Surface coatings for 3-piece freight bogie centre bearings

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Chapter 5: Literature review: Laser cladding

5.1 Introduction

Laser cladding is a surface engineering process by which a laser beam melts wear- and/or corrosion-resistant materials for metallurgical fusion onto a compatible base material [67]. The controlled heat input from the laser ensures a fusion bond with minimal dilution of the clad layer by elements from the steel substrate [68]. The clad materials can be metallic alloys only or consist of ceramic particles bound by a metallic matrix. Popular cladding alloys are based on cobalt, iron and nickel [52]. Ceramic particles such as carbides of tungsten, titanium and silicon, bound by metallic alloys form particle-reinforced metal matrix composites on solidification, providing additional wear resistance [52]. The ceramic particles provide improved wear resistance. Suitable base materials include carbon-manganese and stainless steels, and alloys based on aluminium, titanium, magnesium, nickel and copper [52].

The clad materials can be transferred to the base material by several methods: pre-placed powder on base material, paste injection, powder injection, or wire feeding (Figure 46) [67]. Of these methods, laser cladding by powder injection has been found to be the most effective [67]. The side-fed powder injection method, as illustrated in Figure 46 b2), was used in the laser cladding for this research. A wide range of powder sizes are used in laser cladding, e.g. 2 to 200 μm in diameter [52]. Typical powder injection rates are of the order 25 g/min [52]. Laser clad thicknesses ranging from 0.05 to 2.0 mm and track widths as narrow as 0.4 mm and as broad as 15 mm are possible using the powder blown technique [67, 52].
Figure 46 Different methods of laser cladding. [67].

5.2 Laser material processing

Figure 47 is a schematic of a laser beam traversing a material showing the principal laser material processing variables [52]. Some of these variables will be discussed in the following sections of this chapter.
Figure 47 Laser beam traversing a material showing the principal process variables of laser material processing [52]

5.2.1 Thermal interaction mechanisms

Figure 48 illustrates the three thermal interaction mechanisms that can occur under steady state thermal conditions during laser processing of materials – surface heating, surface melting, and surface vapourisation (keyholing). In laser cladding, the aim is to melt a thin layer of base material together with as much of the alloy addition as possible [52]. As the power density increases the thermal interaction mechanism changes from surface heating to surface melting to surface vapourisation [52]. The power density, E, is defined as the laser beam power divided by the laser beam spot area:

\[ E = \frac{q}{\pi r_B^2}, \quad (23) \]

where \( q \) = laser beam power (W), \( r_B \) = beam radius (mm).

The values of power density required for these different mechanisms are determined by the material thermal properties. For a CO₂ gas laser beam impinging on structural steel,
power densities of the order of 10, 100 and 10 000 W/mm\(^2\) are required for surface heating, surface melting and surface vapourisation, respectively [52].

**Figure 48** The effect of power density on the interaction mechanism between a laser beam and a large block of material under steady state conditions [52].

In reference to Figure 49, Gedda et al. reported that during commercial powder injection laser cladding; 50% of the incident energy (Ql) is reflected off the cladding melt pool (Qr), 10% is reflected off the powder cloud (Qrp), 30% is used to heat the substrate (Qc), and the remaining 10% is used to melt the clad layer (Qp) [69]. Qconvection and Qradiation are not taken into account. Qconvection, the energy lost to shield gas flow, would be negligible. It would be expected that Qradiation would be at least several per cent.
Figure 49 Balance of energy in laser cladding by powder injection [69].

Figure 50 shows the different physical phenomena of the laser cladding zone [67]. The laser beam reaches the substrate and a significant portion of it’s energy (30%) is directly absorbed by the substrate. The powder stream is injected at side (Figure 46 b2)). As discussed above around 10% of the incident energy is absorbed by the powder particles. The energy absorbed by the substrate develops a melt pool [67]. The melted powder particles are simultaneously added into the melt pool. Surface tension gradients drive the Maragoni fluid flow in the melt pool providing rapid mixing of the melted powder particles [67]. In the mushy region, dendrites of the clad alloy begin to form. Due to conduction of heat away from the melt pool, a heat affected zone is always present beneath the solidified clad.
Figure 50 Different physical phenomena of the laser cladding zone [67].

5.2.2 Beam interaction time

The heating time during laser cladding is approximated by the beam interaction time [52], \( t_i \):

\[
    t_i = 2 \frac{r_B}{v_l}
\]

where \( r_B \) = laser beam spot radius (mm), \( v_l \) is the laser beam traverse rate across the sample (mm/s).

If all other laser processing conditions are held constant, then the generalized effects of beam interaction time on the laser cladding processing and thermal conditions, and on the properties of the clad are those listed in Table 9 [70]. In this research only the traverse speed, thus the interaction time, was varied. This table will be referred to in the discussion of Chapter 8.

Please see print copy for Figure 50
Table 9 Effect of beam interaction time [adapted from 70, 69].

<table>
<thead>
<tr>
<th></th>
<th>Short interaction time: $t_i \ll 1$ s</th>
<th>Long interaction time: $t_i \approx 1$ s</th>
</tr>
</thead>
<tbody>
<tr>
<td>Deposition rate (g/mm)</td>
<td>Low</td>
<td>High</td>
</tr>
<tr>
<td>Delivered laser line energy</td>
<td>Low</td>
<td>High</td>
</tr>
<tr>
<td>(J/mm)</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Dilution (%)</td>
<td>Low</td>
<td>High</td>
</tr>
<tr>
<td>Hardness</td>
<td>High</td>
<td>Low</td>
</tr>
<tr>
<td>Average temperature gradient</td>
<td>High</td>
<td>Low</td>
</tr>
<tr>
<td>Cooling rate (K/s)</td>
<td>Fast</td>
<td>Slow</td>
</tr>
<tr>
<td>Microstructure</td>
<td>Fine</td>
<td>Coarse</td>
</tr>
</tbody>
</table>

Figure 51 maps typical values for power density and interaction time for different laser-materials processes and the conditions upon which heating, melting and vapourisation of structural steel occurs. The approximate values used in this research are also marked on the graph as a red bar.

5.3 Laser clad layers

5.3.1 Dilution

A small amount of the substrate must be melted in order to form a strong metallurgical bond. However, molten substrate mixes with the molten clad by convection, effectively diluting the composition of the clad. There are two definitions of dilution based on the methods used for it’s calculation: geometrical and metallurgical.
The geometrical dilution is:

\[
\%\text{dilution} = \frac{b}{(h+b)} \times 100
\]  

(25)

where \(b\) = thickness of substrate melted during laser cladding (mm), and \(h\) = height of clad (mm) as shown in Figure 52.

Figure 51 Power density and interaction time for various laser material processing [67, adapted from 71]. Marked on the graph as a red bar is the approximate range of power density and interaction time values used in this research. NB: The value for power (720 W) equals the incident power (1800 W) multiplied by 0.4 (fraction of incident laser power adsorbed by substrate and clad layer [69]).
The equation for metallurgical dilution [69] as used in this research is:

\[
\%\text{Dilution} = \frac{(M_{Fe} - P_{Fe})}{(S_{Fe} - P_{Fe})} \times 100
\]  
(26)

Where \(M_{Fe}\) = measured concentration of Fe in the clad layer, \(P_{Fe}\) = nominal concentration of Fe in the powder, \(S_{Fe}\) = measured concentration of Fe in the steel substrate.

For a steel substrate, the metallurgical dilution of the clad is calculated using the element iron (Fe) which is the predominant element in steel. This reduces the relative error in the calculation of the dilution [69].

The geometrical dilution provides a bulk value, whereas spot measurements of chemical composition can be made anywhere on the clad layer using an SEM equipped with an energy- and/or wavelength-dispersive spectrometer and, hence the metallurgical dilution can be calculated effectively at any spot on the clad layer.

Dilution of about 5% is sought in laser cladding [52]. Yellup [51] observed a substantial decrease in hardness of stellite 6 laser clad layers above 8% dilution. Dilution increases with increasing laser power but decreases with increasing travel speed [67].
5.3.2 Stellite 6 microstructure

Stellite 6 is a cobalt-based alloy. The chemical composition of the Stellite 6 used in this research is listed in Table 21 (page 155). The solidification microstructure of Stellite 6 comprises mostly primary dendrites of the cobalt rich phase (γ, face-centred cubic) and interdendritic eutectic [72]. The eutectic phase includes $M_7C_3$ carbides, mainly $Cr_7C_3$ carbides [73].

A unique aspect of laser cladding with respect to weld surfacing techniques is the possibility of obtaining very fine microstructures resulting from elevated solidification rates [72]. Figure 53 shows the effect of laser traverse speed on the solidified Stellite 6 microstructure [72]. In relative terms, at low speed, the microstructure is coarse, whilst at fast speed it is fine. Figure 54 shows the secondary arm spacing as a function of cooling rate. The open circles and thick bars represent experimental values, the latter being for laser cladding at (from left to right) 1.67, 13.3, and 166.7 mm/s representing slow, intermediate, and fast laser cladding traverse speeds, respectively.

5.3.3 Hardness profile of Stellite 6 laser clad layers

There are a number of contributing factors to the hardness of laser clad Stellite 6: solid solution strengthening, and precipitation strengthening. Chromium, tungsten and molybdenum provide solid solution strengthening [73]. Chromium is the predominant carbide precipitate former. The predominant carbide found in Stellite 6 is $Cr_7C_3$ [73].

Figure 55 shows a microhardness profile of laser clad Stellite 6 [51]. The average microhardness is around 550 HV. The drop in hardness from 550 HV to 430 HV at 0.05 mm from the interface is likely due to excessive dilution.
Figure 53 Transverse section (perpendicular to the scanning direction) of the Stellite 6 deposits. The dark dendritic phase is the cobalt-based matrix and the white phase is formed by the $\text{M}_7\text{C}_3$ eutectic carbides: (a) slow processing conditions (=1.67 mm/s), (b) fast processing conditions (=167 mm/s) [72].

Figure 54 Secondary arm spacing as a function of cooling rate. The open circles and thick bars represent experimental values [72].
Figure 55 Hardness profile of Stellite 6 [51].
5.3.4 Wear resistance of Stellite 6

There was no literature regarding the dry sliding friction and wear testing of Hadfield steel pins against Stellite 6. The good wear properties of Stellite 6 can be described based on the unusual deformation characteristics imparted by cobalt [74]. The cobalt matrix has a low stacking fault energy, and it’s face centred cubic structure (high temperature stable) can transform to the low temperature stable hexagonal close-packed structure under the action of mechanical stress [74].

The low stacking fault energy is believed to delay fatigue crack initiation through it’s influence on slip behaviour and twinning propensity [74]. The only important slip systems in face centred cubic metals are those associated with slip on the octahedral planes [75]. There are four octahedral planes, each with three close-packed directions providing 12 slip systems being well-distributed in the crystal lattice [75]. It is therefore almost impossible to strain a face-centred cubic metal and not have at least one octahedral plane in a favourable position for slip [75]. Single crystals of face-centred cubic metals exhibit three stages of work-hardening as shown in Figure 56 [76]: Stage I (easy glide), stage II (linear hardening region), and stage III (parabolic hardening region). The work hardening rate in face-centred cubic polycrystals, i.e. for Stellite 6, is about 9 times the rate for single crystals [76]. Stage I immediately follows the yield point and has low rate of work-hardening. Stage II shows a rapid increase in work-hardening rate. Stage III exhibits a low rate of work-hardening and is probably associated with the annihilation of built-up dislocations, that were causing stress to increase more rapidly, as a consequence of cross-slip [76]. In face-centred cubic metals with low stacking fault energy cross-slip is difficult at reasonable stress levels, so the transition from stage II to stage III work hardening is delayed [76]. The stress necessary
to produce twinning in face-centred cubic metals is high, thus face-centred cubic metals do not normally deform by twinning [76]. Solid solution alloying, e.g. for Stellite 6 this is chromium in cobalt, favours twinning [76]. It has also been shown that in copper-based alloys (face-centred cubic) that the twinning stress decreases with stacking fault energy [76]. So, hypothetically for a Stellite 6 surface that has work-hardened such as in stage II, the critical resolved shear stress for twinning may be lower than for slip allowing twinning to proceed, thus delaying fracture. This could explain the improvement in fatigue properties.

Figure 56 Stress-strain curve showing the three stages of work-hardening in face-centred cubic metals [76].

The transformation from face centred cubic to hexagonal close-packed structure is believed to enhance fatigue life by reducing internal strain [74]. Whether the face-centred cubic to hexagonal close-packed transformation proceeds is complicated and
depends on a number of factors such as, the stacking fault energy, the cooling rate, the grain size or carbide spacing in the matrix, and the plastic strain induced after cooling [73]. It is beyond the scope of this thesis to discuss this further.

Figure 57 shows the surface damage due to galling for various non-ferrous alloys [46]. The test used was the pin-on-block galling test. The ASTM standard for galling resistance of materials G98-91 defines galling as: “a form of surface damage arising between sliding solids, distinguished by macroscopic, usually localized, roughening and creation of protrusions above the original surface; it often includes plastic flow or material transfer, or both”, and “A wavy surface is not considered galled. If fracture of any cold welded areas has taken place in the plane of the surfaces and no distinct raised metal is found, galling has not occurred” [77]. The average contact pressure and sliding speed for the data presented in Figure 57 were not given, but a normal range for galling test conditions could include: average contact pressure $\approx 5$ to $150$ MPa, sliding speed $\approx 2.5$ to $30$ mm/s. These are similar to the conditions at the centre bearing and wear testing procedure used in this research. However, the galling test is short-term and the cylindrical flat face pin is rotated against a flat block. However, it could provide a guideline.

Stellite 6 is similar in chemical composition and properties to ERCoCr-A. The 2 layer ERCoCr-A coating was welded gas tungsten arc (GTA). The surface damage in the self-mated condition was an order of magnitude less than when slid against austentic stainless steels 304 and 316 [46]. Bearing in mind the Hadfield steel used in the wear testing in this research also has the austenitic structure. It was beyond the scope of this research to study dry sliding friction and wear of self-mated Stellite 6.
Figure 57 Galling property evaluation of non-ferrous alloys. Stellite 6 is similar in chemical composition and properties to ERCoCr-A. In this case, ERCoCr-A has been gas tungsten arc (GTA) hardfaced in 2 layers [46].