textsubscript{13}Fe\textsubscript{4} compounds, each with distinct thermal and mechanical properties \cite{4-6}.

Among the intermetallics, Al-rich compounds (Al\textsubscript{3}Fe, Al\textsubscript{13}Fe\textsubscript{4} and Al\textsubscript{5}Fe\textsubscript{3}) exhibit relatively lower fracture toughness than Fe-rich compounds (FeAl and Fe\textsubscript{3}Al) \cite{7, 8}. The latter predominates when Fe-diffusion is sufficient; i.e., at high T\textsubscript{AUS}.

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and/or long $t_{\text{AUS}}$. When subjected to plastic deformation, intermetallics like FeAl or Fe$_3$Al show a higher resistance to fracture than Al-rich compounds [9]. In terms of the thermal properties, although both intermetallics have similar thermal expansion coefficients ($\alpha$) at high temperature, the substrate shows relatively higher $\alpha$ [10-12].

During the heating stage, voids are formed at the coating-substrate interface. The formation of voids could be attributed to the Kirkendall effect, in which distinct diffusion coefficients of two differing metals (Fe and Al) [13] result in a net flux of vacancies [14]. Since the diffusivity of Fe is ~14 times larger than that of Al, voids are formed at the interface. The density of Kirkendall voids depends on $T_{\text{AUS}}$ and $t_{\text{AUS}}$, the voids being delimited along the diffusion zone (substrate-coating interface) [15].

On the other hand, the formation of micro-cracks are due to thermal loading and dissimilar thermal expansion coefficients of the evolved intermetallics [16], along with the substrate. Since the formation of intermetallic compounds requires Fe-diffusion, it is assumed, in this study, that the micro-cracks would only exist after the development of intermetallics and voids. Therefore, it is very likely that the micro-cracks are formed during the quenching stage i.e., after heating.

The aim of this work is to investigate the initiation of micro-cracks in Al-Si coating at air-quenching step. The effect of thermal loads on the morphology of coating is examined by altering $T_{\text{AUS}}$, $t_{\text{AUS}}$ and cooling rate in heat treatment experiments. Then, a microscopic through-thickness inspection of the sample is conducted to find the extent of damage and distribution of intermetallics throughout the coating layer. A numerical model is used to investigate crack initiation in the coating during quenching.

2 HEAT TREATMENT EXPERIMENTS

2.1 MATERIALS & METHODS

The Al-Si coating, used for this investigation, consists of 87% aluminum, 10% silicon and 3% iron. Uncoated hot-stamping steel (22MnB5) is hot-dipped into a bath of molten Al with 10 wt.% Si, the average thickness of the coating is ~25 $\mu$m (Fig.1).

The heating step is controlled by austenitization temperature ($T_{\text{AUS}}$) and dwell time ($t_{\text{AUS}}$), while the quenching step by cooling rate (Fig.2). Thermocouple wires were spot-welded to a rectangular sheet specimen, which was then inserted to a pre-heated furnace. The heating rate of the sample could not be controlled since it depends on the furnace temperature, $T_{\text{AUS}}$. The $t_{\text{AUS}}$ begins as the specimen temperature reaches the furnace temperature. For air-quenching, the specimen was taken out of the furnace and suspended inside a cooling fixture through which constant air flow was maintained from a fixed distance. As a result, the cooling rate was altered by varying the air pressure in each case. The temperature of the specimen was recorded during the entire process. The experiments were designed based on varying one parameter at a time to capture the effect of $T_{\text{AUS}}$, $t_{\text{AUS}}$ and cooling rate. After quenching, the cross-sections of the specimen were polished, etched in 4 $\mu$m diamond solution, inspected under an optical microscope and analyzed using SEM/EDX.

2.2 RESULTS

Figure 3 shows the micrographs of Al-Si coating as a function of $T_{\text{AUS}}$, $t_{\text{AUS}}$ and cooling rate. It is evident that the density of cracks and voids increases with increasing $T_{\text{AUS}}$, only density of voids increases with increasing $t_{\text{AUS}}$, while the cooling rate does not affect the crack-void density. Using high cooling rate (50 K/s) via air-quenching, some chunks of coating were blown away due to high air mass flow on the coating surface.

2.3 DAMAGE CHARACTERIZATION OF AI-SI COATING

After heating and quenching steps, voids and cracks were found in the coating. We already speculated that voids were formed due to Fe-diffusion during the heating stage while the cracks were generated afterwards, presumably during
quenching because of existing voids and thermal expansion mismatch between the intermetallics and substrate. In order to characterize damage in the coating, the voids are quantified separately from the micro-cracks (Fig. 4).

In order to have a consistent comparison, an area of interest (AOI) is segmented from the optical images, such that the bubbles on the coating surface that are formed due to surface oxidation were neglected. The AOI is then binarized to segment voids and micro-cracks (Fig. 4). The total fraction of voids and micro-cracks is calculated by taking the ratio of black pixels to total pixels. Damage due to voids is then evaluated via an ellipse-based algorithm, where voids represent ellipses with eccentricity in the range of 0.8-1.0. Using this technique, the voids fraction is measured from each micrograph. Furthermore, since the cracks and voids accumulate to the total damage, the so-called crack fraction is, thereby, calculated by subtracting the void fraction from the total damage.

The homogeneous length scale of AOI to quantify the void/crack fraction is determined by performing sensitivity analysis. The length of AOI was varied between 0.2 and 2 mm (Fig. 5). For each parameter, the crack+void fraction was measured for different coating lengths (0.2, 0.5, 1 and 2 mm). It was found that the fraction converged from a coating length of 0.5 mm onwards. For a 2-mm analyzed coating length, figure 6 shows the representative void-crack fractions as a function of \( T_{AUS} \), \( t_{AUS} \) and cooling rate.

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**Fig. 3** Evolution of Al-Si coating after different (a) \( T_{AUS} \) (b) \( t_{AUS} \) and (c) cooling rate

**Fig. 4** Quantification of voids and cracks in Al-Si coating

**Fig. 5** Crack+void fractions for different coating lengths with change in (a) \( T_{AUS} \) (b) \( t_{AUS} \) and (c) cooling rate
Since the composition of intermetallics is essential to define the coating layers, a SEM/EDX line probe test was performed along the cross-section of the coating. Scanning electron microscopy with electron-dispersive SEM/EDX (Leo 438VP tungsten filament, EDAX liquid nitrogen cooled detector system) with an accelerating voltage of 15 kV was used for this purpose. In this paper, the EDX results corresponding to one austenitization temperature ($T_{\text{AUS}}=1020^\circ\text{C}$) under constant dwell time ($t_{\text{AUS}}=6$ mins) and cooling rate ($\dot{C}=40$ K/s) are presented (Fig.7). The atomic percentage acquired from the EDX probe show that for $T_{\text{AUS}}=1020^\circ\text{C}$, Fe-rich compound (AlFe) predominates in greater proportion than Al$_5$Fe$_2$ in the coating.

3 NUMERICAL QUENCHING MODEL OF Al-Si COATING

3.1 SIMULATION SETUP

A thermal finite element model was setup in MSC Marc to investigate the effect of quenching process in the coating. During the heating stage, although various compounds were formed in the coating layer due to Fe-diffusion, the intermetallics of Al$_5$Fe$_2$ and AlFe stabilize and predominate in greater fraction than other intermetallics throughout the coating layer [17-19]. Therefore, the thermal model was framed assuming that Fe-diffusion transforms the coating into layers of Al$_5$Fe$_2$ and AlFe, the latter being situated near the substrate interface. Furthermore, due to the intrusion of Fe in the coating, the thickness of coating was increased to ~40 μm, which was also taken into account.

The substrate (22MnB5) is modeled as an elasto-plastic material, with temperature dependent Young’s modulus [20] (Fig.8), yield stress and strain hardening at a constant strain rate of 0.01/s [21]. The thermal properties of steel and intermetallic compounds- specific heat capacity ($C_\text{p}$) [10, 22, 23], thermal conductivity (k) [20, 22] and thermal expansion coefficient ($\alpha$) [10, 11] are shown in figure 9. In case of coating, AlFe and Al$_5$Fe$_2$ are also modeled as elasto-plastic materials, the former being perfectly plastic and the latter showing strain hardening in the temperature range of 500-1000°C. The elasticity of the Al$_5$Fe$_2$ intermetallic was derived by extrapolating Young’s modulus data of AlFe, based on the atomic percentage of Al [24]. The yield stress of AlFe was acquired from Baker and Munroe [25] while that of Al$_5$Fe$_2$ was extracted from Hirose, Itoh [26] (Fig.8). An equivalent strain-based failure criterion was imposed to the intermetallic elements such that Al$_5$Fe$_2$ becomes brittle (i.e. fails at $\sigma_y$) below 500°C [26] whereas AlFe follows increased elongation [27] with rise in temperature (Fig.10).
In the quenching simulation, the plane strain elements were initially kept at $T_{\text{AUS}}$, after which temperature history of the coating surface acquired from the experiments was applied on the coating surface to mimic the nature of forced convection heat loss of the experiment (Fig.10). Periodic and symmetry boundary conditions were also applied on the sides and half-thickness, respectively. Furthermore, since voids and intermetallic layers were assumed to have been formed already during the heating stage because of diffusion, the Kirkendall voids were plotted along the interface, according to the measured representative void fraction at $T_{\text{AUS}}$ $= 1020$ °C (Fig.6). In addition, the intermetallics were distributed based on the atomic percentage of Al and Fe (Fig.7) obtained from EDX measurements. Taking the voids and atomic composition into account, the material is distributed, as shown in figure 11, with an element size of 0.25 µm at the vicinity of voids.

3.2 SIMULATION RESULTS & DISCUSSION

The preliminary simulation results show that the thermal expansion coefficient mismatch between substrate and coating coupled with brittle intermetallic compounds caused strain localization around the voids (Fig.12). During the quenching simulation, the material contracts and applies plastic strain to the voids. In particular, from 200 to 50°C, the intermetallics showcase negligible plasticity. Thus, at ~130°C, the localization of strains around the voids initiates longitudinal micro-cracks, which propagates until it reaches the coating surface.

Multiple micro-cracks emerge mainly from voids, reducing the stresses acting around them and also around other voids (Fig.13b). The micro-crack propagates in a vertical direction towards the Al$_5$Fe$_2$ layer, at which the micro-crack tip partially continues along the interface between Al$_5$Fe$_2$ and AlFe. Eventually, one micro-crack reaches the coating surface, causing stress relaxation on Al$_5$Fe$_2$ layer and also on the other leading micro-crack. That’s why, the other micro-crack reaches as far as the AlFe-Al$_5$Fe$_2$ interface (Fig.13c).
At the closing stage of quenching simulation (from 300 to 50⁰C), the crack initiation in the coating layer was sensitive to the intermetallic fracture strains. Therefore, the elongation to fracture of the intermetallics should be acquired with greater precision.

4 CONCLUSIONS

This work provides an extensive analysis of Al-Si coating (on UHSS) during heating as a function of austenitization temperature, dwell time followed by quenching as a function of cooling rate. For a particular heating condition ($T_{AUS} = 1020^\circ C$ and $t_{AUS} = 6$ mins), a quenching simulation (with voids) was performed to analyse the strain and crack profiles in the coating. The experimental and simulation results indicate the following:

- The magnitude of austenitization temperature and dwell time determine the extent of damage (i.e. voids) in Al-Si coating.
- Increased cooling rate (via air-quenching) shows negligible effect on coating damage.
- A coating length of at least 0.5 mm is required to measure the representative damage fraction from the micrographs.
- A quenching simulation with Kirkendall voids shows strain localization at the vicinity of the voids.
- Quenching simulation results suggest micro-cracks emanating from the existing Kirkendall voids.

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