Rapid Solidification Microstructure and Carbide Precipitation Behavior in Electron Beam Melted High-Speed Steel

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The solidified microstructure and carbide precipitation behavior in an S390 high-speed steel processed by electron beam melting (EBM) have been fully characterized. The as-EBM microstructure consists of discontinuous network of very fine primary carbides dispersed in auto-tempered martensite matrix together with a limited amount of retained austenite. The carbide network consists of M_2C/M_6C and MC carbides. Both the columnar and near-equiaxed grain structures were found in as-EBM microstructure and the presence of inter-dendritic eutectic carbides assisted in revealing the dendritic solidification nature. The top-layer microstructure observation confirmed that the columnar dendritic structured grains were located adjacent to the micro-melt pool boundary, indicating an epitaxial growth with the average growth direction parallel to the maximum thermal gradient. At the center of the micro-melt pool, the near-equiaxed grains were developed by dendritic growth parallel to the beam traveling direction. The carbide decomposition was revealed by scanning transmission electron microscopy and confirmed by transmission Kikuchi diffraction. The MC carbides (rich in V followed by W) nucleated at the interface between M_2C (W, Fe, Mo, and Co in the order of significance) and the matrix and then grew from the outside inward, but their nucleation might occur from the M_2C carbide itself. The thermal effect induced by the adjacent scan lines seems to trigger a solid-state phase transformation of MC → M_2C + γ-Fe. The elemental migration was theoretically calculated and compared with the experimental results. The high hardness of ~ 65 HRC and good transverse rupture strength of ~ 2500 MPa in as-EBM S390 means that EBM processing can be used to fabricate highly alloyed tool steels. With the help of the post-processing heat treatment, the best Rockwell hardness of 73.1±0.2 HRC and transverse rupture strength of 3012±34 MPa can be obtained.

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I. INTRODUCTION

Compared to conventional ingot casting and powder metallurgy, electron beam melting (EBM) has successfully demonstrated its capability to fabricate near-net-shape parts with the very fine microstructure and minimum segregation due to its intrinsic rapid solidification. EBM belongs to the group of additive manufacturing (AM). In contrast to selective laser melting (SLM) that operates at lower temperatures, EBM can work at high temperatures of up to 1200 °C. This in situ heating characteristic makes EBM particularly attractive for processing high-performance alloys that are susceptible to thermal stress-induced cracking or solidification/hot cracking. For example, EBM was applied to produce precipitation-strengthened nickel-base superalloys, cobalt-base alloys, as well as titanium aluminide intermetallic alloys. However, little effort has been made regarding EBM high-speed steels that are also susceptible to cracking. Note the EBM work on H13 steel that has a carbon content of 0.37 wt pct and limited degree of alloying does not represent those highly alloyed high-speed steels.

Up until now, there have been few publications on high-speed steels fabricated by SLM and blown powder AM. The carbon content in those steels (including M_2, M_50) and a high-strength Fe85Cr4Mo8V2C1 were all below 1.0 wt pct, except for a very recent work on an SLM M3:2 steel. With
the increasing carbon content, higher hardness and wear resistance can be achieved, but this poses challenges for processing due to material susceptibility to cracking. This paper focuses on S390 high-speed steel containing 1.64C-10.12W-2.00Mo-4.97V-4.90Cr-7.86Co (all in wt pct). Compared to M2 steel containing 0.90C-6.15W-4.89Mo-1.82V-3.97Cr, the S390 steel has a higher carbon content and a greater degree of alloying. For carbon content of higher than 1.3 wt pct, the primary solidification phase is austenite, as opposed to delta-ferrite. In this sense, revealing the solidification sequence and primary carbide precipitation behavior in S390 steel, a highly alloyed high-carbon steel, during the EBM process helps advancing non-equilibrium solidification theories.

High-speed steels generally have good hardenability due to their high alloy contents. For such steels when processed using ingot casting, the presence of coarse primary carbide network that formed during solidification is inevitable. The presence of these carbides and their decomposition during hot forging and hardening/tempering heat treatment is detrimental for mechanical properties of the final part.\(^\text{[10]}\) Powder metallurgy route (e.g., hot-isostatic-pressing) has been successfully used to fabricate high-speed steels, but a complex post-processing is required to achieve desired properties. For example, the heat treatment cycles for an S390 steel include (i) pre-heating of first to 650 °C, then 850 °C, and finally to 1050 °C; (ii) heating to the austenitizing temperature of 1130 °C and soaking, followed by quenching in vacuum environment; (iii) cryogenic temperature treatment; and (iv) triple tempering at 540 °C.\(^\text{[20,21]}\) Bearing all these complexity in mind, it is difficult to obtain the optimum microstructure and mechanical properties with the conventional means. In contrast, the intrinsic heat treatment capability in EBM process could make it as an alternative approach to fabricate high-speed steels with a very good starting microstructure, hence simplifying the subsequent heat treatment procedure.

The high-speed steels in the post-processed condition (i.e., conventional ingot casting followed by hot rolling and full heat treatment cycles) exhibit their excellent hot hardness due to a combination of matrix and precipitation strengthening mechanisms. The microstructure of these steels consists of coarse primary carbides (1 to 2 \(\mu\)m in diameter) dispersed in a tempered martensite matrix, which is further strengthened by much finer secondary carbides (less than 20 nm) which precipitate out during tempering.\(^\text{[22]}\) Despite the large number of high-speed steel types, the sequence of phase transformations and precipitation reactions which occur across different alloys were found to be similar.\(^\text{[23]}\) The precipitation reaction occurs with certain alloy elements that have affinity for carbon. These elements include vanadium (V), tungsten (W), chromium (Cr), molybdenum (Mo), and iron (Fe).

Depending on alloy contents and the solidification path, the formed eutectic carbides include \(\text{M}_2\text{C}, \text{MC}, \text{M}_3\text{C}\), and \(\text{M}_6\text{C}\). \(\text{M}_2\text{C}\) exhibits an even distribution of V, Mo, W, and Fe, whose contents are approx. 20 at. pct.\(^\text{[24]}\) \(\text{M}_2\text{C}\) carbide is a metastable phase and often transforms during high-temperature exposure to MC and \(\text{M}_6\text{C}\) carbides which are thermodynamically more stable.\(^\text{[23,25,26]}\) MC is rich in V but could also contain a significant number of Mo, while \(\text{M}_6\text{C}\) contains high amounts of, in order of magnitude, Fe, Mo, and W.\(^\text{[22]}\)

The main compositional difference between \(\text{M}_2\text{C}\) and \(\text{M}_6\text{C}\) carbides is that the former contains higher Fe.\(^\text{[22,23,27]}\) The effect of cooling rate on carbide formation in high-speed steels can be manifested by their varying composition, volume fraction, and type.\(^\text{[25,26]}\) The MC carbide with different compositions, observed within the same specimen, was attributed to the wide temperature range through which the crystallization of carbides from the melt proceeded.\(^\text{[26]}\) Furthermore, the \(\text{M}_6\text{C}\) carbide can be replaced by \(\text{M}_2\text{C}\) with the higher cooling rates.\(^\text{[25]}\)

In the present work, S390, a highly alloyed powder metallurgical (PM) high-speed steel, was fabricated by EBM. Both the microstructure and mechanical properties in as-EBM condition were assessed using a range of complementary characterization tools. The results are discussed with respect to those already published similar, if not the same grade high-speed steels that were processed by ingot casting, PM, or spray-forming (i.e., more established routes).

II. EXPERIMENTAL

A. S390 Steel Powders

Gas atomized S390 steel powders with the particle size range of primarily between 45 and 140 \(\mu\)m were used for the EBM build. The powder size distribution is illustrated in Figure 1(c) and particle size data of D10, D50, and D90 were determined as 45.1, 80.1, and 140.7 \(\mu\)m, respectively. Representative scanning electron microscopy (SEM) micrographs of atomized powders are shown in Figures 1(a) and (b). In general, the powder particles are spherical, Figure 1(a), albeit the presence of small satellites and rod-shaped ones, Figure 1(b). Table I summarizes the chemical composition of S390 steel powders that contains 1.64C-10.12W-2.00Mo-4.97V-4.90Cr-7.86Co (wt pct). Inductively coupled plasma atomic emission spectroscopy (ICP-AES) was used for the measurements. The powder composition is consistent with the commercial Böhler Microclean PM S390 steel, Table I.

B. EBM Sample Fabrication

EBM S390 steel samples were fabricated in an Arcam A2XX machine with Control Software 3.2 in manual mode. One batch of the EBM fabrication consisted of 12 sample blocks, each with dimensions of 20 \(\times\) 20 \(\times\) 37 mm. EBM samples were built on a stainless-steel start plate of 10 mm thick and 150 mm diameter. Samples were processed using a range of scanning speed \(v\) of 470, 700, and 1400 mm s\(^{-1}\), line offset \(L_{\text{off}}\) of 0.05, 0.10, and 0.15 mm, as well as beam current \(I\) of 5.5 and 4.5 mA. These process parameter sets were previously optimized to obtain good quality EBM S390 steel. They were all within the EBM process window, i.e., no melt ball
formation, smoke or powder splash. Layers were scanned using a snake-like melt strategy (i.e., 0 or 90 deg hatch direction alternation). More details about the EBM sample fabrication can be found elsewhere [4,10].

Table II summarizes these process parameters together with the calculated area energy $E_A$, where $E_A = (IU)/(vL_{off})$. $U$ is the operating voltage and 60 kV was used. Samples 1 and 2 had a similar $E_A$ but $v$ and $L_{off}$ were changed proportionally. This allowed us to assess the effect of $v$ on the as-EBM microstructure. Samples 3 and 4 were fabricated with different $L_{off}$; hence, a different $E_A$ can be seen in Table II.

Figure 2(a) illustrates the temperature history during EBM fabrication that includes pre-heat, melt, and final cooling upon completion. The build temperature was measured using a thermocouple attached to the bottom of the plate. The start plate was pre-heated to a temperature of 820 °C and the whole EBM fabrication was performed at the build temperature of approx. 780 °C for 4 hours, Figure 2(a). This indicates that the EBM process temperature was above the martensite starting temperature of 351 °C.[28] Upon build completion, the samples were cooled slowly within the EBM build chamber. The relative density of as-EBM samples was measured using Archimedes drainage method.

C. Microstructural Characterization

As-EBM samples were sectioned using wire electrical discharge machining to four identical pieces per sample. Each specimen had a dimension of 10 × 7 × 15 mm as illustrated in Figure 2(b) and the microstructural characterization was performed close to center of the EBM build (away from the contour melting areas) at a similar Z height position. In terms of metallographic preparation, specimens were subjected to grinding and polishing down to 0.5 μm, followed by either a vibro-polishing with OPS colloidal silica for 6 hours or a chemical etching with 4 pct nital for 2 minutes. A FEI Quanta 200F FEG-SEM was used for microstructural
observation. The volume/area fraction of carbides and their sizes were measured by quantitative metallography with the assumption that determined 2D mean size from circular cross-sections represents the 3D mean size. A JEOL JSM-7001F SEM equipped with Pegasus XM2 electron backscatter diffraction (EBSD) detector was used to obtain grain orientation maps. A step size of 0.2 μm was used to collect EBSD data and a large field-of-view area of 180 × 180 μm that included >1000 grains was examined. The HKL CHANNEL5 software was used to analyze both the average grain size (defined by a misorientation angle of >15 deg) and micro-texture. The noise reduction was applied prior to the grain size measurement as this helped to minimize the effect of inter-dendritic eutectic carbides on individual grain identification.

The submicron primary carbides observed in as-EBM samples under SEM were further characterized by using scanning transmission electron microscopy (STEM), JEOL JEM-2800 equipped with energy-dispersive X-ray analysis (EDX). Bright-field imaging, selected area diffraction (SAD), as well as compositions were collected from both the matrix and carbide precipitates. The TEM sample was prepared by mechanical polishing both sides to approx. 30 μm in thickness and then cutting out 3 mm disk samples. This was followed by Ar− ion milling with 5 kV by using Gatan 691 PIPS. The thin-foil TEM sample was further investigated using a Zeiss Supra 40 VP FEG-SEM equipped with EBSD detector. Phase constitution maps were obtained using Transmission Kikuchi Diffraction (TKD). An accelerating voltage of 30 kV in high current mode and an aperture size of 120 μm were used to maximize signal-to-noise ratio. Mapping was performed at a working distance of 3.5 mm, with a step size of 5 nm, a sample pre-tilt angle of 20 deg, and a dwell time of 0.18 seconds per pattern. More details about the use of TKD technique can be found in our previous work [29,30].

X-ray diffraction (XRD) was used to provide complementary information for phase identification. A Rigaku D/Max-2000pc diffractometer with Cu Kα radiation with a wavelength of 1.5418 Å was used. All the XRD measurements were performed on polished cross-sectional X–Z plane with a 2θ range from 30 to 70 deg.

D. Hardness and Transverse Rupture Strength

Rockwell HRC hardness measurements were performed using HR-150DT Electric Rockwell Hardness Tester. 150 kg force with the retention time of 10 seconds was used. Fifteen individual measurement points were made per sample and the average value together with standard deviation (STDEV) is reported.

Three-point bend tests were carried out to determine the transverse rupture strength. Dimensions of the test specimens were 5 × 5 × 25.4 mm (Z = 25.4 mm). Sample surfaces were polished down to a SiC grit size of 1000. An E45 SANS 50 kN electro-mechanical testing machine was used with a displacement rate of 0.1 mm/ min. The bending force was applied perpendicular to the Z-build direction. Three tests per sample condition were performed to ensure reproducibility. The transverse rupture strength was calculated based on the measured bend fracture force and the average value is reported. Specimen dimensions, test procedure, and calculations all followed the recommendation given in ASTM B528-16. This is a standard test method for PM materials. SEM fractography examination was performed to reveal the fracture mode.

III. RESULTS AND DISCUSSION

A. Density and Bulk Chemical Composition

Table III summarizes the relative density of as-EBM samples 1 to 4 with respect to the fully dense solid S390 that has a density of 8.1 g cm−3. Sample 2 that was fabricated with the lowest scanning speed of v = 470 mm s−1 exhibited the highest relative density of ~97.7 pct. The other three samples had a similar value, Table III. SEM micrograph in Figure 3 shows the presence of spherical-shaped gas pores with a typical size of 45.1 ± 10.6 μm in sample 1, which is the
representative of the other three samples. The presence of gas pores in as-EBM samples is very likely to be a result of gas atomization process and no lack-of-fusion defects were found. In addition, there was little change in alloy compositions when compared to the pre-alloyed powders, Table I. This is important as the primary carbide precipitation behavior is strongly influenced by the exact composition of the alloy steel.

B. Solidification Microstructure

Figures 4(a) through (d) show the typical microstructures of samples 1 to 4, respectively. The carbide distribution (i.e., a discontinuous network of carbides) is characteristic of eutectic solidification. The enlarged view in Figures 4(a) and (b) highlights that the eutectic carbide area tended to be larger for sample 1 with a higher scanning speed of \( v = 1400 \text{ mm s}^{-1} \) when compared to sample 2 with \( v = 470 \text{ mm s}^{-1} \). The eutectic cluster size of sample 1 was measured as \( 3.2 \pm 0.1 \text{ \mu m}^2 \), while that for sample 2 was \( 2.2 \pm 0.2 \text{ \mu m}^2 \). The morphology of the matrix phase can be characterized as a combination of near-equiaxed and columnar grains in shape. The typical regions for columnar grains are outlined by cyan color in Figures 4(c) and (d). The area fraction of the columnar grains was smaller for sample 3 (20.3 pct) when compared to sample 4 (48.7 pct). A very small amount (< 10 pct) of columnar grains was found in samples 1 and 2; both samples had a higher \( E_A \) value compared to sample 4. This observation suggests that a smaller \( E_A \) promotes a higher proportion of columnar grains.

Figures 5(a) and (b) show the EBSD orientation map and pole figures for sample 3. No preferred texture for the matrix phase can be found. This is consistent with the previous work on SLM M2 high-speed steel, where EBSD results revealed that the matrix grains had little micro-texture. EBSD scans were performed on the other three samples 1, 2, and 4 to clarify whether different EBM melt parameters applied would result in any preferred texture. EBSD orientation maps are presented in Figures 6(a) through (c). As evidenced by the very weak pole figure intensity (the maximum intensity of less than 2), Figures 6(d) through (f), none of the EBM S390 samples exhibited a strong texture.

The EBSD pattern of the steel matrix was best matched by the body-centered cubic bcc-Fe in the majority. This means that the primary matrix phase in as-EBM S390 steel is either auto-tempered martensite or ferrite, or a combination of these two. It is referred as auto-tempering as the EBM process offers the opportunity of tempering during the part fabrication. In addition, 5 pct volume fraction of face-centered cubic fcc-Fe was found in as-EBM S390 steel matrix, indicating a small amount of retained austenite.

Table II summarizes the measured grain sizes (bcc-Fe) for all as-EBM samples, in the order of increasing, sample 4 (\( E_A = 3.1 \text{ J mm}^{-2} \)) with 1.69 ± 0.53 \( \mu m \), sample 3 (\( E_A = 4.7 \text{ J mm}^{-2} \)) with 1.74 ± 0.53 \( \mu m \), sample 1 (\( E_A = 4.7 \text{ J mm}^{-2} \)) with 1.90 ± 0.57 \( \mu m \), and sample 2 (\( E_A = 4.7 \text{ J mm}^{-2} \)) with 2.08 ± 0.59 \( \mu m \). It seems that the grain size is not strongly dependent on the EBM parameters considered here.

XRD spectra collected from all four samples are shown in Figure 7. Characteristic peak positions for bcc-Fe, body-centered tetragonal bct-Fe, and fcc-Fe obtained from ICDD-PDF cards are also indicated in Figure 7. The presence of characteristic XRD peaks for bcc-Fe, MC, \( \text{M}_2\text{C} \), and \( \text{M}_6\text{C} \) can be easily identified. But it was unable to discern the characteristic peaks associated with the martensite bet-Fe. Furthermore, there might be a small characteristic XRD peak at 20 angle of ~ 43 deg for the retained austenite, but the peak seems to be overlapped with that of MC carbide.

| Sample ID | \( v (\text{mm s}^{-1}) \) | \( E_A (\text{J mm}^{-2}) \) | Relative Density (Pct) | HRC Hardness | Transverse Rupture Strength (MPa) |
|-----------|----------------|-----------------|----------------------|-------------|-----------------------------------|
| 1         | 1400           | 4.7             | 95.7                 | 65.5 ± 0.5  | 2391.3 ± 100.3                    |
| 2         | 470            | 4.7             | 97.7                 | 65.3 ± 0.7  | 2601.6 ± 161.6                    |
| 3         | 700            | 4.7             | 95.6                 | 65.0 ± 0.7  | 2445.1 ± 126.6                    |
| 4         | 700            | 3.1             | 95.5                 | 65.1 ± 0.7  | 2561.9 ± 86.3                     |

Fig. 3—A representative BSE SEM micrograph showing the presence of gas pores in as-EBM sample 1. The sample was polished without further etching.
Fig. 4—BSE SEM micrographs of as-EBM S390 steels showing the distribution of primary carbides: (a) sample 1 with $v = 1400$ mm s$^{-1}$; (b) sample 2 with $v = 470$ mm s$^{-1}$ together with enlarged views highlighting different sizes of the eutectic carbide regions in these two samples; (c) sample 3 with $E_A = 4.7$ J mm$^{-2}$ and (d) sample 4 with $E_A = 3.1$ J mm$^{-2}$. Note: Samples were polished without further etching.

Fig. 5—(a) The EBSD orientation map with respect to the ND (Y-direction) on sample 3; (b) the corresponding pole figures showing a texture-free in as-EBM condition. The EBSD color coding applies to both Figs. 5 and 6 IPF orientation maps (Color figure online).
C. Carbide Precipitation and Volume Fraction

Backscattered electron BSE SEM images at high magnification, the insets of Figures 4(a) and (b), revealed two distinct types of primary carbides: (i) the brighter white phase suggests that these carbides contain higher-atomic-number elements (e.g., W) and (ii) the darker gray phase suggests that these carbides contain lower-atomic-number elements (e.g., V). The brighter phase is likely to be M$_2$C/M$_6$C carbides, while the darker phase is likely to be MC carbide. The chemical composition and crystal structure of these primary carbides were determined later by STEM-EDX and TEM-SAD.

Figure 8(a) shows the discontinuous network of eutectic carbides that consists of rod-shaped M$_2$C/ M$_6$C carbides and irregular shaped MC carbides. The size of MC carbides ranged from a few tens to hundreds of nanometers* and they co-existed with M$_2$C/M$_6$C.

*The accurate size measurement for each individual particle cannot be carried out due to their irregular shape. In addition, both the TEM and TKD are judged to be too high resolution to provide a statistically meaningful size measurement.
and carbides to form cell boundaries. It is well documented that $M_2C$ carbide is a metastable phase that can decompose into a mixture of $M_6C$ and MC carbides during high-temperature exposure. The carbide size observed in the EBM samples is much smaller compared to that of ingot casting and similar to spray-forming. For comparison purpose, the microstructure of a PM S390 steel sample in the fully heat-treated condition is shown in Figure 8(b). It can be seen that the primary carbides in PM S390 steel are much larger than those in the EBM processed one. The refined microstructure in EBM S390 steel can be attributed to the non-equilibrium rapid solidification during the EBM process.

The volume fraction of primary carbides was determined by examining a large number of carbides (i.e., several SEM images similar to the one as shown in Figure 8(a)), and the results are summarized in Table IV. Z-contrast under BSE imaging mode was used to distinguish the MC carbide from the $M_2C/M_6C$ carbides. The total carbide fraction does not seem to change with different EBM process parameters, although their relative proportion is different for samples 1 to 4. In particular, sample 4 with the smallest $E_A = 3.1 \text{ J mm}^{-2}$ had the lowest relative proportion of $M_2C/M_6C$-type carbides (24.8 pct), compared to sample 1 (41.1 pct), sample 2 (47.4 pct), and sample 3 (37.6 pct). The relative proportion was calculated by dividing the measured $M_2C/M_6C$ volume fraction by the total primary carbides. The carbide volume fraction in the heat-treated PM S390 steel is also given in Table IV. It seems that the volume fraction of primary carbides in the present EBM S390 steel is higher than that in PM S390 steel, indicating a very effective in situ heat treatment that already occurred during the EBM process.

D. Matrix Phase Identification

The TEM micrograph shown in Figure 9(a) indicates corresponding locations where SAD patterns were collected. Taking an example of the inset in Figure 9(a), the interplanar spacing was measured to be $d_{110} = 2.0338 \text{ Å}$, that is within 1 pct difference to the theoretical value of 2.0268 Å for bcc-Fe. The theoretical value for bct-Fe $d_{110}$ was reported as 2.0128 Å, hence not preferred here. The tetragonality ($c/a$ ratio) of the matrix phase was determined as 1.03, which seems to indicate a bct-Fe. We thereby denote bct-Fe in the SAD pattern in Figure 9(a). Note the $c/a$ ratio for an unambiguously identified martensite phase should be between 1.02 and 1.08.

An attempt was made to examine whether EBSD band contrast can be used to differentiate between the bcc-Fe ferrite and bct-Fe martensite. Figure 10(a) presents the band contrast EBSD map in which the darker region (i.e., poor band contrast) indicates heavily strained grains, whereas the lighter region indicates less strained ones. The interior of those grains with a small degree of deformation reveals very limited number of subgrain boundaries, Figure 10(b). They were confirmed as retained austenite fcc-phase based on EBSD phase identification. Therefore, for the majority of grains that have the poor band contrast shown in Figure 10(a), they are most likely the bct-Fe martensite. Combining experimental results from TEM-SAD and EBSD band contrast, we may conclude that the primary matrix phase is auto-tempered martensite which has very little tetragonality left.

For carbon content lower than 1.3 wt pct (e.g., M2 steel, 0.9 pct), $\delta$-Fe ferrite is the initial solidification phase. But with the increasing carbon content (e.g., S390 steel, 1.48 pct in the bulk sample in Table I), primary austenite would form from the liquid followed by partial phase transformation to martensite during cooling, resulting in the matrix that is composed of retained austenite and martensite. During the EBM part fabrication, the temperature of the material close to the top layers would be much higher than the austenite transformation temperature of 837 °C and a lot of heat would be released from the top to the bottom of the part. Since the build temperature as measured by the thermocouple attached to the start plate was approx. 780 °C, this
indicates that during the part fabrication, the already solidified material was subjected to a high-temperature within the austenite-phase region. As a result, the formation of auto-tempered martensite most likely happened during the final cooling.

The TEM-SAD result as shown in Figure 9(b) confirms the presence of submicron-sized steel matrix phase with an fcc crystal structure. This fcc-phase has a lattice parameter of \( a = 3.90 \) Å. According to Reference 32, a solid solution of 3 at. pct Mo in austenite would only increase its lattice parameter by 0.01 Å. Furthermore, the residual lattice strain that would be required to cause such a large lattice expansion was calculated to be 83333 le, which differed by a factor about 100, i.e., two orders of magnitude higher than what is expected. According to our neutron diffraction measurements on
steel samples, the magnitude of residual lattice strains should be in the range of a few hundred microstrains. As a consequence, such a large lattice expansion cannot be simply attributed to residual stresses due to rapid cooling.

According to References 24, 28, 35, and 36 that cover a wide range of high-speed steels, only three phases with the fcc crystal structure should be expected; they are retained austenite, MC, and M₆C carbides. The MC carbide has a lattice parameter of $a = 4.13$ Å (ICDD-PDF 89-5055), while $a = 11.09$ Å for M₆C carbide (ICDD-PDF 78-1990). Therefore, it is unlikely that the observed phase in Figure 9(b) with $a = 3.90$ Å is the M₆C carbide, neither for MC carbide as this type of carbide agreed very well with the SAD pattern as shown in Figure 9(d). Finally, STEM-EDX elemental mapping confirmed that this fcc-phase is rich in Fe; hence, it is a matrix phase.

E. Carbide Phase Identification

The SAD patterns in Figures 9(c) and (d) were analyzed to determine the interplanar spacing. We thereby confirmed the presence of M₂C carbide (Figure 9(c)) and MC carbide (Figure 9(d)). It is noteworthy to mention that the hexagonal M₂C carbide identification was based on the three-fold interplanar spacing as shown in Figure 9(c) and the difference between the calculated lattice parameters (a and c values) and the theoretical values was within 10 pct.

The STEM-EDX elemental mapping was performed to understand the carbide decomposition behavior. Figure 11(a) shows overlay maps of Fe, V, and Mo and their individual elemental maps are presented in Figures 11(b) through (d). The W elemental distribution was very similar to Mo and hence it is not shown for brevity. As expected, the M₂C carbide is rich in W, V, and Mo but lean in Fe compared to the matrix. The decomposed M₂C carbide region is highlighted in Figure 11(a), where the V-enriched MC can be seen at the interface between the M₂C and bcc-Fe matrix, Figure 11(c). This suggests that MC carbide nucleated at the interface between M₂C and bcc-Fe and then the growth occurred from the outside inward. In addition, the presence of much smaller-sized discrete MC carbides (rich in V region at the upper left area of Figures 11(a) and (c)) indicates that MC carbide nucleation may also take place from the M₂C carbide itself.

The carbide decomposition behavior was further validated by TKD analysis of the thin-foil TEM. As shown in region 1 of Figure 12, a V-enriched MC carbide is adjoined to M₂C carbide, indicating the presence of carbide decomposition during EBM process. It is clear that the elemental distribution within the M₂C carbide is not homogeneous, exhibiting Mo-enriched and W-enriched areas. In addition, the presence of M₆C carbides in both regions 2 and 3 of Figure 12 provides another evidence for the carbide decomposition that occurred during EBM process. Since the M₆C carbides were revealed from the TEM and their XRD characteristic peaks were relatively small (Figure 7), their content in the present EBM S390 steel is judged to be very low. This seems to be consistent with the effect of solidification rate on the types of carbides that precipitate; a faster cooling rate favors the formation of M₂C carbide over M₆C.[37] Finally, the presence of Fe₃C carbide in region 2 of Figure 12 is another evidence to indicate that the primary matrix phase is an auto-tempered martensite.

F. Chemical Compositions of Carbides and Primary Matrix Phase

Average compositions of the two commonly observed primary carbides (e.g., MC and M₂C) in EBM S390, as determined by the STEM-EDX point analyzes, are given in Table V. It is evident that the MC carbide contains a much higher V content than that of M₂C. In contrast, the M₂C carbide takes up a fair amount of W, Fe, and Mo elements with the W being the highest. Table V.

Cobalt was also found in M₂C carbide, but not in MC carbide. This is consistent with the previous data on cobalt-containing high-speed steels.[22-23] Chromium does not partition to either MC or M₂C carbide;[24] therefore, the measured Cr content is close to that in the bulk sample, Table I.
For comparison purpose, the carbide compositions in the spray-formed ASP30 as measured by STEM-EDX are shown in Table V. The MC carbide seems to show a similar composition between the EBM S390 and spray-formed ASP30 high-speed steel. Compared to the bulk alloy composition for S390 steel containing 1.64C-10.12W-2.00Mo-4.97V-4.90Cr-7.86Co, the ASP30 steel contains a lower amount of W (6.4 wt pct) but higher in Mo (5.0 wt pct). This could explain the compositional difference in terms of W and Mo for the MC carbide. A similar trend can be found for the M 2C carbide. The primary difference in terms of the M 2C carbide composition between the EBM S390 and spray-formed ASP30 is that the Fe content in the former is high. This might be attributed to the rapid solidification during the EBM process. Fischmeister
et al. studied a wide range of high-speed steels and it was claimed that the Fe content in M₂C carbide increases strongly with the increasing cooling rate. Of course, this could be also attributed to the partial decomposition of metastable M₂C to M₆C carbide, Figure 12, that is known to contain a much higher Fe. By studying a range of high-speed steels (M₂, ASP23, ASP30, ASP60), it was found that the M₆C carbide can contain 40 to 50 at. pct of Fe.

The carbon content associated with the MC and M₂C carbides was firstly calculated with the assumption of their exact stoichiometry. The amount of carbon consumed by forming both carbides was then calculated with their volume fractions (Table IV). By knowing the carbon concentration in the bulk EBM S390 steel (1.48 wt pct in Table I), the carbon content in the auto-tempered martensite as the primary matrix phase was calculated from the difference. Table VI presents the compositions of the steel matrix as measured by STEM-EDX together with the calculated carbon. It can be seen that, because of the formation of primary carbides, the carbide-forming elements (e.g., V, W and Mo) was significantly lower in the auto-tempered martensite matrix compared to the bulk alloy composition in Table I. The remaining carbon concentration in the matrix of as-EBM S390 steel is similar to what has been measured in other high-speed steels. This indicates that the matrix microstructure in as-EBM S390 is as good as that made by conventional means (e.g., PM route), and an EBM post-processing heat treatment might promote the formation of secondary hardening carbides as there is still a fair amount of carbon remained in the matrix.

G. Mechanical Properties

The average transverse rupture strength of ~ 2500 MPa was obtained for as-EBM S390 steel. Table III; this value is higher than that of 2094 MPa for the fully heat-treated PM S390 steel processed by hot-isostatic-pressing. Fractography of the failed sample was examined by SEM to understand the nature of the failure. The low-magnification SEM image in Figure 13 shows that cracks were not originated from gas pores and the inset highlights that most of the fracture surfaces exhibited a cleavage fracture mode. This fracture observation applies to all as-EBM samples, i.e., no evidence indicating that the gas pores acted as the crack source. This probably explains why a desirable transverse rupture strength can be obtained in the as-EBM S390 steel.

HRC hardness of all as-EBM samples 1 to 4 are summarized in Table III. Regardless of parameter sets selected in the present work, all S390 steel samples had a similar hardness value of ~ 65 HRC. It is important to note that the hardness value found in these as-EBM samples are very similar to that in the fully heat-treated PM S390 steel; 65 to 66 HRC was reported in Reference 20 and 61.1 to 64.4 HRC in Reference 28. It is hence reasonable to conclude that the microstructure generated by the EBM process is very promising and the measured hardness value meets the high hardness requirement for the wear-resistant tool steel. The mechanical property in the present EBM S390 steel could be further enhanced with the help of post-processing heat treatment and this will be discussed in Section IV–C.

IV. OVERALL DISCUSSION

A. Microstructure Formation During EBM Non-equilibrium Solidification

The solidified microstructure and carbide precipitation behavior in EBM S390 high-speed steel have been fully characterized using a range of complementary techniques. The steel matrix consists of auto-tempered martensite with very little tetragonality left together with limited amount of retained austenite (5 pct in volume fraction). The discontinuous carbide network basically delineated the grain structure, Figure 4, and a combination of columnar and near-equiaxed grains was found in as-EBM condition. The eutectic carbides, predominantly M₂C and MC-type (Figures 9(c) and (d)), at the inter-dendritic region have a larger degree of refinement (typically a few tens to hundreds of nanometers). This can be evidenced by comparing with those primary carbides found in PM S390 fabricated by hot isostatic pressing and in heat-treated condition, Figure 8. The microstructure refinement is very likely to be due to the applied localized high energy input combined with the rapid cooling rate when the electron beam sweeps over the melted region. The high cooling rates (10⁵ to 10⁷ °C/s) during EBM process, as determined in previous EBM process modeling work, are much higher than that of 0.1 to 5 °C/s in ingot casting. Microstructural refinement observed in as-EBM S390 arises primarily from the limited time for dendrite growth.

The presence of both near-equiaxed and columnar grains, Figure 4, can be attributed to the EBM solidification and line-by-line remelting processes. By examining the top layers of as-solidified microstructure, the periodically appeared distinct grain structures can be seen in Figure 14. Two micro-melt pool boundaries are indicated in Figure 14(a). The melt pool depth is approx. three times the powder layer thickness of 50 µm. Figures 14(b) and (d) show the magnified view of the locations adjacent to the melt pool boundary. Figure 14(c) confirms the columnar dendritic nature of the grains located close to the melt pool boundary. These columnar dendritic structured grains were exclusively revealed at the locations adjacent to the melt pool boundary. This term ‘cellular and dendritic’ is often used to represent different quantities in the AM field (e.g., References 18, 40 and 41), with the result that much confusion develops. In the welding field, the difference between the cellular and dendritic growth is that the cellular solidification is significantly controlled by the heat flow direction, while dendritic growth is also affected...
It should be noted that the distinction between the two is not consistent in the literature. The alternative approach is to define dendrites as having branched structures, whereas cells have no branches, without regard to their growth direction. However, without any compositional variations in the substructure of grains, it is almost impossible to reveal the dendritic nature of the solidification microstructure. In other words, without noticeable branching, the columnar dendritic structure would be very similar to the cellular solidification structure. Fortunately, the formation of eutectic carbides at inter-dendritic regions in the present EBM S390 steel helped to reveal the dendritic nature of solidification microstructure.

In welding, the fusion zone grain structure is determined by the base metal grain structure and the welding condition. Because the weld pool is formed by melting regions of the base metal, the base metal that is in contact with the liquid acts as an ideal substrate on which growth of the solid phase occurs. This means that the initial solidification would occur epitaxially at the partially melted grains in the base metal. For the present EBM S390 steel, the formation of the micro-melt pool is analogous to that of the welding. The solidification process associated with the next adjacent scan line occurs epitaxial at the partially melted grains. This results in a columnar dendritic growth behavior at the region close to the melt pool boundary. This applies to both the side and bottom of the micro-melt pool, Figures 14(b) and (d). The average growth direction of the dendrites is approximately perpendicular to the solid/liquid interface and parallel to the maximum thermal gradient, namely the heat flow direction. At the center of the micro-melt pool that has not been remelted, the initial solidified microstructure would be remained and no dendritic-to-cellular transition in the solid state should be expected.

It then becomes interesting to interrogate the formation mechanism of near-equiaxed grains in the center of micro-melt pool. By the grown crystallography, it should be noted that the distinction between the two is not consistent in the literature. The alternative approach is to define dendrites as having branched structures, whereas cells have no branches, without regard to their growth direction. However, without any compositional variations in the substructure of grains, it is almost impossible to reveal the dendritic nature of the solidification microstructure. In other words, without noticeable branching, the columnar dendritic structure would be very similar to the cellular solidification structure. Fortunately, the formation of eutectic carbides at inter-dendritic regions in the present EBM S390 steel helped to reveal the dendritic nature of solidification microstructure.

Table V. Average STEM + EDX Analyzes of Chemical Compositions of $M_2C$ and MC-Type Carbides

| Alloy Type       | Carbide Type | Chemical Composition (Weight Percent, $\sum_{M} = 100$) |
|------------------|--------------|--------------------------------------------------------|
| EBM S390         | MC           | 35.5 W 6.2 Mo 49.2 V 5.8 Cr 3.3 Fe 0 Co               |
|                  | $M_2C$       | 49.5 W 9.5 Mo 3.2 V 5.4 Cr 28.3 Fe 4.1 Co             |
| Spray-Formed ASP30 | MC           | 26.7 W 18.2 Mo 45.9 V 5.2 Cr 4.0 Fe 0 Co              |
|                  | $M_2C$       | 36.4 W 29.5 Mo 13.7 V 6.9 Cr 10.4 Fe 3.1 Co           |

Table VI. Chemical Composition of Martensite Matrix in EBM S390 Steel

| Alloy Type | Chemical Composition of Matrix (Weight Percent) | Bulk Total (Weight Percent) |
|------------|-------------------------------------------------|-----------------------------|
|            | C* W Mo V Cr Co Fe                              | C**                         |
| EBM S390   | 0.6 2.5 0.6 2.0 4.6 9.7 bal. 1.48               |                             |
| Atomized T15 powder | 0.6 7.3 1.3 4.4 5.9 bal. 1.52               |                             |
| Conventional M2 | 0.6 3.8 3.5 1.0 4.3 bal. 0.94               |                             |
| Spray-formed ASP30 | 0.3 3.5 2.8 0.1 3.9 8.7 bal. 1.28               |                             |

*Calculated data from the bulk alloy analysis and theoretical carbon content corresponding to the measured volume fraction of carbides as well as the approximate chemical formulas for MC and $M_2C$-type carbides. The retained austenite was not considered due to its low volume fraction as the matrix phase.

**The carbon content reflects the total carbon in the bulk EBM sample as measured by ICP-AES, Table I.

Fig. 13—SEM fractography examination on as-EBM sample 2 showing that gas pores did not act as the crack source for the transverse rupture strength test.
Three micro-melt pools with two melt pool boundaries were illustrated in Figure 15(a). The yellow box highlights the region where the EBSD scan was made with respect to the micro-melt pool. This region contains columnar grains in the left and near-equiaxed grains in the right, separated by a melt pool boundary, Figures 15(b) and (c). EBSD orientation map and corresponding pole figures for the columnar grains and near-equiaxed grains are shown in Figures 15(d) and (e). The maximum pole figure intensity was found close to 5 and there is a rotational symmetry between the two distinct grain regions. For the columnar grains, the maximum pole figure intensity was found to be at (111) plane normal, while (111) plane normal for the near-equiaxed grains. The angle between the two crystallographic planes was calculated as 70.5 deg. Considering such a rotational symmetry, it is very likely that the near-equiaxed grains were developed by columnar dendritic growth and their average growth direction followed the maximum thermal gradient, namely parallel to the electron beam travel direction (X direction in Figure 15(a)).

Based on the EBSD-based crystallographic orientation analysis, it seems that the grains formed at the center of micro-melt pool were most likely formed by dendritic solidification. The reason why they have a totally different grain morphology to those close to the melt pool boundary is probably due to the direction change of the maximum thermal gradient at the center of the micro-melt pool. In other words, to reveal the micro-melt pool morphology in a 2D SEM viewing plane (Y–Z plane), the sample was longitudinal sectioned in such a way that the plane is actually perpendicular to the traveling direction of the melt lines, X direction as indicated in Figures 14(a) and 15(a). Unfortunately, this naturally creates a sectioning effect when observing any dendritic structure with the average growth direction aligned with X direction.

The schematic diagrams shown in Figure 16 summarize the solidification sequence during the S390 steel EBM process. The initial solidification occurs by the formation of columnar dendritic primary austenite phase, probably followed by the formation of proeutectic MC carbides around dendrites, the left side of Figure 16(c). Columnar grains were found close to the melt pool boundary, while near-equiaxed ones at the center of the micro-melt pool, Figure 16(a), controlled by the maximum thermal gradient direction. With the decreasing temperature, the remaining liquid phase that is more enriched with solute atoms is solidified by forming eutectic carbides (MC and M2C carbides), Figure 16(c). Note MC can occur both as eutectic and as primary crystals from the melt. When the next adjacent melt line arrives, part of the already solidified dendritic structure in the previous melt pool will be melted and then re-solidified to form columnar dendritic
structured grains close to the new melt pool boundary, Figure 16(b). When the part fabrication is finished, the high-temperature austenite phase transforms to martensite phase and its auto-tempering occurs during the final cooling.

The above-envisaged solidification path is similar to that found in gas atomized high-speed steel T15 powders and spray-formed high-speed steel. The presence of a fair amount of metastable M$_2$C carbide in the as-EBM microstructure can be attributed to the high-carbon concentration in S390 steel (1.64 wt pct in the original powder and 1.48 wt pct in the bulk sample, Table I). Sare and Honeycombe found that a high-carbon content in splat-quenched M1 tool steel promotes the M$_2$C carbide formation by broadening the composition and temperature range. Since no solidification/hot cracking was found in the as-EBM S390 steel (see SEM fractography in Figure 13), the presence of liquid film in as-solidified sample is very unlikely; hence, the left part of Figure 16(b) shows that columnar dendrites adjoined each other. The effect of solute re-distribution during solidification on the material susceptibility to cracking can be assessed by performing thermodynamic calculation using Scheil–Gulliver solidification model, as already performed elsewhere.

**B. M$_2$C and MC Carbide Precipitation and Elemental Migration**

The relative proportion of M$_2$C carbide over the total carbide volume fraction depends on the area energy input; sample 4 with the smallest $E_A$ had the lowest value of 24.8 pct, Table IV. When the quantitative carbide measurement was performed by the use of threshold grayscale-to-binary method, it was found that columnar grains often had a much higher proportion of MC carbides, whereas the brighter white M$_2$C carbides were predominantly present in the near-equiaxed grains. The schematic diagram in Figure 16(c) depicts the phase constitution difference between two distinct regions. Recall that the area fraction of columnar grains in sample 4 (48.7 pct, Figure 4(d)) is higher than that of sample 3 (20.3 pct, Figure 4(c)). This means that the smallest relative proportion of M$_2$C carbide in sample 4 could be also attributed to the highest area fraction of the columnar grains. As shown in Figure 16(b), the line-by-line remelting could induce a heat flow from the micro-melt pool to the already solidified metal. This may cause a solid-state carbide phase transformation from MC to M$_2$C that probably involves MC $\rightarrow$ M$_2$C $+$ $\gamma$-Fe. If this hypothesis is temporarily accepted, a further theoretical calculation can be made to check the consequent effect of this unusual reaction.

From Table IV, it is known that the volume fraction of MC carbide decreased from 17.9 pct (sample 4, $E_A = 3.1$) to 13.6 pct (sample 3, $E_A = 4.7$) as a result of reduced area energy input $E_A$. In the meantime, the M$_2$C carbide increased from 5.9 pct to 8.2 pct. This indicates that 4.3 pct of MC transformed to 2.3 pct of M$_2$C by involving 2.0 pct of $\gamma$-Fe. Furthermore, by knowing the chemical compositions for both the MC and M$_2$C (Table V), together with their reported density...
\( \rho \) (in g/cm\(^3\)),\(^{24}\) the elemental migration for each major element (i.e., V, W and Fe) can be derived, according to:

\[
\Delta \text{Element} = \text{wt pct}(M_{MC}) \times \Delta V_{MC} \times \rho_{MC} - \text{wt pct}(M_{M2C}) \times \Delta V_{M2C} \times \rho_{M2C}
\]

where wt pct \( (M_{MC}) \) is the weight percent of certain metallic element in MC carbide, \( \Delta V_{MC} \) is the volume fraction difference of the MC carbide between samples 3 and 4, and \( \rho_{MC} \) is the density of the MC carbide. For the present calculation, the carbide density of \( \rho_{MC} \approx 7.28 \text{ g/cm}^3 \) and \( \rho_{M2C} \approx 8.81 \text{ g/cm}^3 \) were used. Figure 16(d) provides a schematic diagram to explain the elemental migration associated with the solid-state phase transformation (MC \&rightarrow; M\(_2\)C + \( \gamma\)-Fe) and the number of atoms reflect the amount of elemental migration.

Based on Eq. [1] the following elemental migration magnitude (in wt pct) was obtained for vanadium, tungsten and iron elements, respectively; \( \Delta V = +14.7 \text{ pct} \), \( \Delta W = +0.04 \text{ pct} \), and \( \Delta Fe = -4.7 \text{ pct} \). A positive sign indicates that the element migrates from MC carbide to the matrix, while a negative sign suggests the migration to the M\(_2\)C carbide. Thus, the following aspects can be summarized through the hypothetical reaction of MC \&rightarrow; M\(_2\)C + \( \gamma\)-Fe: (i) a significant content of vanadium in MC carbide migrates to the matrix; (ii) very limited tungsten migration occurs; (iii) a certain content of iron migrates to the M\(_2\)C carbide. Considering that very limited iron was detected in the MC carbide (Table V), it is most likely that the iron migration involves the \( \gamma\)-Fe matrix. Importantly to note, the vanadium migration to the matrix is consistent with the STEM-EDX results shown in Table VI, where the resulting martensite matrix in the final EBM S390 steel contains the highest vanadium content compared to the conventional means. The high vanadium content found in the auto-tempered martensite matrix could be responsible for the very promising mechanical property in the EBM S390 steel.

\[\text{C. Microstructure and Mechanical Property}\]

The very fine microstructure in as-EBM S390 steel removes the need to apply the hot forging to break up those blocky primary carbide networks prior to performing the full cycle of heat treatments—one of the common problems encountered in the manufacture of high-speed steel, particularly near the center of large cast ingots.\(^{19,44,45}\) For the present PM S390 high-speed tool steel, the highest HRC hardness value obtainable for the hot-isostatic-pressed (HIP) + austenitized + tempered condition as reported in the literature so far is between 62 and 68 HRC.\(^{20,28,46}\) By contrast, for the EBM S390 steel, even without the post-processing heat treatment, the as-EBM microstructure already comes with the good
hardness value of 65 HRC (Table III). The average transverse rupture strength for the as-EBM S390 steel achieves a high value of ~2500 MPa, that is higher than that offered by the conventional means.[39] The high hardness value and bending strength in the as-EBM S390 steel are likely to be a combined effect of refined primary carbide precipitation and matrix strengthening. Figure 17 shows the presence of vanadium cluster in the auto-tempered martensite matrix, suggesting that these nanometer length-scale V-rich carbides could act as secondary carbide strengthening mechanism.

To fully appreciate the great potential in the use of EBM to fabricate high-speed steels, EBM processed samples 3 and 4 were subjected to the full post-processing heat treatment cycles (first austenitization: 1170 °C/4 min/oil quenching, followed by triple tempering: 560 °C/2 + 2 + 2 hours). The resulting hardness for the heat-treated EBM S390 steel reached to HRC value of 73.1 ± 0.2 and 72.6 ± 0.2 for sample 3 and sample 4, respectively. The corresponding transverse rupture strength was further increased to 2784 ± 51 MPa and 3012 ± 34 MPa. Prior to the heat treatment, the as-EBM samples were subjected to a HIP treatment (150 MPa, 1150°C, 2 hours) so that the densification of the material can be increased. The extremely high HRC hardness value obtained in the post-processed EBM S390 steel appears to be the first-ever. Therefore, this exploratory work demonstrates a new cost-effective process route to fabricate a wide range of high-speed tool steels whose mechanical properties are primarily relied on the size and distribution of carbide precipitation. In the follow-up paper, the microstructural evolution and material strengthening mechanisms in post-processed EBM S390 steel will be provided.

V. CONCLUSIONS

The solidification microstructure, primary carbide precipitation behavior (both the composition and volume fraction), and the carbide decomposition in an Fe-1.64C-10.12W-2.00Mo-4.97V-4.90Cr-7.86Co (S390 steel) fabricated by EBM have been critically assessed. The following conclusions can be reached:

1. The crack-free microstructure of the as-EBM S390 is characterized by the discontinuous network of primary carbides that are dispersed in the auto-tempered martensite matrix. Limited amount of retained austenite was found in the matrix.

2. The discontinuous network of eutectic carbides is characterized by both the M₂C and MC carbides and their relative proportions in as-EBM microstructure depends on the area energy input. The M₂C carbides takes up tungsten, iron, molybdenum, and cobalt in the order of significance.
while MC carbide contains primarily vanadium followed by tungsten.

3. The solidification sequence starts with the formation of primary austenite dendrites followed by proeutectic MC carbides around the dendrites. Further solidification continues with the formation of eutectic MC and M₂C and austenite phase at inter-dendritic areas. Upon the final cooling, the martensite formed and auto-tempered.

4. The transverse rupture strength as well as the Rockwell HRC hardness in as-EBM condition is as good as that achieved in a post fully heat-treated HIP S390 steel. A typical cleavage fracture mode was identified based on SEM fractography and cracks does not seem to be originated from gas pores.

5. With the help of EBM post-processing heat treatments, the hardness and transverse rupture strength were further enhanced to 73.1 ± 0.2 HRC and 3012 ± 34 MPa. The detailed interpretations of the EBM S390 steel in heat-treated condition will be given in the follow-up paper.

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