Effects of laser energy density on carbide dissolution, element distribution and microstructure evolution of AISI P20 steel after laser surface quenching

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Abstract
Laser surface quenching (LSQ) was performed on AISI P20 mould and hot-working die steel with an objective to improve surface characteristics. The steel was treated under three different process parameter conditions. The microstructure, element distribution, and residual stresses were investigated through SEM, EDS, and XRD analyses. The effect of laser energy density on carbide dissolution/ablation and microstructure evolution was thoroughly investigated. The dissolution/ablation of carbides significantly affected the formation of martensite and retained austenite, and the distribution of elements and phase in the microstructure. The results of the study and analyses of treated surface revealed that the LSQ treatment significantly improved the microstructure and eliminated the pores or other defects. Furthermore, the degree of carbide dissolution/ablation was closely related to the laser energy density. Comparing to Cr7C3, Cr3C2 was more difficult to dissolve at lower laser energy density. Thus, those incompletely dissolved Cr3C2 would hinder the growth of austenite and reduce the carbon content in austenite and lead to the formation of low-carbon martensite. The highest laser energy density (150 J/mm²) was able to produce finer microstructure and significantly reduced the inhomogeneity in distribution of Cr between the poor and the rich Cr areas.

Keywords Laser surface treatment · Laser quenching · Carbide dissolution/ablation · Element distribution · AISI P20 · Microstructure

1 Introduction
As a typical mould and hot-working die steel AISI P20 (3Cr2Mo) is widely used in the plastic mold and the low melting point metal forming die-casting mold due to the excellent combination of properties such as toughness, hardness, and comprehensive performances [1, 2]. The AISI P20 is commercially supplied in pre-hardened condition to 30–36 HRC and can be directly employed in molding processing. The service life of molds can reach 500,000 up to times. However, the molds usually fail by surface degradation [3]. When AISI P20 is employed in plastic molds, it encounters a temperature in the range of 200–250 °C. The molten plastic flows into the mould cavity at a high velocity under high pressure which results in serious friction and abrasion on the surface of the cavity [4]. In addition, the strong corrosive gases and vapors (HCl, HF, etc.) produced by the melting and decomposition of plastics can also corrode the mould surface. During moulding and metal forming, AISI P20 steels are also exposed to cycles of repetitive cooling and heating (at temperature of the order of 400–700 °C) along with severe dynamic/static cyclic load. Such conditions lead a set of severe adverse conditions involving corrosion, erosion, thermal fatigue, and abrasive/adhesive wear, etc., and greatly reduce the service life of the
molds. [1, 5] A surface treatment of AISI P20 to strengthen and harden up to 57–60 HRC can improve the service life by 1 million times [6]. The adverse service conditions require careful selection of effective treatment process which can effectively control of thickness, hardness, interface bonding, and ability to endure the environment. The ideal process not only keeps the original chemical composition of the surface, which saves the production cost caused by the post-treatment, but also brings great economic benefits.

The laser surface quenching (LSQ) and laser melting are among the best of the available alternatives, which can well overcome the shortcomings of conventional surface strengthening processes. In comparison to laser melting, the LSQ can significantly improve the microstructure and surface performances without melting or damaging substrate. Telasang et al. [5, 7] studied the effect of LSQ and laser melting on the microstructure and surface properties of AISI H13 and reported that the LSQ resulted in superior corrosion resistance and mechanical properties with the yield strength improved to as high as 1460 MPa. The authors attributed the overall improvement to the better interface characteristics and the formation of low-carbon martensite in case of LSQ. However, the precipitation of brittle carbide at the grain boundary and presence of dendritic structure in the laser-melted samples caused poor corrosion resistance and lack of toughness. The yield strength was also found to be 1310 MPa. Therefore, the better performance in stringent environment has attracted huge interest in the investigations on wear and corrosion resistance of different steels by LSQ and made a great deal of scientific research achievements [8–10].

The surface properties depend on the microstructure in the quenched zone of the LSQ-treated surface, which is characterized by the phase composition, the dissolution and diffusion of the initial carbides, the uniformity of element distribution, and state of residual stresses. Maharjan et al. [11] performed LSQ on AISI 4150 to enhance the corrosion resistance and reported that formation of uniform protective oxide layer containing Cr in quenched surface was a key factor which resulted in enhancing the corrosion resistance of treated samples by about 3 times. Zhang et al. [12] combined element distribution and microstructure morphology demonstrated that laser heat treatment can effectively dissolve Ti and N elements in the alloy 800-H phase. The dissolution significantly reduced the amount of TiN as well as probability of crack initiation. Chen et al. [13] reported that the surface performances of 40-Cr steel could be improved by LSQ treatment. The authors demonstrated that the resistance to impact abrasive wear of treated samples was improved by 2 times. The results of EDS analysis showed that the SiO_{2} and Fe_{2}O_{3} particles existed before the scratch, and the larger particles can damage the metal under the impact.

The constituent phases comprising of carbides are the main strengthening phases in the microstructure. The ablation, dissolution, and diffusion of carbide during LSQ seriously affect the properties of microstructure in the quenched zone. The C acts as austenite stabilizer, expand the austenitic region significantly, and its effect is about 30 times stronger than Ni [14, 15]. The distribution of C leads to uniformity and compactness of microstructure (mainly α-Fe), as well as the uniform transformation of austenite into martensite. The Cr can significantly reduce the intergranular corrosion and improve the resistance to Cl^{-} corrosion [16, 17]. The alloying elements and diverse phase changes encountered during surface treatments produce complex set effects which greatly alter the surface properties. Therefore, a detailed study on the ablation/dissolution of carbides and the diffusion and distribution of elements in the LSQ process is important and useful.

Current work reports a comprehensive analysis on the microstructure characterization of AISI P20 steel surface modified by the LSQ process. Extensive analyses on phase composition, element distribution, residual stress, and dissolution/ablation of carbide have been investigated in detail. The effects of the carbide dissolution/ablation on the formation of martensite, retained austenite, and the element distribution were thoroughly investigated. The intrinsic relationships among carbide dissolution/ablation, element distribution, microstructure formation, and residual stress were clarified. This work is of great significance for the improvement of LSQ microstructure and the reasonable match between the microstructure and properties.

### 2 Experimental procedure

#### 2.1 Materials

The elemental concentration of AISI P20 steel is given in Table 1. The steel received has been hardened as follows: (a) heating: the AISI P20 steel was heated to 600 °C and kept at this temperature for 2 h; (b) hardening: heating the steel to 900 °C for 1.5 h followed by quenching in oil; and (c) tempering: reheating the steel to 600 °C, and keeping it at this temperature for 2 h followed by air cooling.

| Table 1 Elemental concentration (wt.%) in AISI P20 steel |
|--------------------------------------------|
| Elements | C   | Cr  | Mo  | Si  | Mn  | Ni  | S   | P   | Fe  |
| Composition | 0.39 | 1.95 | 0.35 | 0.45 | 0.6 | 0.1 | 0.001 | 0.012 | Bal |
**2.2 Laser treatment system and laser surface quenching**

The laser treatment system utilized in the experiments is presented in Fig. 1 [18]. The IPG YLS-4000 fiber laser produced a Gaussian beam with a wavelength of 1070 nm and a maximum power of 4 kW. The rectangular flat-top beam of $25 \times 4$ mm$^2$ was shaped by RayTools AK390 laser quenching head from the Gaussian beam. Coaxial protective gas (99.999% Ar) with 8 L/min was utilized during LSQ process. A digital control system was employed to adjust the processing parameters and to control KUKA KR30HA robotic arm. The workpieces with polished working surface were fastened on the platform and processed with different laser parameters respectively by controlling robotic arm.

The three sets of process parameters of LSQ process were given in Table 2. Laser power and scanning speed were variation parameters. Laser energy density comprehensively represented the interaction time and laser power density. The LSQ modified regions in three samples of AISI P20 were measured as $200 \times 25$ mm$^2$. After the first LSQ process is completed, the second process would not be carried out until the temperature of the hardened zone is reduced to room temperature. Each hardened zone was kept 50 mm apart.

**2.3 Characterization**

Each sample with the size of $20 \times 10 \times 10$ mm$^3$ was cut from the center of each LSQ region (marking in Fig. 1) and was polished by standard metallographic procedure. The microstructure of the polished samples was analyzed using Quanta 400FEG SEM equipped with EDS. The phase identification of the LSQ samples was carried out by using Bruker D8-ADVANCE XRD with Cr radiation, operating at 40 kV.

| Sample | Laser energy density (J/mm$^2$) | Laser power (kW) | Laser scanning speed (mm/s) | The focusing beam (mm$^2$) |
|--------|-------------------------------|-----------------|-----------------------------|---------------------------|
| #1     | 150                           | 1.8             | 3                           | $25 \times 4$             |
| #2     | 142                           | 1.7             | 3                           | $25 \times 4$             |
| #3     | 113                           | 1.8             | 4                           | $25 \times 4$             |
and 40 mA. The residual stress measurement was performed by PROTO LXRD SYSTEM XRD with parallel beam and Cr radiation operated at 40 kV and 40 mA.

3 Results and discussion

3.1 Microstructure and phase composition

The LSQ modified surface of steel is found to possess three distinct zones in quenching layers: quenched zone, intermediate zone, and thermally affected zone [6, 18]. A detailed analysis of strengthening phases, element, and phase distribution in quenched zone shall be important for the process optimization and performance improvement.

The microstructure of the untreated samples and all the three LSQ processed samples is shown in Fig. 2. The microstructure of the untreated steel was found to be comprised of tempered sorbite (Fig. 2a). The microstructure also revealed the presence of small number of pores distributed on the ferrite, which would seriously affect the service life of the AISI P20 steel mold. It is clear from Fig. 2b–d that the pores were eliminated after LSQ treatment and the microstructure of treated samples was significantly improved. This could be attributed to the formation of new austenite grains because of the lattice recombination and elements diffusion during the heating process. The ferrite phase with pores possessed higher phase-changing energy, which promoted nucleation and growth of new austenite. This illustrated that the LSQ process was beneficial in eliminating pores. Further, it could be shown that the microstructure of quenched zone was mainly consisted of lath martensite (LM), plate martensite (PM), retained austenite (RA), and a small number of carbide particles. Figure 2b, c reveals that the more PM with high-carbon content distributed in the primary austenite grains in sample #1 than in sample #2, and the blocky RA were disappeared. The primary austenite grains formed during austenitizing in sample #1 were faintly visible and the grains were larger as well. The cooling process promoted rapid nucleation of PM in the primary austenite. Under the influence of cooling rate and alloying elements on Ms line, some untransformed RA still existed between the PM in the primary austenite. For sample #1, the laser power and laser energy density were higher, which resulted in the coarsening of primary austenite grains [17]. Furthermore, the high austenitizing temperature and heating rate caused more alloying elements (such as C) to diffuse into austenite, which helped in the high-carbon PM formation during cooling. The high-carbon PM formation has also been confirmed from the literature [18]. The temperature and heat input were relatively lower in sample #2. In sample #2, carbides were not fully dissolved during the austenitizing and suppressed
the growth of austenitic grains, resulting in relatively finer grains. Moreover, the diffusion of alloying elements retarded dissolution of carbon in austenite, which led to the more formation of low-carbon martensite.

A comparison of sample #1, #2, and #3 showed that the sample #3 was comprised of coarse microstructure with poor uniformity, which contained more LM and some blocky RA distributed in a concentrated and local way (as shown in Fig. 2d). The high scanning speed in case of sample #3 also resulted in a lower laser heating temperature [19], which led to the lower carbon content in primary austenite and the formation of more LM. The laser power and scanning speed of sample #3 were higher (in comparison to sample #2), but the laser energy density was relatively lower, which reduced the laser-to-matrix interaction time and increased the cooling rate. Such conditions caused the non-uniform of temperature distribution and element diffusion in the localized regions of the quenched zone. Additionally, low carbon in the austenite and high cooling rate is conducive to the formation of low-carbon martensite during the phase change, which was also confirmed by XRD analysis as shown in Fig. 3.

In addition to detail microstructural investigation, XRD analyses were also performed on the surface of quenched zone, as shown in Fig. 3 and Table 3. On comparing the XRD peaks and relevant data between the untreated and LSQ samples, it was evident that the microstructure of the untreated samples contained ferrite, Cr-rich carbides M7C3 and M3C2. The XRD results of the LSQ samples were consistent with the phase composition observed by SEM, including martensite, retained austenite, and a small number of carbides. However, no carbides could be detected in sample #1 and sample #3, which was probably because almost all carbides were dissolved during heating due to the high laser energy density. In comparison to the untreated samples, the XRD peaks of LSQ samples were significantly broadening. The half-height width of the martensite (α-Fe) diffraction peak of sample #1 (0.586) was larger than the ferrite (α-Fe) diffraction peak of untreated sample (0.167), and larger than the martensite (α-Fe) diffraction peak of sample #2 (0.540) and sample #3 (0.531). This indicated that the LSQ process refined microstructure to a larger extent (as shown in Fig. 3). Additionally, broadening of peak was also related to the lattice micro-strain variables generated from the solid-state phase transformation during the heating/cooling process. The peak broadening of the conventional heat treatment is similar to that of laser heat treatment [5]. Therefore, it can be concluded that the broadening of the diffraction peaks of martensite (α-Fe) in this process was mainly caused by the microstructure refinement during the phase changing process. It should be noted that, after LSQ, the diffraction peaks of martensite (α-Fe) shifted to the left, in which sample #1 possessed the largest offset (2θ = 44.50°), and the crystal plane spacing d was the largest (as shown in Table 3, d = 2.0343). This may attribute to the fact that the sample #1 was irradiated under high laser energy density

Table 3 Analysis data of α-Fe peaks

|         | 2θ (°) | d (nm) | Height | Area | FWHM (2θ) |
|---------|--------|--------|--------|------|-----------|
| Untreated steel | 44.73  | 2.0243 | 1180   | 11,927 | 0.167     |
| #1      | 44.50  | 2.0343 | 565    | 19,993 | 0.586     |
| #2      | 44.58  | 2.0310 | 434    | 14,159 | 0.540     |
| #3      | 44.67  | 2.0268 | 290    | 9289  | 0.531     |

Fig. 3 XRD patterns of untreated AISI P20 steel and different LSQ samples
(150 J/mm²), more solute atoms (C, Cr, Mo, etc.) dissolved in austenite during austenitization, but these atoms could not precipitate during the rapid cooling, which would enhance the formation of high-carbon martensite. This was also confirmed in the discussion given in Sect. 3.1 (Fig. 2c). Furthermore, as shown in Table 3, the diffraction peak area (19,993) and peak value of martensite (α-Fe) in sample #1 were the largest, indicating that sample #1 possessed higher crystallinity and contained more high-carbon martensite. The higher the crystallinity, the more regular the arrangement of atoms in the crystal, the greater deformation resistance was obtained [20, 21]. For the sample #3, the diffraction peak area (9289) and peak value of martensite (α-Fe) were the smallest, illustrating that the crystallinity of the obtained microstructure was lower, and form less martensite, so that the microstructure uniformity of sample #3 was poor as described in Sect. 3.1 (Fig. 2d).

3.2 Residual stress

The LSQ process is capable of changing the state of internal stresses and strengthening surface. The stress distribution also significantly affects the corrosion and wear resistance. The distribution of residual stress in the longitudinal and transverse directions in the strengthened layer of the LSQ-treated surface is shown in Fig. 4. The initial stress on the untreated surface was compressive, and the average value of the measured stress was −48.7 MPa. The compressive stress in the as received material would have developed during surface machining. The residual stress was measured on the strengthened layer of the LSQ-treated surface and found to be compressive in nature and significantly increased (average value: -225.1 MPa). It may be noted that the flat-topped laser employed in this work has a uniform energy distribution across the beam cross-section. The surface-treated by such a beam generated a residual stress profile which was more or less uniformly distributed in the longitudinal and transverse directions. A large number of carbon atoms diffused into austenite during heating and caused distortion in the face-centered cubic (FCC) structure; on subsequent quenching, the martensitic transformation changed the lattice from FCC to body-centered cubic (BCC), which developed high volumetric stress in the already distorted lattice and produced significant compressive stress. An increase in the laser energy density increased the diffusion of C atoms (associated lattice distortion too) and in turn increased the compressive residual stress. Furthermore, in addition to lattice distortion, the increase of laser energy density also promoted the formation of martensite, which also added to the residual stresses. This was also confirmed by the XRD analysis. The compressive residual stress on the surface effectively reduced the fatigue cracks and improved the fatigue and corrosion resistance of the treated surface [22–24].

3.3 Element distribution

The LSQ process usually completes in a very short time; consequently, the alloying elements usually do not get enough time to diffuse evenly, leading to fluctuations in concentration in different areas. This raises concern on the homogeneity of microstructure and surface performances. Figure 5 shows the results of EDS of the untreated and LSQ treatment AISI P20 steel under different laser energy density. The two key elements of C and Cr were analyzed emphatically here, and the EDS results are listed in Table 4.
The distribution of C is indicative of the microstructural homogeneity and distinguishes the type of martensite present on sight. Similarly, the distribution of Cr is indicative of microstructural regions of poor or rich chromium which accordingly defines the evaluation of corrosion resistance property (especially to the corrosion of Cl−).

It is evident from Fig. 5a that the concentrations of Cr and C largely varied and were present on the ferrite boundary in the untreated steel. The microstructure of untreated steel consisted of tempered sorbite and the C and Cr mainly existed as carbides. The laser heating during LSQ caused diffusion/dissolution/redistribution of elements (C and Cr).

Fig. 5 SEM and EDS map scanning data of untreated AISI P20 steel and LSQ samples: a untreated steel, b #1, c #2, d #3
and the fluctuations in the concentration were evened out (Fig. 5b–d). A comparison of sample #2 with sample #1 indicated that the distribution of C in sample #1 was more uniform (Fig. 5b). Almost all the carbides were dissolved under the effect of higher laser energy density, the higher temperature, and the lower scanning speed. These factors also increased the diffusion rate of C, prolonged the effective diffusion time [25], and resulted in uniform distribution of C. The uniform distribution of C enhanced the uniform and dense microstructure, as shown in Fig. 2b. Combining with the point analysis revealed that the carbon content of RA in sample #1 was the highest in all the samples, which was 1.7% (point A). This indicated that more C atoms diffused into austenite and retained during the austenitizing. Meanwhile, it is also evident that more PM with high-carbon content were formed in sample #1. The lower temperature and lower heat input resulted in some carbides to remain undissolved in sample #2, which caused very high nonuniformity of C distribution than sample #1. Further, the number of C atoms in solid solution was lower which was found to be only 1.1% (point B). The distribution of C in sample #3 was found to be similar to that of untreated steel, with large fluctuation of carbon concentration and poor distribution (refer to Fig. 5d). Most of C atoms were found on the body and boundary of RA surrounded by LM. This was attributed to the fact that high scanning speed caused insufficient diffusion of C and significantly affected the uniformity in the microstructure. The distribution of C was found to be more uniform after LSQ treatment which was desirable in reducing the local corrosion caused by the presence of a large number of poor chromium regions [11, 16]. Therefore, the corrosion resistance of mold steel can be significantly improved by LSQ treatment.

### 3.4 Carbide ablation and microstructure evolution

#### 3.4.1 Carbide ablation

The dissolution/ablation of carbide imparts significant effects on the diffusion and distribution of C. When the laser power is low, there may be some incompletely dissolved carbides remaining in the quenched zone. Figure 6 shows the high magnification SEM micrographs of two carbides with different shapes which were distributed in the quenched zone in sample #2. The EDS point analysis data (listed in Table 5) and the results of XRD could confirm that the block-shaped carbides (as shown in Fig. 6a) were Cr$_7$C$_3$. The rod-shaped or elliptic carbides were identified as Cr$_3$C$_2$ (as shown in Fig. 6b). It can be evident that the irregular ablation layers and the particles which were broken and exfoliated due to ablation exist at the carbide boundary (Fig. 6a). Cr$_7$C$_3$ has better thermal stability as compared to Cr$_3$C$_2$ [26, 27]. Consequently, the Cr$_3$C$_2$ ablated relatively lower and the ablation layer (broken particles characterizing ablation) at the boundary of Cr$_3$C$_2$ was not obvious (as shown in Fig. 6b). Moreover, Cr$_7$C$_3$ underwent secondary reaction with C at high temperature and formed Cr$_3$C$_2$ (as given by Formula 1) [28–30].

$$\frac{3}{7}Cr_7C_3 + \frac{5}{7}C \rightarrow Cr_3C_2 \tag{1}$$

High magnification SEM at the local regions could be effectively utilized in the analysis of undissolved carbides in the quenched zone of each sample (refer to Fig. 7). For sample #2 which was treated with lower laser power, undissolved

### Table 4 EDS analysis of P1–P3 (wt. %)

|       | Cr  | C   | Mn  | Si  | Fe      |
|-------|-----|-----|-----|-----|---------|
| Point A | 1.5%| 1.7%| 0.5%| 0.2%| 96.1%   |
| Point B | 1.2%| 1.1%| 0.5%| 0.4%| 96.8%   |
| Point C | 1.3%| 1.3%| 0.4%| 0.3%| 96.7%   |

### Table 5 EDS analysis of C1–C2 (wt. %)

|        | Cr  | Mn  | Si  | Fe  |
|--------|-----|-----|-----|-----|
| C1     | 86.5%| 0.4%| 0.5%| 3.0%|
| C2     | 83.4%| 0.3%| 0.3%| 2.8%|

Fig. 6 Higher amplification of LSQ layer and carbides

![Fig. 6 Higher amplification of LSQ layer and carbides](image)
carbides in the quenched zone were observed, most of which were rod-shaped Cr$_3$C$_2$ with high thermal stability (as shown in Fig. 7b). In comparison to Cr$_7$C$_3$ blocks, the Cr$_3$C$_2$ mostly appeared with relatively clear boundaries with slight dissolution/ablation phenomenon (as analyzed and discussed in Fig. 6 of the preceding section). The sample #1 was treated with the highest laser energy density (as shown in Fig. 7a), and it did not possess any trace of undissolved carbides in the quenched zone. It can be inferred that almost all the Cr$_7$C$_3$ and Cr$_3$C$_2$ were completely dissolved because of the high heat input and long interaction time. An observation of Fig. 7c reveals that more undissolved carbides with irregular shape were present in sample #3, especially the presence of Cr$_7$C$_3$ with large block shape. The carbides were embedded in the shallow surface of the matrix, and their interface with the matrix appeared fuzzy. In comparison to sample #2, the size of undissolved carbides in sample #3 was larger, and the degree of ablation was lower. Furthermore, as the Cr$_7$C$_3$ generally dissolve completely at about 870 °C [31], it can be inferred that the temperature of sample #3 might have reached between Ac1 and Ac3 lines.

### 3.4.2 Microstructure evolution

Figure 8 shows the microstructure evolution of the quenched zones treated under different laser energy density. During treatment the irradiation of high-energy laser beam heated up the untreated surface rapidly, the nucleation of austenite began at the interface of ferrite and carbides (tempered sorbite) when the temperature exceeded Ac1. During transformation, the C atoms continuously dissolved by diffusion into austenite and promoted the growth of austenite. The diffusion and dissolution are a time- and temperature-dependent process, and different combinations of LSQ parameters produce conditions of different interaction time and temperatures. Consequently, different heat input and interaction time result in a very widely different microstructure of the quenched zone. The variations in undissolved carbides, type

![Fig. 7 High magnification SEM of the quenched zones of different samples: a #1, b #2, c #3](image)

![Fig. 8 Schematic diagram of the microstructure evolution in quenched zone of different LSQ samples](image)
of transformed martensite, number of RA, and the carbon content (as shown in Fig. 8) were produced as a consequence of LSQ treatment under different process parameters.

The sample #2 was treated under the lowest laser power and median laser energy density. The SEM and results of EDS reveal the presence of some undissolved carbides, which restricted the growth of austenite grains and promoted the formation of fine LM with low-carbon content. Moreover, if the cooling rate during quenching is lower than the critical cooling rate, some untransformed austenite will be retained. For the sample #1, the laser energy density was the highest, the carbides were fully dissolved, and complete austenitic transformation occurred. This resulted in the formation of the high amount of high-carbon PM after cooling. However, the sample #3 was treated at the lowest laser energy density (the scanning speed was highest). Under such a condition, a small number of undissolved carbides were observed in the microstructure. The lower temperature and higher laser scanning speed result in non-uniform element diffusion. The heavily nonuniform diffusion causes the evolution of heterogeneously distributed blocky RA (or ferrite) and produces the highly nonuniform microstructure, and the formation of low-carbon LM.

The mechanisms of laser surface quenching involve martensitic strengthening, retained austenite solution strengthening, grain refinement strengthening, and residual stress strengthening, etc. The dominant mechanism depends on the phase composition, phase distribution, and phase size. The results of studies show that the undissolved carbides in the transformed phase are beneficial in improving the wear resistance property, but the phase interface in multiphase microstructure increases the susceptibility of corrosion and reduces the corrosion resistance [7, 32, 33]. Having demonstrated a widely varied microstructural feature produced under different LSQ conditions, it is important to select the appropriate processing parameters according to the in-service performance requirements.

4 Conclusion

1. LSQ can effectively eliminate the pores and other defects on the original AISI P20 steel surface. PM, LM, RA, and some undissolved carbides are the main component of the quenched zone microstructure. Higher laser energy density is beneficial to obtain the uniform and dense microstructure, and to the formation of high-carbon martensite and the elimination of blocky RA and carbides.

2. The compounds Cr₇C₃ and Cr₅C₂ are the main carbides in the quenched zone of LSQ samples. Cr₅C₂ is harder to be dissolved and ablated under lower laser energy density irradiation because of its better thermal stability, and those undissolved Cr₅C₂ will hinder the austenite growth and enhance the formation of low-carbon martensite. When the laser energy density reaches 150 J/mm², almost all carbides dissolve, and the element distribution is the most uniform in all the samples.

3. LSQ can implant uniformly distributing residual compression stress and fine the microstructure in the quenched zone. When the laser energy density reaches 150 J/mm², the residual compression stress is the largest, with the average value of about -280.5 Mpa.

Author contribution Zhiyuan Li contributed to all parts of this work: designing the analysis, collecting data, performing analysis, writing paper. Jian Zhang conceived the idea of the paper, supervised the experiments, analyzed data, optimized details, and wrote the paper. All authors contributed to refining the ideas and finalizing this paper.

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Declarations

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References

1. Chen J-Y, Conlon K, Xue L, Rogge R (2010) Experimental study of residual stresses in laser clad AISI P20 tool steel on pre-hardened wrought P20 substrate. Mater Sci Eng, A 527:7265–7273. https://doi.org/10.1016/j.msea.2010.07.098
2. Hoseiny H, Caballero FG, Högman B, San Martín D, Capdevila C, Nordh L-G, Andrén H-O (2012) The effect of the martensitic packet size on the machinability of modified AISI P20 prehardened mold steel. J Mater Sci 47:3613–3620. https://doi.org/10.1007/s10853-011-6208-y
3. Hoseiny H, Caballero FG, M’Saoubi R, Högman B, Weidow J, Andrén H-O (2015) The influence of heat treatment on the microstructure and machinability of a prehardened mold steel. Metall and Mater Trans A 46:2157–2171. https://doi.org/10.1007/s11661-015-2789-4
4. Li HX, Qi HL, Song CH, Li YL, Yan M (2018) Selective laser melting of P20 mould steel: investigation on the resultant microstructure, high-temperature hardness and corrosion resistance. Powder Metall 61:21–27. https://doi.org/10.1080/00325899.2017.1368965
5. Telasang G, Dutta Majumdar J, Padminabhram G, Mannan I (2014) Structure–property correlation in laser surface treated AISI H13 tool steel for improved mechanical properties. Mater Sci Eng, A 599:255–267. https://doi.org/10.1016/j.msea.2014.01.083
6. Park C, Sim A, Ahn S, Kang H, Chun E-J (2019) Influence of laser surface engineering of AISI P20-improved mold steel on wear and corrosion behaviors. Surf Coat Technol 377:124852. https://doi.org/10.1016/j.surfcoat.2019.08.006

7. Telasang G, Dutta Majumdar J, Padmanabhama G, Manna I (2015) Wear and corrosion behavior of laser surface engineered AISI H13 hot working tool steel. Surf Coat Technol 261:69–78. https://doi.org/10.1016/j.surfcoat.2014.11.058

8. Moradi M, Arabi H, Nasab SJ, Benyounis KY (2019) A comparative study of laser surface hardening of AISI 410 and 420 martensitic stainless steels by using diode laser. Opt Laser Technol 111:347–357. https://doi.org/10.1016/j.optlastec.2018.10.013

9. Moradi M, KaramiMoghadam M (2019) High power diode laser surface hardening of AISI 4130; statistical modelling and optimization. Opt Laser Technol 111:554–570. https://doi.org/10.1016/j.optlastec.2018.10.043

10. Lei J, Xie J, Zhou S, Song H, Song X, Zhou X (2019) Comparative study on microstructure and corrosion performance of 316 stainless steel prepared by laser melting deposition with ring-shaped beam and Gaussian beam. Opt Laser Technol 111:271–283. https://doi.org/10.1016/j.optlastec.2018.09.057

11. Maharanj N, Murugan VK, Zhou W, Seita M (2019) Corrosion behavior of laser hardened 50CrMo4 (AISI 4150) steel: A depth-wise analysis. Appl Surf Sci 494:941–951. https://doi.org/10.1016/j.apssurfc.2019.07.172

12. Zhang W, Jiang T, Li J, Liu L (2019) Effect of laser heat treatment on the microstructure and properties of alloy 600H. Metals 9:379. https://doi.org/10.3390/met9030379

13. Chen Z, Zhu Q, Wang J, Yun X, He B, Luo J (2018) Behaviors of 40Cr steel treated by laser quenching on impact abrasive wear. Opt Laser Technol 103:118–125. https://doi.org/10.1016/j.optlastec.2018.01.039

14. Karthik D, Swaroop S (2017) Laser shock peening enhanced corrosion properties in a nickel based Inconel 600 superalloy. J Alloy Compd 694:1300–1319. https://doi.org/10.1016/j.jallcom.2016.10.093

15. Wolozs P, Baran A, Polanski M (2020) The influence of laser engineered net shaping (LENSTM) technological parameters on the laser deposition efficiency and properties of H13 (AISI) steel. J Alloy Compd 825:153840. https://doi.org/10.1016/j.jallcom.2020.153840

16. Bonagania SK, Bathulab V, Kain V (2018) Influence of tempering treatment on microstructure and pitting corrosion of 13 wt.% Cr martensitic stainless steel. Corros Sci 131:340–354. https://doi.org/10.1016/j.corsci.2017.12.012

17. Kim B, Kim S, Kim H (2018) Effects of alloying elements (Cr, Mn) on corrosion properties of the high-strength steel in 3.5% NaCl solution. Adv Mater Sci Eng. https://doi.org/10.1155/2018/7638274

18. Li Z, Zhang J, Dai B, Liu Y (2020) Microstructure and corrosion resistance property of laser transformation hardening pre-hardened AISI P20 plastic die steel. Opt Laser Technol 122:105852. https://doi.org/10.1016/j.optlastec.2019.105852

19. Hung T-P, Shi H-E, Kuang J-H (2018) Temperature modeling of AISI 1045 steel during surface hardening processes. Materials 11:1815. https://doi.org/10.3390/ma11101815

20. Jiao X, Wang J, Wang C, Gong Z, Xinxing Pang SM, Xiong, (2018) Effect of laser scanning speed on microstructure and wear properties of Ti5M clad coating fabricated by laser cladding technology. Opt Lasers Eng 110:163–171. https://doi.org/10.1016/j.optlaseng.2018.05.024

21. Wang X, Wang J, Gao Z, Xia D-H, Wenbin Hu (2018) Fabrication of graded surfacing layer for the repair of failed H13 mandrel using submerged arc welding technology. Journal of Materials Processing Tech 262:182–188. https://doi.org/10.1016/j.jmatprotec.2018.06.040

22. Liverani E, Lutey AHA, Ascari A, Fortunato A, Tomesani L (2016) A complete residual stress model for laser surface hardening of complex medium carbon steel components. Surf Coat Technol 302:100–106. https://doi.org/10.1016/j.surfcoat.2016.05.066

23. Bailey NS, Tan W, Shin YC (2009) Predictive modeling and experimental results for residual stresses in laser hardening of AISI 4140 steel by a high power diode laser. Surf Coat Technol 203:2003–2012. https://doi.org/10.1016/j.surfcoat.2009.01.039

24. Chen X, Fang Y, Zhang S, Kelleher IF, Zhou J (2015) Effects of LSP on micro-structures and residual stresses in a 4 mm CLAM steel weld joints. Fusion Eng Des 94:54–60. https://doi.org/10.1016/j.fusengdes.2015.03.019

25. Sehyeok Oh, Ki H (2017) Prediction of hardness and deformation using a 3-D thermal analysis in laser hardening of AISI H13 tool steel. Appl Therm Eng 121:951–962. https://doi.org/10.1016/j.applthermeng.2017.04.156

26. Zhang C, Wang Y, Zhang Y, Li JinHua, Zeng H, Zhang DeQiang (2015) Microstructure and wear-resistant properties of NiCr-Cr-C coating with Ni45 transition layer produced by laser cladding. Rare Met 34(7):491–497. https://doi.org/10.1007/s12598-015-0492-7

27. Hebbale AM, Srinath MS (2018) Microstructural studies of cobalt based microwave clad developed on martensitic stainless steel (AISI-420). Trans Indian Inst Met 71(3):737–743. https://doi.org/10.1007/s12666-017-1206-7

28. Music D, Kreissig U, Mertens R, Schneider JM (2004) Electronic structure and mechanical properties of Cr$_3$C$_2$. Phys Lett A 326:473–476. https://doi.org/10.1016/j.physleta.2004.04.068

29. Xiao B, Feng J, Zhou CT, Jiang YH, Zhou R (2011) Mechanical properties and chemical bonding characteristics of type multicomponent carbides. J Appl Phys 109:023507. https://doi.org/10.1063/1.3532038

30. Chong XiaoYu, Jiang YeHua, Zhou R, Feng J (2017) Multialoying effect on thermophysical properties of Cr$_7$C$_3$-type carbides. J Am Ceram Soc 100:1588–1597. https://doi.org/10.1111/jace.14694

31. Zhou J, Ma S, Chi X, Chen Z, Li X (2013) Microstructure and properties of hot working die steel H13MOD. J Iron Steel Res Int 20(9):117–125. https://doi.org/10.13228/j.boyuan.iss.1006-706x.2013.09.011

32. Boztepe E, Alves AC, Ariza E, Rocha LA, Cansever N, Toptan F (2018) A comparative investigation of the corrosion and tribocorrosion behaviour of nitrocarburized, gas nitrided, fluidized-bed nitrided, and plasma nitride plastic mould steel. Surf Coat Technol 334:116–123. https://doi.org/10.1016/j.surfcoat.2017.11.033

33. Sundqvist J, Manninen T, Heikkinen H-P, Anttila S, Kaplan AFH (2018) Laser surface hardening of 11% Cr ferritic stainless steel and its sensitisation behaviour. Surf Coat Technol 344:673–679. https://doi.org/10.1016/j.surfcoat.2018.04.002

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