Laser treatment of dual matrix cast iron with presence of WC particles at the surface: Influence of self-annealing on stress fields

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A B S T R A C T
Laser control melting of dual matrix cast iron surface is carried out. A carbon film containing 15% WC particles is formed at the surface prior to the laser treatment and the spiral tracks are adopted for laser scanning at the workpiece surface. Morphological, metallurgical, microhardness, and scratch resistance of the laser treated parts are examined using analytical tools. Temperature and stress fields in the laser irradiated region are predicted incorporating ABAQUS finite element code. Predictions of temperature and residual stress at the laser treated surface are validated with the thermocouple and the X-ray diffraction data. It is found that surface temperature and residual stress predictions agree well with their counterparts corresponding to thermocouple data and findings of X-ray diffraction technique. Laser treated surface is free from asperities including voids and micro-cracks despite the mismatch of thermal expansion coefficients of WC and dual matrix cast iron. This behavior is attributed to the self-annealing effects of recently formed spiral tracks on the previously formed tracks during the laser treatment process; in which case, the self-annealing effect modifies the cooling rates and lowers thermal stress levels in the laser treated layer. Laser treated layer consists of a dense region composed of fine grains and WC particles, dendritic and featherlike structures below the dense layer, and the heat affected zone.

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1. Introduction

Laser treatment of metallic surfaces improves tribological and mechanical characteristics of the surfaces [1] and extends the fatigue life [2]. In addition, laser processing has several advantages over the conventional methods because of the local treatment, precision of operation, and the low cost. In laser surface processing, many factors affect the end product quality assessed by surface texture, microhardness, scratch resistance, microstructure, morphology, and etc. The selection of the process parameters plays a major role in achieving the desired end product quality of the laser treated surface. Since laser treatment involves high temperature processing, rapid solidification under high cooling rates results in high thermal stress levels in the laser treated region. This, in turn, causes asperities such as voids and micro-crack networks in the treated region and limits the practical applications of the laser treated surfaces. Laser control melting improves the surface quality; however, the residual stress formed at the surface region lowers the fracture toughness and the performance of the surface response to the tribological tests [3]. The self-annealing effect created during laser treatment is one of the solutions to minimize the high stress field formed in the laser treated layer. In this case, heat conduction across the path of laser scanning and laser repetition rate during the surface processing may create a self-annealing effect in the treated layer. Moreover, the laser treatment of metallic surfaces with presence of hard particles, such as WC, require proper control and appropriate selection of the laser treatment parameters due to the mismatch of thermal-mechanical properties of the constituting elements in the treated layer. On the other hand, the model studies provide useful information about the physical processes, in terms of thermal and stress fields, taking place during the laser treatment process. Consequently, investigation of the laser treatment of metallic surfaces with presence of hard particles and thermal stress field developed in the treated layer becomes essential.

Considerable research studies were carried out to examine laser treatment of steel and iron based composites. Laser treatment of dual matrix structured cast iron surface was investigated by Sun et al. [4]. They showed that the wear resistance of the treated layer was almost 1.6 times of that of the nodular cast iron substrate. The improvement in wear resistance was due to combined results of
the grain refining effect, the solution strengthening effect, the distribution of the hard phases, the work hardening effect of the retained austenite, and the good bonding between these hard phases and the Fe-based matrix. Laser alloying of cast iron and nano-mechanical properties were examined by Yang et al. [5]. They demonstrated that the resistances of plastic deformation and oxidation reaction in alloying zones were main reasons for inhibiting cracking. The alloying zone could also be considered as enlarged hard phase because of the high nano-hardness, which contributed to cracks blocking. Laser surface melting of nodular cast iron was studied by Benyounis et al. [6]. They indicated that laser treatment modified the surface morphology. Investigation of the wear characteristics of laser surface treated cast iron cylinder was carried out by Duffet et al. [7]. They showed that high density of lamellae structure was formed in the laser treated layer, which improved the tribological properties of the surface. The influence of laser surface texturing on the frictional behavior of cast iron was studied by Kim et al. [8]. They demonstrated that the aspect ratio of the dimples formed during the laser treatment was found to be the most significant factor influencing the friction coefficient; however, the effect of the surface density of the dimples on the coefficient of friction was only marginal. The friction and wear behavior of nodular cast iron modified by a laser micro-precision treatment were investigated by Xia et al. [9]. Their findings revealed that the substantial increase in the wear-resistance of the cast iron was observed after the laser treatment process. This could be attributed to a significant increase in the surface hardness of the laser-modified layers. Laser surface hardening of austempered ductile iron grades was examined by Soriano et al. [10]. They showed that a coarse martensite with retained austenite structure was found in the treated area, which resulted in a wear resistant and high hardness effective layer of 0.6–1 mm thickness. Laser cladding of Co-based alloy on cast iron was carried out by Ocelík et al. [11]. They developed the relationships between the laser cladding parameters (i.e., laser beam scanning speed, laser output power, and powder feeding rate) and the geometrical characteristics of a single laser track (height, width, and dilution). Surface morphology of stainless steel irradiated by a nanosecond Nd:YAG pulsed laser was studied by Liu et al. [12]. They indicated that irradiation due to several consecutive pulses caused significant damage and enhanced the stainless steel surface roughness. The characteristics of treated zone processed by pulsed Nd:YAG laser surface remelting on hot work steel were investigated by Zhang et al. [13]. They showed that different combinations of average peak power density and effective peak power density could vary the appearance of cross-sectional morphology, microstructure and hardness. Investigation of the different surface morphologies formed on AISI 304 stainless steel via millisecond Nd:YAG pulsed laser was carried out by Cui et al. [14]. They demonstrated that the smooth surface was not obtained under over-high laser energy densities; in addition, the schematic relationship was used to describe the formation process and mechanism of different surface morphologies. Surface properties of low alloy steel treated by plasma nitrocarburizing prior to laser quenching process were investigated by Wang et al. [15]. They indicated that the laser quenched layer exhibited enhanced wear resistance, due to the lubrication effect and optimized impact toughness, which was contributed to the formation of oxide film consisting of low nitrogen compound and iron oxidation. Laser surface modification of thermally-sprayed Diamalloy 2002 coating was carried out by Gisario et al. [16]. They indicated that laser treatment modified the surface morphology and the surface textures with low roughness. Laser surface treatment of aluminum based composite mixed with B4C particles was examined by Yilbas et al. [17]. They demonstrated that fine grains and ultra-short dendrites were formed in the surface region of the
laser treated layer. In addition, partially dissolved B4C particles were observed in the treated layer.

Although laser surface treatment of alloys were studied previously [17,18], the main focus was to examine the surface morphology when hard particles were present and the microhardness in the laser treated layer. In addition, the corrosion resistance of the laser treated surface due to the presence of SiC particles is investigated in the previous study [19] and the mechanical properties including microhardness, residual stress, and friction coefficient of the surface were left for the future study. Thermal stress analysis and the residual stress formed at the treated surface of the dual matrix cast iron with presence of high concentration of WC were left for the future study. Since thermal expansion coefficient of WC particles differs from that of SiC particles, stress levels formed around the particles are expected to be different than those corresponding to SiC particles in the laser treated layer. In addition, the differences in thermal properties of WC and SiC particles affect the cooling rates and solidification in the laser treated layer, particularly in the close region of the hard particles. This modifies the microstructure around the hard particles in the surface region while modifying the residual stress, microhardness, and friction coefficient of the resulting surface. Consequently, in the present study, laser surface treatment of dual matrix cast iron was investigated. A carbon film accommodating 15% WC is formed at the workpiece surface prior to the laser treatment process. The surface characteristics including surface morphology, microhardness, and the residual stress formation are examined using the analytical tools. Metallurgical changes in the laser treated region and the scratch resistance of the treated surface are also studied. Thermal stress field developed in the cooling cycle of the laser treatment process is predicted numerically using the ABAQUS finite element code.

2. Experimental

The dual matrix structured cast iron was used as the workpiece. The chemical composition of the workpiece is given in Table 1. To produce dual matrix structures (DMS) with different ausferrite volume fractions (AFVF), as cast specimens were intercritically produce dual matrix structures (DMS) with different ausferrite

<table>
<thead>
<tr>
<th></th>
<th>WC</th>
<th>N</th>
<th>Fe</th>
</tr>
</thead>
<tbody>
<tr>
<td>As received</td>
<td>15</td>
<td>0.0</td>
<td>Balance</td>
</tr>
<tr>
<td>Laser treated</td>
<td>15</td>
<td>6.2</td>
<td>Balance</td>
</tr>
</tbody>
</table>

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The dual matrix structured cast iron was used as the workpiece. The chemical composition of the workpiece is given in Table 1. To produce dual matrix structures (DMS) with different ausferrite volume fractions (AFVF), as cast specimens were intercritically austenitized at the dual phase region of 810 °C for 90 min and then rapidly transformed to a salt bath containing 50% KNO3+50% NaN03 held at 315 °C and 375 °C for austempering for 120 min. The details of the heat treatment process are given in the previous study [20]. The circular workpieces with 28 mm x 2 mm (diameter x thickness) were prepared. The water soluble phenolic resin was mixed with 15%(wt) of WC powders of about 2 μm particle size and homogeneous mixing was ensured prior to applying at the workpiece surface. The workpieces were placed in a furnace with a controlled chamber at 8 bar pressure and 175 °C for 2 h to form a 40 μm thick film at the surface. The workpieces were then heated to 370 °C in an argon environment for 6 h to ensure the conversion of the phenolic resin into carbon. The pre-prepared sample surfaces were scanned by a laser beam according to the parameters given in Table 2.

The CO2 laser (LC-ALPHAIII) delivering a nominal output power of 2 kW was used to irradiate the workpiece surface. The nominal focal length of the focusing lens was 127 mm and the diameter of the laser beam focused at the workpiece surface was ~0.3 mm. Nitrogen was used as the assisting gas and it was applied coaxially with the laser beam using a conical nozzle. Several tests were conducted to obtain asperity free laser treated surface. In this case, laser parameters resulting in controlled melting of the surface with a minimum of surface defects, such as very small cavities without crack networks, were selected and the laser treatment conditions are given in Table 2. It should be noted that initial tests were carried out to identify the appropriate laser scanning parameters. This is due to that increasing the laser output power by 10% caused the formation of deep cavities with the irregular shapes at the surface due to evaporation during laser scanning. In addition, similar effects were noted when the laser scanning speed was reduced by 6%. On the other hand, increasing laser scanning speed by 5% reduces partial melting at the workpiece surface. The workpiece surface was scanned by a laser through a spiral orientation as shown in Fig. (1). This arrangement provides self-annealing effect of the recently formed laser tracks on the previously formed tracks.

Material characterization of the laser treated surfaces was conducted using an optical microscope, SEM, EDS, and XRD. A Jeol 6460 Scanning Electron Microscope was used for SEM examinations and a Bruker D8 Advanced X-ray Diffractometer using CuKα radiation was used for XRD analysis. Typical settings of the XRD were 40 kV and 30 mA with the scanning angle (2θ) ranging from 20° to 130°.

The residual measurement is based on the position of the diffraction peak, which exhibits a shift as the specimen is rotated by an angle ψ. The magnitude of the shift is related to the magnitude of the residual stress. The relationship between the peak shift and the residual stress (σ) is given [21,22]:

$$\sigma = \frac{E}{(1 + \nu) \sin^2 \psi} \left( d_o - d_s \right) / d_o$$  \hspace{1cm} (1)

where E is Young’s modulus, ν is Poisson’s ratio, ψ is the tilt angle, and ds are the d spacing measured at each tilt angle. If there are no shear strains present in the specimen, the d spacing changes linearly with sin2ψ. The (114) reflection from the ε-Fe3N was used for residual stress measurements. The measurements were carried out normal to the workpiece surface and Fig. (2) shows the linear dependence of d(114) on sin2ψ in the region of laser treated surface. The ε-Fe3N peak takes place at 128.445°, which corresponds to d(114) plane with the inter-planer spacing of 0.8570 Å. The linear dependence of d(114) in Fig. (2) results in the slope of $-1.714 \times 10^{-13}$ m/degrees and the intercept of 0.8570 Å. The residual stress, therefore, determined from the XRD technique at the surface vicinity is on the order of 300 ± 20 MPa. XRD measurements are repeated three times and the error related to the measurements is on the order of 3%.

A Microphotronics digital microhardness tester (MP-100TC) was used to determine the microhardness at the surface. The standard test method for Vickers indentation hardness of advanced ceramics (ASTM C1327-99) was adopted. Microhardness was measured at the surface of the workpiece after the laser treatment process and the measurements were repeated five times at each location to determine the consistency of the results.

A linear micro-scratch tester (MCTX-S/N: 01-04300) was used to determine the scratch resistance and friction coefficient of the laser treated and untreated surfaces. The equipment was set at a contact load ranging from 0.03 to 5 N. The scanning speed was 5 mm/min with a loading rate of 5 N/s and the total length for the scratch tests was 5 mm.

To validate temperature predictions, a thermocouple was used to monitor the temporal variation of surface temperature at the
data were sampled along the time span due to the transient nature of the heating process. The time corresponding to the maximum temperature was marked and temperature variation due to the repeat was noted. Since the maximum temperature variation remained within 5% over the repeats, the confidence for the measurements was found to be satisfactory. However, to validate the confidence level established for the measurements, the extra repeats were carried out to measure temperature. In this case, twenty more repeats were carried out. It was observed that the maximum measured temperature remained within 5% error limit.

3. Mathematical analysis

In thermal analysis of laser heating, the phase change is considered in the irradiated region. In addition, the solid body heat conduction with temperature-dependent conductivity, internal energy (including latent heat effects), and convection and radiation boundary conditions are considered. In the case of a moving heat source along the \(x\)-axis with a constant velocity \(U\) (Fig. (1)), the Fourier heat transfer equation, for the laser heating process, can be written as [24]:

\[
\rho \frac{\partial (\rho C_T T)}{\partial t} = (\nabla \nabla) (\rho C_T T) + \rho U \frac{\partial (\rho C_T T)}{\partial x} + S_o
\]

(2a)

where

\[
Q = I_o (1 - r_f) e^{-\left(\frac{x - r \sin^2 \psi - y - r \cos \theta}{a^2}\right)}
\]

(2b)

\(S_o\) is the volumetric source term resembling the absorbed laser intensity during the scanning and \(I_o\) is the peak power intensity. In the above equation, \(r\) and \(\theta\) are used to present Archimedean spiral, where the laser follows the track during the scanning at the workpiece surface (Fig. (1)). However,

\[
r = p + q \theta
\]

(3)

with \(p\) and \(q\) are the real numbers. In this case, changing the parameter \(p\) turns the spiral while \(q\) controls the distance between the successive turnings along the track. In the present case, the following relation is used among the parameters and the laser scanning speed (\(U\)).

\[
\theta = \left[\frac{U}{q}\right] t
\]

(4)

The values of \(U\), \(q\), and \(p\) are selected according to the experimental conditions, which are \(U = 0.1\) m/s, \(q = 25 \mu m\), and \(p = 100 \mu m\).

In Eq. 1, \(I_o\) is laser power peak density, \(\delta\) is the absorption coefficient, \(a\) is the Gaussian parameter, \(r_f\) is the surface reflectivity, \(\rho\) is the density, and \(x\) and \(y\) are the axes while the laser beam scans the surface along the \(x\)-axis. The absorption coefficient of the incident radiation is considered to be \(6.17 \times 10^6\) m\(^{-1}\). The laser beam axis is the \(z\)-axis (Fig. (1)). It should be noted that the laser beam intensity distribution is assumed to be Gaussian at the irradiated surface.

The convective and radiation boundary conditions are considered at the free surface of the workpiece. Therefore, the corresponding boundary condition is:

<table>
<thead>
<tr>
<th>Scanning speed (m/s)</th>
<th>q ((\mu m))</th>
<th>p ((\mu m))</th>
<th>Laser power (W)</th>
<th>Frequency (Hz)</th>
<th>Nozzle gap (mm)</th>
<th>Nozzle diameter (mm)</th>
<th>N(_2) pressure (kPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.1</td>
<td>25</td>
<td>100</td>
<td>110</td>
<td>1500</td>
<td>1.5</td>
<td>1.5</td>
<td>600</td>
</tr>
</tbody>
</table>

Table 2
Laser processing parameters.
At the irradiated surface (top surface) along the scanning region, the convection and radiation boundary is considered:

$$\frac{dT}{dz} = \frac{h_f}{k} (T_f - T_{amb}) + \frac{\varepsilon_s}{k} (T_s^4 - T_{amb}^4)$$

(5)

where $h_f (= 3000 \text{ W/m}^2 \text{K})$ [25]) is the forced convection heat transfer coefficient due to the assisting gas.

At the bottom surface:

$$\frac{dT}{dz} = h (T_i - T_{amb}) + \frac{\varepsilon_s}{k} (T_s^4 - T_{amb}^4)$$

(6)

where $h (= 20 \text{ W/m}^2 \text{K})$ is the heat transfer coefficient due to natural convection, and $T_i$ and $T_{amb}$ are the surface and ambient temperatures, respectively, $\varepsilon$ is the emissivity ($\varepsilon=0.9$ is considered), $\sigma$ is the Stefan-Boltzmann constant ($\sigma=5.67 \times 10^{-8} \text{ W/m}^2\text{K}^4$). At far away boundary (at edges of the solution domain) constant temperature boundary is assumed ($T=293 \text{ K}$), i.e.

$$x = \infty; y = \infty; z = t_h \rightarrow T = 293 \text{ K}$$

(7)

where $t_h$ is the thickness of the workpiece ($t_h=0.002 \text{ m}$). Initially (prior to laser scanning), the substrate material is assumed to be at constant ambient temperature, i.e. $T=T_{amb}$, which is considered as constant ($T_{amb}=293 \text{ K}$).

Eq. 1 is solved numerically with the appropriate boundary conditions to predict the temperature field in the substrate material. Table 3 gives the data used in the simulations in line with the experimental conditions. In order to incorporate the melting during the heating process, the enthalpy method is used [26]; in which case, the specific heat capacity is associated with the latent heat of melting of the substrate material, i.e. $C_p(T) = \frac{U}{\Delta T}$, where $U$ is the internal energy gain during the phase change and it is associated with the latent heat of fusion. The melting is treated separately in terms of solidus and liquids temperatures (the lower and upper temperature bounds of the phase change range) and it is the part of the total internal energy gain [17]. Since the primary interest is the stress field developed in the laser irradiated section, the flow field generated in the liquid phase in the surface vicinity during the laser melting process is omitted. This is because that the thermal stress field is considered to be almost zero in the melt pool, since only negligibly small hydrodynamic pressure is present in the melt pool.

Since the solidification involves a small strain, the assumption of small strain is adopted when modeling the thermal stresses. The thermal strains which dominate thermo-mechanical behavior during solidification are on the order of only a few percent [27]. Several previous solidification models [27] confirm that the solidified metal part undergoes only small deformation during solidification. The displacement spatial gradient is small $\nabla u = \sigma/\kappa$, so $\sigma u = \nabla u$ and the linearized strain tensor becomes [28]:

$$\varepsilon = \frac{1}{2} [\nabla u + (\nabla u)^T]$$

(8)

The small strain formulation can be used, where Cauchy stress tensor is identified with the nominal stress tensor $\sigma$, and $b$ is the body force density with respect to initial configuration.

$$\nabla \cdot \sigma + b = 0$$

(9)

The rate representation of total strain in this elastic-viscoplastic model is given by [29]:

$$\dot{\varepsilon} = \dot{\varepsilon}_{el} + \dot{\varepsilon}_{pl} + \dot{\varepsilon}_{th}$$

(10)

where $\dot{\varepsilon}_{el}$, $\dot{\varepsilon}_{pl}$, $\dot{\varepsilon}_{th}$ are the elastic, inelastic (plastic + creep), and thermal strain rate tensors respectively. Stress rate $\dot{\sigma}$ depends on elastic strain rate, and in this case of linear isotropic material and
negligible large rotations, it is given by Eq. 11 in which “·” represents inner tensor product.

\[
\dot{\epsilon} = D \left( \mathbf{e} - \dot{\epsilon}_e - \dot{\epsilon}_s \right)
\]

(11)

\[\mathbf{D} = \frac{2}{\mu} \mathbf{I} + \left( K_0 - \frac{2}{3} \mu \right) \mathbf{I} \otimes \mathbf{I}
\]

(12)

Here \(\mu, K_0\) are the shear modulus and bulk modulus respectively and \(\mathbf{I}, \mathbf{I}\) are fourth and second order identity tensors and “\(\otimes\)” is the notation for outer tensor product.

Inelastic strain includes both strain-rate independent plasticity and time dependent creep. Creep is significant at the high temperatures of the solidification processes and is indistinguishable from plastic strain [30]. The inelastic strain-rate is defined here with a unified formulation using a single internal variable [31,32]. Equivalent inelastic strain \(\dot{\epsilon}_e\) characterizes the microstructure.

The equivalent inelastic strain-rate \(\dot{\epsilon}_e\) is a function of equivalent stress \(\sigma\), temperature \(T\), equivalent inelastic strain \(\epsilon_{ie}\).

\[
\dot{\epsilon}_e = f(\sigma, T, \epsilon_{ie})
\]

(13)

\[
\dot{\sigma} = \frac{2}{3} \dot{\epsilon}_e \delta_{ij}
\]

(14)

\[\delta_{ij} = \sigma_{ij} - \frac{1}{\sigma} \delta_{ijk} \delta_{ij}
\]

(15)

The workpiece is assumed to harden isotropically, so the von Mises loading surface, associated plasticity and normality hypothesis in the Prandtl–Reuss flow law is applied [32]:

\[
\dot{\epsilon}_{ie} \epsilon_{ij} = \frac{2}{3} \dot{\epsilon}_e \delta_{ij} \sigma
\]

(16)

Thermal strains arise due to volume changes caused by both temperature differences and phase transformations, including solidification and solid-state phase changes, i.e.

\[
\dot{\epsilon}_{th} \epsilon_{ij} = \int_{T_0}^{T} \delta_{ij} \alpha \mathbf{T} d\mathbf{T}
\]

(17)

where \(\alpha\) is coefficient of thermal expansion, and \(T_0\) is the reference temperature and \(\delta_{ij}\) is Kronecker’s delta. It is normally accounted for in stress analyses through a temperature-dependent differential thermal expansion coefficient, \(\alpha(T)\). ABAQUS/Standard analysis a spatially varying thermal expansion can be defined for homogeneous solid continuum elements by using a distribution, which includes the tabulated values for the thermal expansion [26]. ABAQUS uses an implicit backward-difference scheme for time integration of both temperature and displacements at every material integration point [26].

4. Numerical simulation

Finite element discretization was carried out using the ABAQUS software [26]. The fixed boundary conditions are applied on both ends of the workpiece resembling the experimental laser heating situation. Laser heat flux with Gauss distribution and prescribed velocity of 10 cm/s along the x-axis through user sub-routine DFLUX is applied to the thermal model. The Gauss parameter “a” is \(a = 0.0003\) m, in accordance with the experimental power intensity distribution. The thermal model consisted of two steps. The first step, which lasts 0.05 s, simulates the response of the plate under moving laser heat flux. The second step, which lasts for 1000 s, simulated the continued cooling in the model. Cooling was allowed to continue until all of the plate reaches initial temperature (room temperature). The temperature–time history resulted from the thermal analysis is used as input to the thermal stress analysis. The workpiece is considered as an elastic body, which is modeled as von Mises elastic–plastic material with isotropic hardening and with a yield stress that changes with temperature. Table 3 gives the properties of dual matrix cast iron used in the simulations [33].

5. Results and discussion

Laser control melting of dual matrix cast iron with presence of 15% WC particles at the surface is carried out and metallurgical and morphological changes in the treated laser are examined using the analytical tools. Thermal stress field developed in the treated layer is predicted using the ABAQUS finite element code [26] while the residual stress formed in the surface region is measured incorporating the XRD technique.

5.1. Morphological, metallurgical, and surface properties

Fig. (3) shows top view of SEM micrographs of laser treated surfaces. SEM micrographs reveal that WC hard particles are apparent at the treated surface. The scanning tracks are formed by overlapping of laser irradiated spots (Fig. (3a) and (b)), which is associated with the repetitive pulses of the laser beam. During the scanning, the laser beam irradiates the workpiece surface via repetitive pulses of a high frequency (1500 Hz) and the workpiece movement results in 68% overlapping ratio of the laser irradiated spots while forming the tracks. Since the laser pulse intensity distribution at the irradiated spot is Gaussian, the peak intensity occurs at the spot center. This in turn gives rise to a partial evaporation of the surface during the laser scanning. Therefore, small scale cavities are formed within the central region of the irradiated spot. However, the cavity size in width and depth is limited at the center of the irradiated spot because of: i) the sharp decay of the intensity across the irradiated spot (Gaussian intensity distribution), and ii) the melt flow in the near region of the cavity, which modifies the cavity shape in such a way that the cavity size and its depth become small and shallow at the surface (Fig. (3c)). Over flow of the melt across the laser scanning tracks is not observed from the SEM images. This indicates that laser control melting localizes the phase change at the workpiece surface, which in turn lowers surface roughness. It should be noted that excessive phase change (melting) at the surface causes the melt flow due to surface tension force and gives rise to undulation at the surface. The surface roughness of the as received surface is in the order of 0.6 \(\mu\) and the laser treated surface is in the order of 1.6 \(\mu\). Consequently, laser treatment increases the surface roughness of the workpiece because of the formation of the regular laser scanning tracks. The volume shrinkage around the hard particles (WC) causes scattered few micro-voids at the surface (Fig. (3d)), which are attributed to the mismatched between the thermal expansion coefficients of dual matrix cast iron (11.5 \(\times\) 10^{-6}/K [33]) and WC (10^{-6}/K [34]) and although the thermal expansion coefficients of WC and dual matrix cast iron are different, laser treated surface is free from microcracks or crack networks. It should be noted that thermal diffusivity mismatch between the constituting elements in the surface region results in differences in thermal expansion and contraction. This gives rise to development of micro-stresses around the hard particles. However, no cracks around the hard particles are observed (Fig. (3d)). Therefore, it is possible that the stress developed around the hard particles due to mismatch of
thermal expansion coefficients remains low. This can be explained in terms of the self-annealing effect of the laser spiral scanning of multiple tracks, which suppresses the high stress levels in the laser treated layer. Therefore, the heat conducted from the recently formed laser scanning tracks acts as a heat source lowering the cooling rates at the surface. Self-annealing affect also suppresses the thermal stress developed due to convection cooling of the assisting gas, since nitrogen is used as an assisting gas during the laser treatment process. Therefore, the self-annealing affect created during the laser spiral scanning modifies the cooling rates. It should be noted that the heat conduction below the surface in between the scanning tracks modifies the cooling rates in the region below the surface. This results in self-annealing effect in the laser treated region. This situation can be observed from temperature variation along OB in Fig. (9), i.e. the maximum and minimum temperature difference remains low.

(Figs. (4) and 5) show SEM micrographs of the cross-section of the untreated and laser treated section, respectively. Ausferrite microstructure with presence of graphite spheroids is seen in Fig. (4), which resembles the untreated workpiece. Laser treated layer extends uniformly with a thickness of about 40 μm below the surface (Fig. (5a)). In general, the laser treated layer is free from large scale defect sites including voids and micro-cracks. Although volume shrinkage and grain refinement in the surface region results in high thermal stress levels in this region, presence of almost uniformly extending treated layer thickness with no voids and cracks indicates that stress levels in this region is not significantly high. This is attributed to the self-annealing effect of lately formed laser scanning tracks on the previously formed tracks. Since the laser scanning follows a spiral motion, heat conduction among the scanning tracks modifies the cooling rate in the laser treated region. In this case, low cooling rates initiates the slow cooling in the treated region while reducing the thermal stress levels in the treated layer. Moreover, the laser treated layer, mainly, consists of three regions. The first region corresponds to the fine grains with WC particles while forming a dense layer (5b). Since the melting
temperature of WC is higher than that of the dual matrix cast iron, they remain in the solid phase and dissolution of WC particles is not observed in the dense layer. Due to high rates at the surface including the convection cooling of the assisting gas, fine grains are formed in the surface region of the treated layer. The mismatch between the thermal expansion coefficient of dual matrix cast iron [34] and WC particles [35], causes formation of micro-stresses in the near region of WC particles due to the contraction in the cooling period. Since the micro-cracks are not observed around WC particles, the micro-stress levels are not sufficiently high causing the thermally induced micro-cracks in this region. This is also attributed to the self-annealing effect of the lately formed laser scanning tracks on the initially formed tracks. The second layer composes of phase mixture of ferrite and carbides, which transforms into martensite and very fine carbides in the vicinity of the dense layer (Fig. 4c). This behavior is associated with the high cooling rates in the vicinity of the surface region. However, increasing depth below the surface vicinity modifies the cooling rates; in which case, a feathery like structure is observed (Fig. 5d). The formation of the feathery like structures can be related with the nitrogen diffusion from the surface along the grain boundaries [36]. It should be noted that high pressure

![Fig. 5. SEM micrographs of cross-section of laser treated layer: a) thickness of laser treated layer, b) dense layer with WC, c) Martensitic structure (marked in red circles), d) Feathery like structures, e) Fine dendrites, f) Pearlite formation.](image-url)
nitrogen is used during the laser scanning to avoid oxidation reactions at the surface during the laser treatment process. The oxidation reactions cause reduction of carbon from WC through forming W2C and carbonic gases (CO and CO2) at high temperature [37]. Moreover, the presence of nitride phase is also evident from X-ray diffractogram, which is shown in Fig. (6). Some fine dendrites (Fig. (5e)) are also formed towards the edges of the second region while indicating the high cooling rates in this region. The orientation of the dendrites in this zone is multi-directional because of the non-uniform cooling during the solidification cycle. However, martensite is transferred to ferritic-pearlitic structures (Fig. (5f)) in the third region because of relatively lower cooling rates as compared to that of the first and the second regions. In this region, micro-cracking, which, in general, occurs during

Table 4
Elemental composition of laser treated surface obtained from EDS data (wt%). Spectrum corresponds to a location at the surface and EDS data represent weight percentile of constituting elements.

<table>
<thead>
<tr>
<th>Spectrum</th>
<th>N</th>
<th>W</th>
<th>O</th>
<th>Fe</th>
</tr>
</thead>
<tbody>
<tr>
<td>Spectrum 1</td>
<td>5.3</td>
<td>14.1</td>
<td>2.4</td>
<td>Balance</td>
</tr>
<tr>
<td>Spectrum 2</td>
<td>4.6</td>
<td>14.8</td>
<td>2.8</td>
<td>Balance</td>
</tr>
<tr>
<td>Spectrum 3</td>
<td>6.1</td>
<td>14.7</td>
<td>1.7</td>
<td>Balance</td>
</tr>
<tr>
<td>Spectrum 4</td>
<td>6.3</td>
<td>14.1</td>
<td>1.9</td>
<td>Balance</td>
</tr>
</tbody>
</table>

Table 5
Microhardness and residual stress at the workpiece surface prior to and after the laser treatment process.

<table>
<thead>
<tr>
<th></th>
<th>Hardness (HV)</th>
<th>Residual stress (MPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>As received surface</td>
<td>286 (+10/–10)</td>
<td>–</td>
</tr>
<tr>
<td>Laser treated surface</td>
<td>720 (+20/–20)</td>
<td>300 (+10/–10)</td>
</tr>
</tbody>
</table>

Fig. 6. X-ray diffractogram of laser treated surface.

Fig. 7. Friction coefficient of laser treated and untreated surfaces.

Fig. 8. Temporal variation of surface temperature predicted from the simulations and obtained from the thermocouple data at point T (Fig. (1)). The location of T is 0.45 mm away from the laser beam in the final turn during the spiral scanning.

Fig. 9. Temperature variation along the lines OA and OB (Fig. (1)) for different cooling periods.
The predictions of laser heating process are presented in line with the experimental conditions. Temperature and residual stress predictions are validated through the experimental data.

Fig. (8) shows surface temperature predicted from the simulations and obtained from the thermocouple data. Temperature predictions agree well with the thermocouple data. The small discrepancies are associated with the experimental error, which is in the order of 5%, and the assumptions made in the simulations, such as isotropic properties are incorporated in the simulations. Nevertheless, the discrepancies are considerably small. In the case of the Residual stress, the residual stress predicted in the surface region is in the order of $-280$ MPa and the residual stress obtained from the XRD technique is in the order of $-300$ MPa. Consequently, residual stress predicted in the surface region agrees well with its counterpart obtained from the XRD technique.

Fig. (9) shows temperature variations predicted along the paths OA and OB (Fig. (1)) for various cooling periods while Fig. (10) shows temperature contours inside the workpiece at the onset of cooling period. It should be noted that the cooling period is initiated after 1 s of the heating duration. Temperature remains high in the region where the laser beam is last located prior to initiation of the cooling period. The peak temperature remains higher along the path OB as compared to that corresponding to along the path OA. This is attributed to the spiral scanning of the laser beam, which is close to the point 0.0074 m along the OB line. Temperature gradient along the irradiated spot is high; however, as the cooling period progresses, temperature and the temperature gradient reduce along the both paths. As the cooling period progresses further, temperature reduces to the initial temperature and cooling period ceases. This corresponds to the cooling period of 500 s. Fig. (11) shows von Mises stress distribution along the paths OA and OB for various cooling periods while Fig. (12) shows von Mises stress contours in the workpiece surface at the initiation of the cooling period. von Mises stress attains lower values in the region where temperature is high at the initiation of the cooling period, which is particularly true along the path OB. This behavior is attributed to temperature dependent elastic modules, which reduces with increasing temperature (Table 3). However, von
Mises stress remains high in the central region of the laser treated workpiece where laser treatment first initiated (Fig. (1)). This is particularly as the cooling period progresses. The attainment of high values of von Mises stress is associated with the relatively higher cooling rates as compared to that occurring in the other regions. The maximum von Mises stress is in the order of 300 MPa, which is lower than that reported in the previous study [38], which is in the order of 1 GPa. The values of von Mises stress is related to the spiral scanning of the laser beam during the treatment process, which generates a self-annealing effect while modifying cooling rates in the laser treated layer. It should be noted that in the previous study [38], laser scanned the surface along the parallel tracks and self-annealing effect was minimal.

Fig. (13) shows temperature variation along the $z$-axis for various cooling periods. It should be noted that $z=0$ m corresponds to the free surface and temperature variation is shown as reference to point O at the surface (Fig. (1)). Temperature remains high in the surface region at the onset of cooling period and as the distance increases towards the workpiece bulk it decays sharply. This results in high temperature gradients in the region below the surface. As the cooling period progresses, temperature reduces to initial temperature; in which case, the cooling period ends.

Fig. (14) shows von Mises stress along the $z$-axis for various cooling periods and location is O at the surface, which is similar to that is indicated in Fig. (13); von Mises stress attains the high values in the region close to the surface during the late periods of
the cooling cycle. In this case, high temperature gradient is responsible for the attainment of the high stress levels in this region. However, during the early period of the cooling cycle, von Mises stress becomes small in the surface region. This behavior is attributed to reduced elastic modules at high temperatures (Table 3). However, during the early period of the cooling cycle, von Mises stress becomes small in the surface region. This behavior is attributed to the attainment of the high stress levels in this region. However, temperature and stress fields are predicted using the ABAQUS finite element code. Surface temperature and residual stress predictions are validated with the experimental data. It is found that temperature predictions agree well with the thermocouple data. The residual stress predicted in the surface vicinity also agrees well with that obtained from the X-ray diffraction technique. Laser treated surface is free from the defect sites including voids and cracks despite the micro-stresses formed in the near region of the hard particles due to thermal expansion coefficient mismatch of WC and the dual matrix cast iron. This behavior is attributed to the spiral scanning of the laser beam at the surface, which generates a self-annealing effect on the previously formed laser scanning tracks. Laser treated layer consists of a dense region composed of fine grains and WC particles, dendritic and featherlike structures below the dense layer, and the heat affected zone. The formation of the featherlike structures is associated with the nitrogen diffusion in the vicinity of the surface, which is also observed from X-ray diffraction pattern. The orientation of the dendrites is multi-directional because of the non-uniform cooling taking place in the solidification cycle. In the near region of the heat affected zone, martensite is transferred to ferritic–pearlitic structures because of relatively lower cooling rates occurring in this region. Microhardness of the workpiece surface increases considerably after the laser treatment process because of the presence of dense layer, WC carbides, and nitride compounds formed at the surface. The residual stress formed after the laser treatment process is in the order of 300 MPa in the surface region. Temperature gradient remains high in the surface vicinity while progressing time, particularly in the late cooling periods.

6. Conclusion

Laser control melting of dual matrix cast iron surface with presence of 15% WC particles is carried out. A carbon film of about 40 μm and containing uniform distribution of WC is formed at the workpiece prior to the laser scanning. In order to generate self-annealing effect in the laser treated layer, laser beam scanned the surface in spiral motion. Morphological and metallurgical changes in the laser treated layer are examined incorporating, scanning electron microscope, X-ray diffraction, and energy spectroscopy. Microhardness and scratch resistance of the laser treated surface are measured. Temperature and stress fields are predicted using the ABAQUS finite element code. Surface temperature and residual stress predictions are validated with the experimental data. It is found that temperature predictions agree well with the thermocouple data. The residual stress predicted in the surface vicinity also agrees well with that obtained from the X-ray diffraction technique. Laser treated surface is free from the defect sites including voids and cracks despite the micro-stresses formed in the vicinity of the surface, which is also observed from X-ray diffraction pattern. The orientation of the dendrites is multi-directional because of the non-uniform cooling taking place in the solidification cycle. In the near region of the heat affected zone, martensite is transferred to ferritic–pearlitic structures because of relatively lower cooling rates occurring in this region. Microhardness of the workpiece surface increases considerably after the laser treatment process because of the presence of dense layer, WC carbides, and nitride compounds formed at the surface. The residual stress formed after the laser treatment process is in the order of 300 MPa in the surface region. Temperature gradient remains high in the surface vicinity while progressing time, particularly in the late cooling periods.

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References


