

# 多激光粉床熔融成形 GH3536 合金搭接区域组织与性能特征研究

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**摘要** 利用多激光粉床熔融(ML-PBF)设备成形了GH3536高温合金,系统研究了单激光区、双激光区以及四激光区成形试样的缺陷、显微组织和拉伸性能的变化规律。结果表明:随着激光束的增多,试样表面质量得到改善,但致密度由99.82%(单激光区)降低到92.35%(双激光区)和98.97%(四激光区)。同时,多激光搭接区的重熔导致部分晶粒再结晶,沿(001)的择优生长取向更加显著,织构指数也由单激光成形试样的3.040分别提升到了3.403和3.465。但是,单激光区、双激光区和四激光区的大角度晶界比例逐渐降低,由65.9%分别降低到50.1%和46.3%。此外,搭接区的残余应力呈现降低趋势,单激光区、双激光区和四激光区分别为192.3 MPa,106.5 MPa和44.1 MPa。三者的室温拉伸强度相差较小,均达到了800 MPa以上,断裂延伸率由30.3%(单激光区)降低到25.9%(双激光区)和25.4%(四激光区),这主要是搭接区内的孔洞和裂纹缺陷造成的。研究结果有望为多激光粉床熔融制备大尺寸镍基高温合金零部件提供有效的参考。

**关键词** 激光技术; 多激光粉床熔融; 激光选区熔化; 镍基高温合金; 显微组织; 力学性能

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## 1 引言

作为增材制造(AM)技术之一,激光粉床熔融(L-PBF)技术也称为激光选区熔化(SLM)技术,被认为是21世纪最具潜力的金属激光制造技术<sup>[1-3]</sup>。它通过高能激光束选择性地熔化金属粉末层,能够直接制造出近乎全致密、高精度、形状复杂的金属零件<sup>[4-6]</sup>。该技术已成功应用于铝合金、铁合金、钛合金和镍基高温合金上,在生物医学、航空航天以及冶金等领域中具有非常广阔的应用前景<sup>[7-9]</sup>。然而,传统的L-PBF技术由于成形尺寸和工作效率的限制,无法实现大尺寸零件的整体成形,例如涡轮机叶片、涡轮盘等绝大多数难加工的复杂构件的尺寸均超过L-PBF设备的成形范围(250 mm×250 mm),因此该技术在航空航天领域的进一步应用受到了限制<sup>[10]</sup>。

多激光粉床熔融(ML-PBF)技术通过多个激光器和振镜系统的同时工作,分别加工不同的区域,在保证高制造精度的同时还具有高效率、一体成形大尺寸结构的优点,更适合制造结构复杂的大尺寸构件<sup>[11-12]</sup>。Wiesner等<sup>[13]</sup>通过ML-PBF技术成形了大型气缸盖

(零件尺寸为494 mm×210 mm×143 mm,体积超过5 L),减少了50%的成形时间,证明了ML-PBF在成形大尺寸构件时具有巨大优势。虽然ML-PBF设备在原理上与传统的单激光L-PBF设备没有明显差别,但为了保证每个单激光加工区域之间具有良好的连接质量,多激光扫描时都会保证一定宽度的重叠区域。而重叠区域会受到激光束的多次扫描,试样在致密度、微观组织和性能上与单激光成形试样还存在差距。Taheri等<sup>[14]</sup>发现,双激光粉床熔融成形时会产生溅射,从而形成较大的反冲压力,造成未熔合缺陷,并降低成形试样的拉伸性能。Zhang等<sup>[15]</sup>通过自制的四激光ML-PBF设备成形了近乎致密的AlSi10Mg合金,发现激光重叠区域的致密度和晶粒尺寸比单激光成形区域的略低。Wei等<sup>[16]</sup>发现,在ML-PBF成形过程中,多激光同时作用于重叠区域会产生较多的匙孔缺陷。因此,关于不同激光束的ML-PBF成形试样的组织与性能及其影响因素还需要进一步研究。

本文研究的GH3536合金(对应国外Hastelloy X合金<sup>[17]</sup>)是一种铁含量高、由铬和钼元素固溶强化的镍基高温合金,能在900℃的高温环境下长时间工作,在

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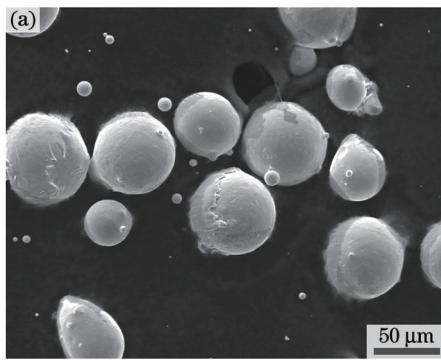
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1080 °C下也能短时使用,适用于航空航天发动机的燃烧室及其他高温部件<sup>[18-19]</sup>。L-PBF技术在成形复杂整体构件方面具有潜在优势,目前已有L-PBF成形GH3536合金的大量报道,但针对ML-PBF成形GH3536的研究还比较少。因此,本文采用ML-PBF技术成形了GH3536合金,研究了不同激光束作用下ML-PBF成形GH3536合金的致密度、缺陷、显微组织、残余应力和拉伸性能的变化规律,为多激光粉床熔融成形高性能复杂结构大尺寸GH3536零部件提供了参考。

## 2 试验材料及装备

### 2.1 试验材料

试验材料选用宁波众远新材料科技有限公司制备



的气雾化GH3536粉末,其化学成分如表1所示。图1是粉末微观形貌以及粒径分布,粉末为球形或近球形,颗粒球形度良好,采用激光粒度仪测得粉末粒径范围为21.2~58.9 μm,中值粒径为35.3 μm,满足L-PBF成形试验的要求。试验前将粉末置于90 °C恒温环境中干燥12 h,以去除水分,提高流动性。基板采用316L不锈钢材质,试验前用乙醇充分清理表面。

表1 GH3536粉末的化学成分

Table 1 Chemical compositions of GH3536 powder

Element	C	Cr	Co	W	Mo	Fe	Ni
Mass fraction / %	0.09	20.43	2.28	0.70	8.65	16.61	Bal.

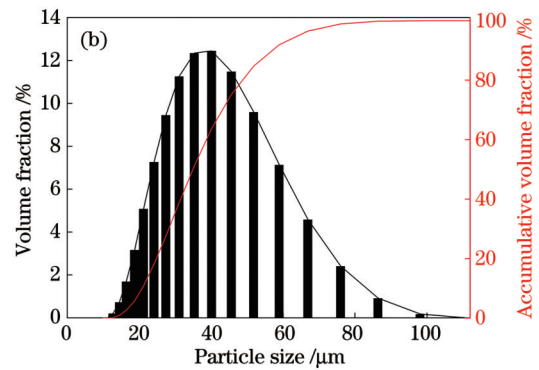


图1 GH3536合金粉末特征。(a)粉末形貌;(b)粉末粒径分布

Fig. 1 Characteristics of GH3536 alloy powder. (a) Powder morphology; (b) powder particle size distribution

### 2.2 试验方法与设备

采用四激光大尺寸增材制造设备进行多激光粉床熔融成形试验。设备配置4个500 W单模连续光纤激光器,波长为1.07 μm,焦斑直径为50 μm,焦距为

130 mm。为保证ML-PBF成形过程的一致性,各光学系统采用相同品牌、型号、生产批次的元器件。该设备允许4个激光器一起工作,成形区域尺寸为600 mm×600 mm,如图2(a)所示。

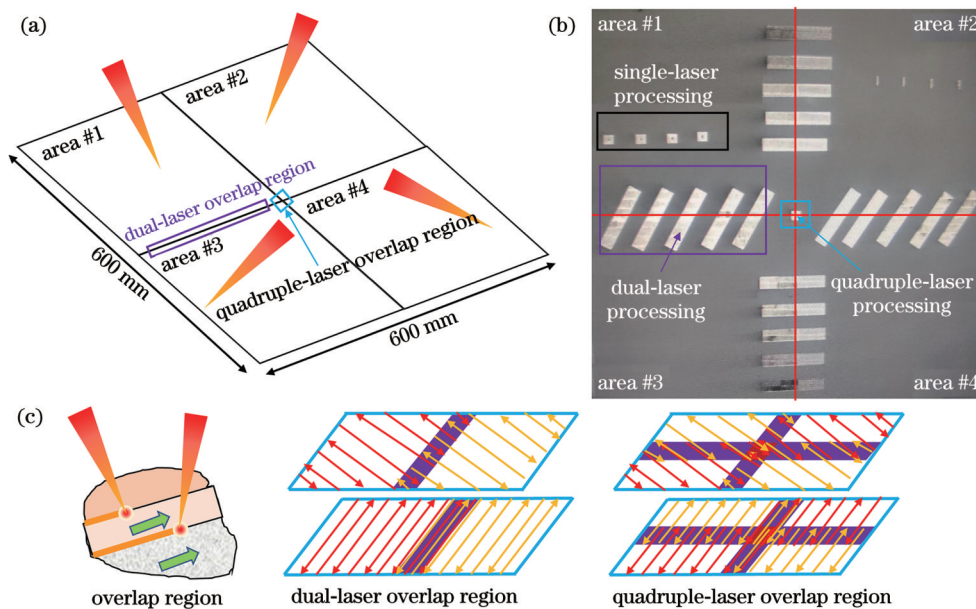


图2 试样成形示意图。(a)四激光L-PBF成形示意图;(b)L-PBF成形试样;(c)单激光成形区域以及搭接区域示意图

Fig. 2 Schematics of sample forming. (a) Diagram of quadruple-laser L-PBF forming; (b) L-PBF forming samples; (c) schematics of single-laser processing region and overlap region



试验在室温环境以及氩气保护下进行,氧气含量(体积分数)控制在  $200 \times 10^{-6}$  以下。采用课题组前期试验优化后的工艺参数<sup>[20]</sup>:激光功率为 240 W,扫描速度为 600 mm/s,铺粉厚度为 40  $\mu\text{m}$ ,扫描间距为 100  $\mu\text{m}$ ,扫描方式为倾斜分区,相邻层间扫描夹角为  $67^\circ$ 。分别在单激光、双激光以及四激光作用区域成形了块体和长条试样,用于后续的微观组织表征以及力学性能测试,如图 2(b)所示。四个激光束分别加工四个不同成形区域,为保证不同激光束成形区域之间良好结合,区域边界都会存在 1~3 mm 的多激光搭接区域,如图 2(c)所示。在该区域内,激光束分时进行扫描,即产生激光重熔。后续对多激光成形试样微观组织及残余应力的表征都取自搭接区域。

### 2.3 性能检测与表征

采用电火花线将成形试样切割成  $8 \text{ mm} \times 8 \text{ mm} \times 5 \text{ mm}$  方块。采用阿基米德排水法测试试样的致密度,每组试样取三个样品测试,结果取平均值。通过 Image J 软件分析试样的孔隙率,每个样品取 5 张光镜图进行统计,结果取平均值。试样经 A2 电解液(含体积分数为 10% 的  $\text{HClO}_4$  和体积分数为 90% 的  $\text{C}_2\text{H}_5\text{OH}$ )电解腐蚀后,利用光学显微镜观察裂纹情况,利用扫描电镜(SEM)观察显微组织,同时采用场发射扫描电镜进行电子背散射衍射(EBSD)分析。使用 X 射线衍射试验机进行残余应力测试。拉伸试样按照国际拉伸件标准加工制备,使用材料高

温持久性能试验机进行室温拉伸试验,拉伸速度为 1 mm/min,测试中未加入引伸计,通过试样上的标距计算断后伸长率;同种状态下取 5 个拉伸件进行测试,测试结果取平均值。

## 3 结果与讨论

### 3.1 微观形貌

图 3 为不同激光束成形试样的上表面熔化道形貌,能观察到明显的条状熔化道特征,这是金属粉末在高速移动的激光束作用下快速熔化并凝固而形成的。从单激光成形 GH3536 试样的上表面[图 3(a)]可以看到大量不规则颗粒以及球状凸起,表面质量较差。经过激光重熔后,双激光成形试样表面的不规则颗粒和球状凸起减少,且熔化道变得更加平整光滑,粗糙度降低[图 3(b)]。四激光成形试样表面的不规则颗粒和球状凸起进一步减少,熔化道宽度变得更窄,表面质量得到改善,如图 3(c)所示。在 ML-PBF 成形时,激光作用在已成形的块体上,会造成试样重熔并凝固,从而熔化道宽度减小。同时,熔化的液态金属受到表面张力的作用,表面形状会发生向球形转变的趋势,进而发生球化现象,形成表面不规则颗粒。而 ML-PBF 成形时的激光重熔过程会进一步熔化上表面黏附的未熔粉末并将粉层穿透熔化,从而有助于消除球化效应以及未熔粉末产生的表面凸起,并且激光束越多,表面质量的改善效果越好<sup>[21-23]</sup>。

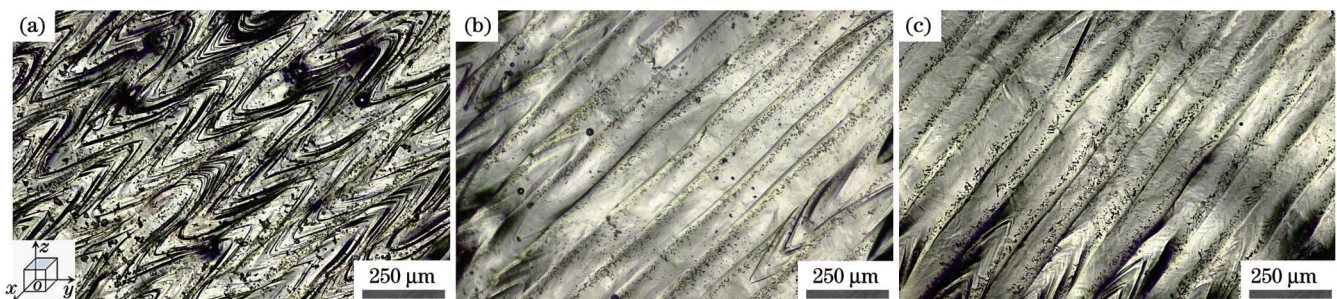


图 3 试样上表面熔化道光镜图。(a)单激光成形区域;(b)双激光成形区域;(c)四激光成形区域

Fig. 3 Optical micrographs of melting tracks on sample upper surface. (a) Single-laser processing region; (b) dual-laser processing region; (c) quadruple-laser processing region

不同激光束成形试样的致密度对比如图 4 所示。在优化后的成形参数下,无论是显微图像分析还是阿基米德法测量,单激光成形样品的致密度都达到了 99% 以上。在图 4 所示的三组数据中,两种测试方法得出的致密度差值都在 1% 以内,这是由于显微图像分析依赖于有限的检测区域,并且阿基米德法测量结果的准确度要优于显微图像分析<sup>[24]</sup>。而双激光和四激光成形试样的孔隙率提高了,并且随着激光束的增加,致密度的衰减趋势更加明显。其中四激光成形试样的致密度最低,仅为 98.97%。从图 4 可以看出,多激光成形试样的主要缺陷为球形孔洞,而在单激光成形区域,样品中的孔洞及裂纹缺陷相对较少。当采用 ML-PBF

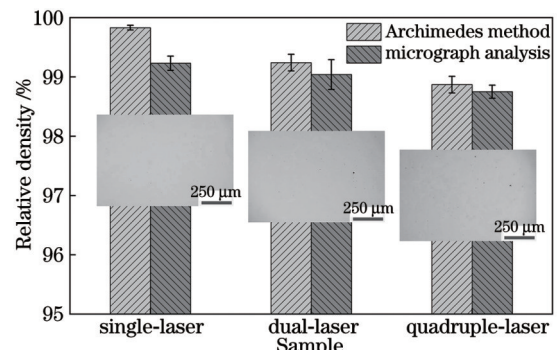


图 4 不同试样的相对致密度及光镜图

Fig. 4 Relative densities and optical micrographs of different samples



成形试样时,在搭接区域试样经历了激光重熔过程,重叠区域的能量输入过高,引起严重的 Marangoni 对流,导致金属在熔化时不能完全流动,并在凝固后形成孔洞缺陷<sup>[25-26]</sup>。此外,当多激光同时作用时,极大的能量输入导致熔池回流并且填充滞后,金属蒸气不能完全溢出,导致孔洞的形成,同时气化金属对熔池的反冲压力会引起飞溅,也会形成孔洞<sup>[27]</sup>。随着激光束的增多,诱导产生的孔洞增多,恶化了成形试样的致密化程度。

不同激光束成形试样的水平方向的显微组织形貌

如图 5 所示。从图 5(a)~(c)所示的低倍形貌图中可以观察到明显的熔池形貌,条状熔池交替叠加。这是由于采用了 67° 分区的激光扫描策略,每一层熔池与上一层存在一定的角度。单激光粉末床熔融成形样品的熔池宽度在 100  $\mu\text{m}$  左右,这与试样采用的 100  $\mu\text{m}$  扫描间距相匹配。而采用多激光成形的试样的熔池宽度变窄,并且随着激光束的增多,熔池形貌的变化增大。这是由于多激光重熔过程会促使相邻熔化道和相邻层之间具有更好的冶金结合,从而熔池之间的搭接率得到提高<sup>[28]</sup>。

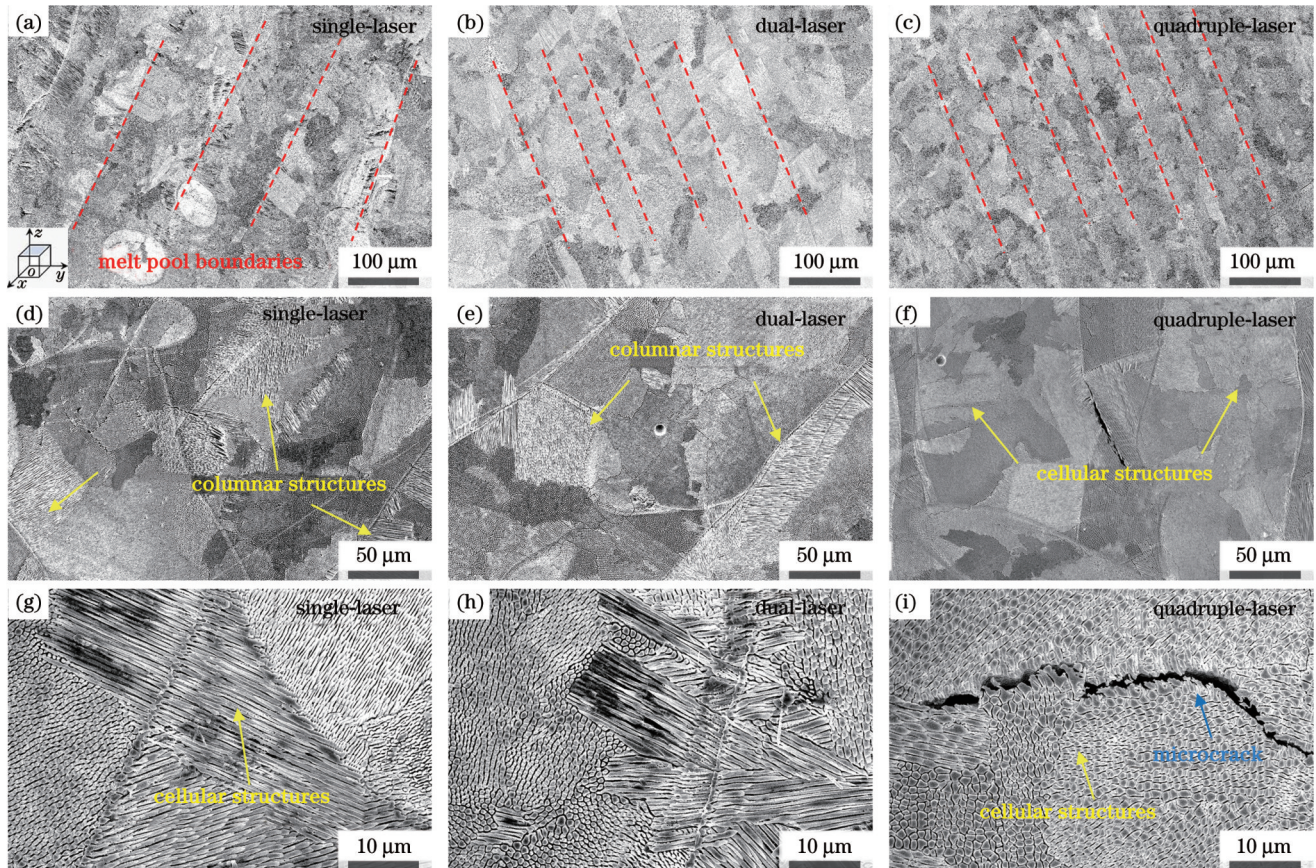


图 5 不同试样的 SEM 图。(a)~(c)低倍;(d)~(i)高倍

Fig. 5 SEM images of different samples. (a)~(c) Low magnification; (d)~(i) high magnification

图 5(d)~(i)是 GH3536 试样的高倍形貌图,熔池内部由近似垂直于熔池边界的逆温度梯度方向生长的细小柱状晶和等轴晶组成,外延生长的柱状晶可以穿过多个熔化道,这些细小晶粒是 L-PBF 成形过程中产生的极大温度梯度和冷却速度造成的。并且随着激光束的增加,晶粒尺寸没有明显差异,但等轴晶的数量变多[图 5(d)~(f)]。在 ML-PBF 成形时,激光重熔区域的晶粒是在已凝固试样上生长的,阻碍了柱状晶的生长,随着激光束的增多,大量柱状晶逐渐转变为等轴晶。此外,如图 5(f)、(i)所示,在四激光成形区域中还存在微裂纹缺陷,裂纹附近能观察到明显的晶粒撕裂特征。成形试样熔池中心散热方向与成形方向相近,凝固生长速度大于熔池边缘;当竞争生长的中心晶粒

与边缘柱状晶接触时,晶粒会停止生长,接触区域的液态金属为凝固的末端,在其固液界面前沿的残余液相中会富集一些难熔元素,最终形成低熔点液膜。当熔融金属的凝固收缩以及 L-PBF 成形过程中的高温梯度所产生的拉应力超过液膜的临界拉断力时,裂纹便会在晶界处产生并沿着晶界扩展<sup>[29-30]</sup>。

### 3.2 晶粒特征

图 6 为单激光、双激光和四激光成形 GH3536 试样沿水平面方向的 EBSD 结果。单激光成形试样的晶粒取向比较均匀,随着激光束的增多,搭接区域的晶粒变细小,沿(001)方向的强度逐渐变强,而沿(101)和(111)两方向的强度相对变弱,这与面心立方(FCC)型结构材料的晶粒择优生长取向一致<sup>[29-31]</sup>。这主要是



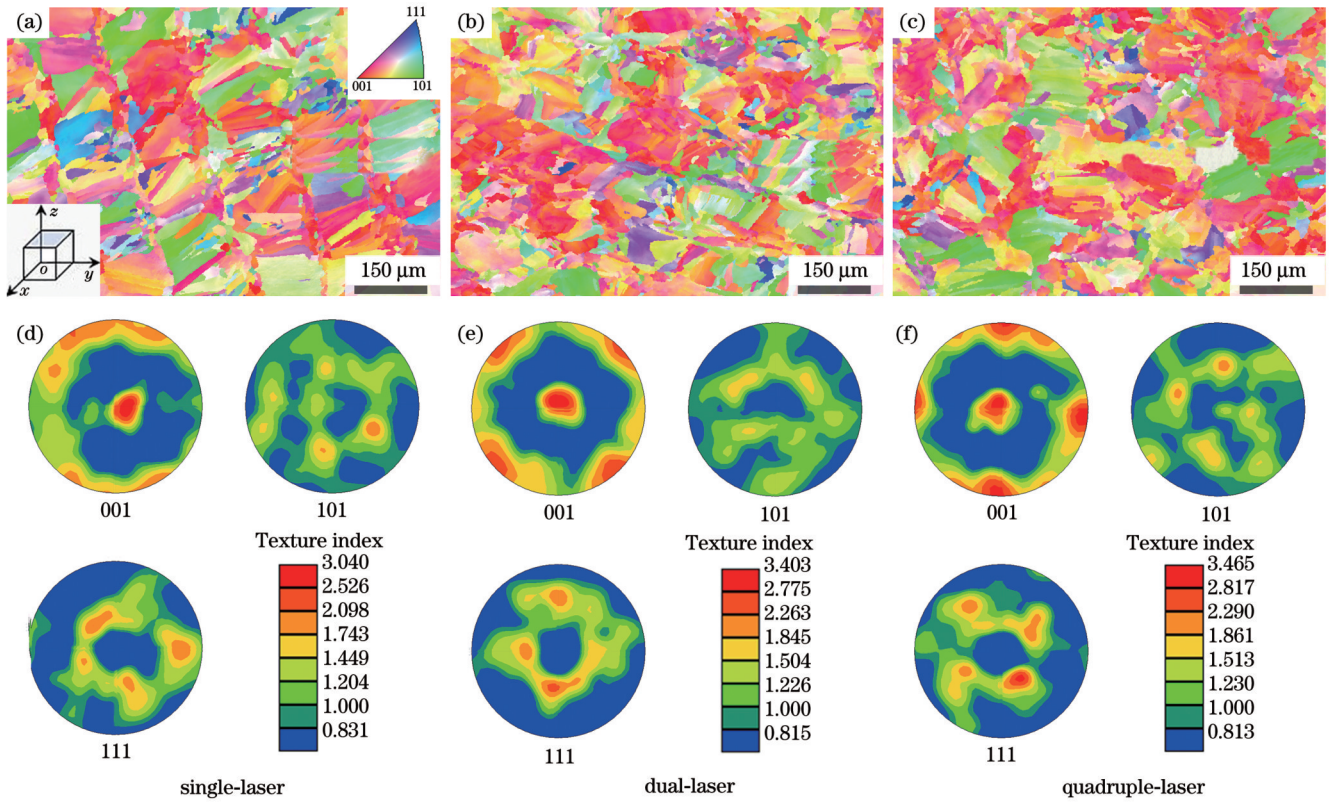


图 6 不同激光束成形试样沿水平方向的EBSD结果。(a)~(c)晶粒取向;(d)~(f)织构

Fig. 6 EBSD results of samples formed by different laser beams along horizontal direction. (a)~(c) Grain orientation; (d)~(f) texture

由于在 ML-PBF 成形过程中,晶粒经历重熔过程,发生了部分再结晶,因此晶粒的择优生长更加显著。图 6(d)~(f)为不同试样的织构指数的计算结果,可以看出,单激光成形试样在(001)方向的织构指数为 3.040,而双激光以及四激光成形试样在(001)方向的织构指数分别达到了 3.403 和 3.465,说明随着激光束的增多,试样的织构指数增大。织构指数是反映晶体织构程度的重要指标,织构指数为 1 时说明试样为各向同性材料,织构指数越大,试样的各向异性程度越大<sup>[32]</sup>。因此,在多激光作用下,试样的各向异性程度升高了。由前文可知,在多激光成形试样中,水平面的柱状晶转变为等轴晶,意味着垂直方向上的晶粒主要为垂直生长的柱状晶粒,这种方向性的微观组织导致材

料表现出更加明显的各向异性。

图 7 为不同激光束成形试样的晶间取向偏差角分布,可以看出,随着入射激光束的增加,大角度晶界(HAGBs)的比例逐渐降低,从单激光成形试样的 65.9% 逐渐降低到双激光成形试样的 50.1% 以及四激光成形试样的 46.3%。在 L-PBF 的快速熔化和凝固过程中,大量亚晶粒形核并长大,这些亚晶粒主要以小角度晶界(LAGBs)的形式出现<sup>[33]</sup>。由于 L-PBF 成形过程中的逐层熔化堆积作用,这些亚晶粒发生部分再结晶,因此小角度晶界向大角度晶界转变<sup>[34]</sup>。由于 L-PBF 过程中的极高冷却速度,小角度晶界的初始含量很高,并且不是所有的小角度晶界都转变为大角度晶界,在成形试样中还保留着一定量的小角度晶界。

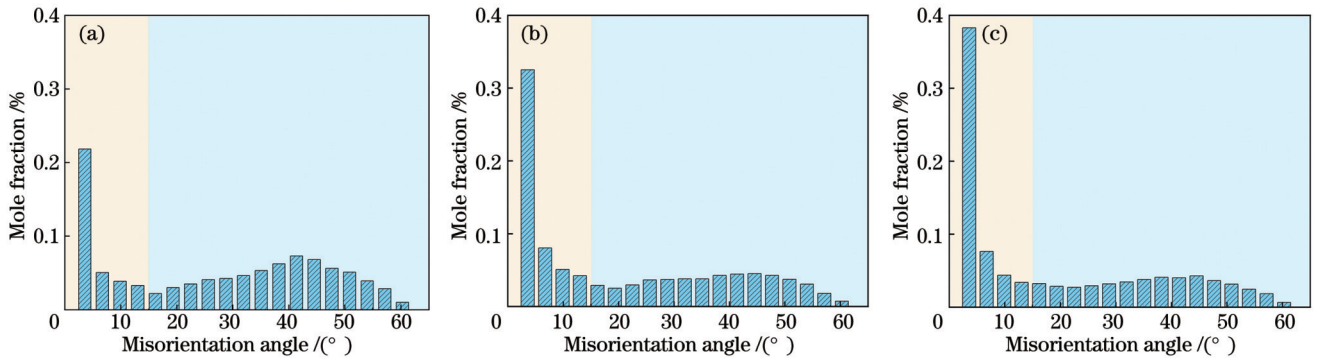


图 7 不同试样的取向偏差角分布。(a)单激光成形试样;(b)双激光成形试样;(c)四激光成形试样

Fig. 7 Misorientation angle distributions of different samples. (a) Single-laser processing sample; (b) dual-laser processing sample; (c) quadruple-laser processing sample



在 ML-PBF 成形过程中,样品受到多次加热而熔化,导致大量亚晶粒形核与长大。同时,与单激光成形时激光束作用在粉末层上不同,多激光成形时激光束直接作用在上一层已凝固的试样上,由于凝固层固体的导热系数比粉末床的导热系数高<sup>[35]</sup>,因此凝固热量能够迅速散失,减轻了小角度晶界向大角度晶界的转变程度。同时,多激光重熔会积累收缩应力,由于过饱和固溶体的高晶格畸变和溶解度,位错会逐渐累积在枝晶间区域,晶粒内部发生轻微的倾斜,从而形成了小角

度晶界<sup>[36]</sup>。

### 3.3 残余应力

利用 X 射线衍射 (XRD) 方法测试了不同激光束成形试样的残余应力,图 8(a)~(c) 为不同试样在同一个晶面不同倾转角度下的 XRD 图谱。根据前期得到的 L-PBF 成形 GH3536 试样的 XRD 测试结果<sup>[17]</sup>,选择衍射角度在 70°~80° 之间的 (220) 衍射峰进行分析。通过拟合图中不同角度下的 X 射线数据,可以对残余应力进行计算分析<sup>[37]</sup>。

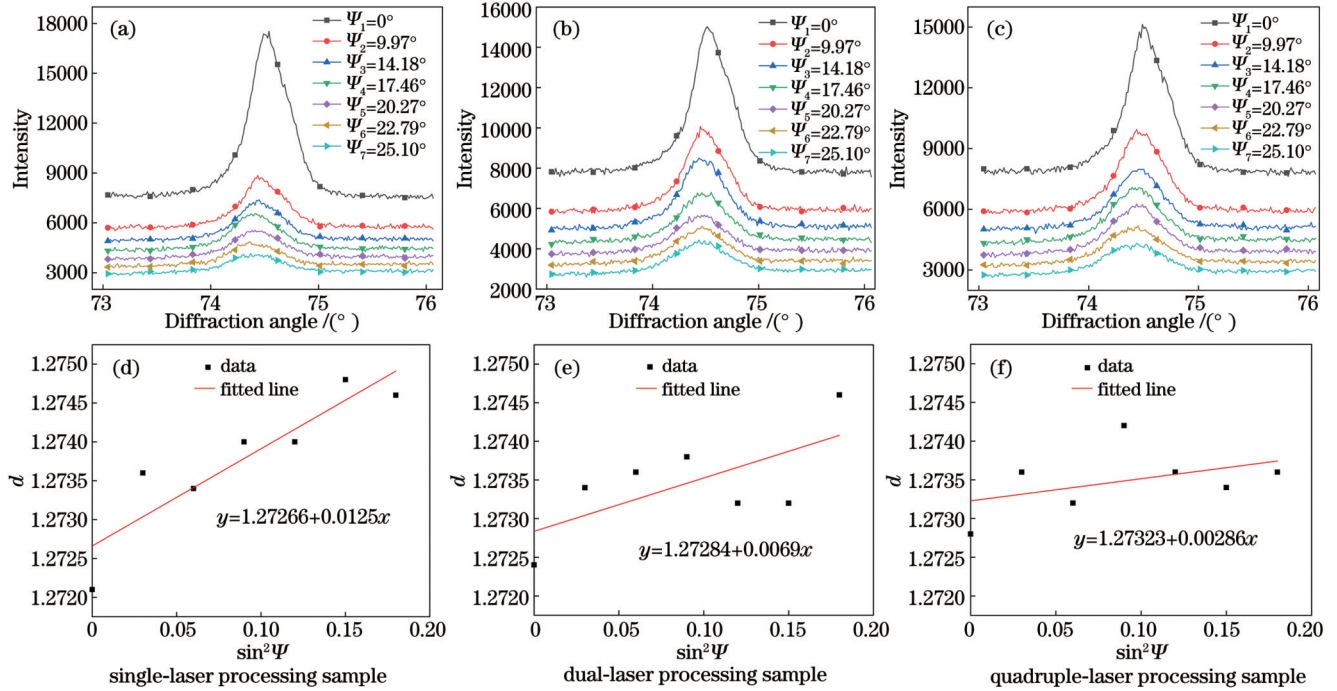


图 8 不同激光束成形试样的残余应力。(a)~(c) (220) 晶面在不同倾转角度下的 XRD 图谱;(d)~(f)  $d$  随  $\sin^2\Psi$  的变化

Fig. 8 Residual stress of samples formed by different laser beams. (a)~(c) XRD patterns of (220) crystal planes at different tilting angles; (d)~(f)  $d$  versus  $\sin^2\Psi$

当扫描晶面的法向和应力方向存在倾转  $\Psi$  角度时,晶格应变 ( $\epsilon_\Psi$ ) 为

$$\epsilon_\Psi = \frac{d_\Psi - d_0}{d_0} = \epsilon_\beta \sin^2\Psi + \epsilon_n \cos^2\Psi, \quad (1)$$

考虑其他角度的晶格应变与倾转角  $\Psi=0^\circ$  时的应变,有

$$\epsilon_\Psi - \epsilon_n = (\epsilon_\beta - \epsilon_n) \sin^2\Psi = \frac{d_\Psi - d_n}{d_0} \approx \frac{d_\Psi - d_n}{d_n} \approx \frac{1 + \nu}{E} \epsilon_\beta \sin^2\Psi, \quad (2)$$

式中:  $\Psi$  为晶面法线与样品表面法线之间的夹角;  $\epsilon_\beta$  为扫描晶面的法向与应力方向垂直时的晶格应变;  $d_\Psi$  为存在倾转角时的晶面间距;  $\epsilon_n$  为扫描晶面的法向与应力方向平行时的晶格应变;  $d_n$  为扫描晶面的法向与应力方向平行时的晶面间距;  $d_0$  为无应力时的晶面间距;  $E$  和  $\nu$  分别为样品的杨氏模量和泊松比。其中,倾转角度分别为  $\Psi_1=0^\circ$ ,  $\Psi_2=9.97^\circ$ ,  $\Psi_3=14.18^\circ$ ,  $\Psi_4=17.46^\circ$ ,  $\Psi_5=20.27^\circ$ ,  $\Psi_6=22.79^\circ$ ,  $\Psi_7=25.10^\circ$ 。拟合

晶面间距 ( $d$ ) 随  $\sin^2\Psi$  的变化可以得到残余应力值,图 8(d)~(f) 分别为单激光、双激光以及四激光成形试样的数据点及其拟合曲线。可以看出不同条件下曲线斜率的变化规律,这表明残余应力水平从大到小的试样依次为单激光成形试样、双激光成形试样、四激光成形试样。选择 GH3536 合金的杨氏模量和泊松比分别为 200 GPa 和 0.3<sup>[38]</sup>, 计算得出的残余应力值分别为 192.3、106.5、44.1 MPa。由前文可知,多激光成形试样经历了激光重熔过程,试样受到多次的激光高温作用。残余应力是材料内部自身保持平衡的一种内应力,其分布状态与材料屈服强度相关。高温条件下材料的屈服强度较低,当熔覆层内部的应力超过该温度下自身的屈服强度时,会发生塑性变形而释放应力;当应力小于该温度下材料的屈服强度时,也会产生局部弛豫,同样能达到降低应力的效果<sup>[39]</sup>。

### 3.4 拉伸性能

图 9 为不同激光束成形 GH3536 试样沿水平方向的室温拉伸试验结果。可以看出,试样的抗拉强度



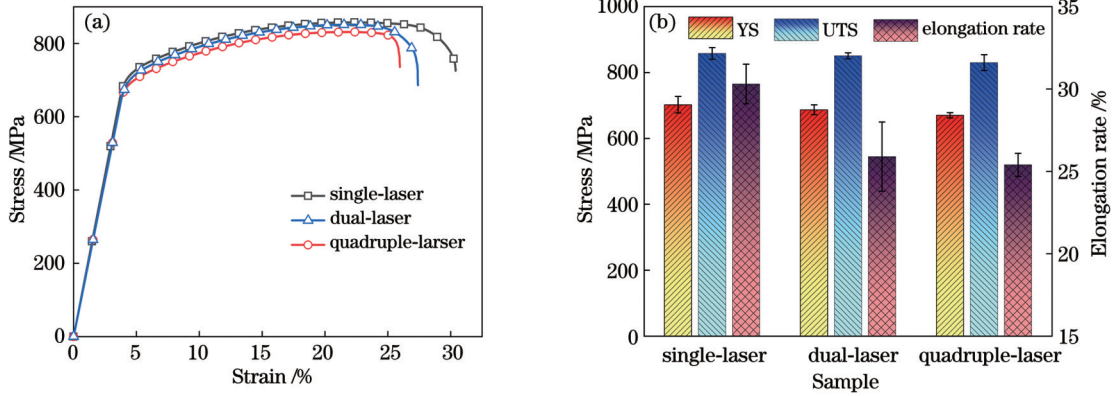


图 9 不同试样的拉伸性能。(a)水平方向的应力-应变曲线;(b)水平方向的屈服强度、抗拉强度及延伸率

Fig. 9 Tensile properties of different samples. (a) Stress-strain curves along horizontal direction; (b) yield strength, ultimate tensile strength and elongation rate along horizontal direction

(UTS)均超过了 800 MPa, 远高于锻造态<sup>[40]</sup>, 这主要归功于 L-PBF 成形试样致密的奥氏体组织以及更加细小的晶粒。其中, 单激光、双激光以及四激光成形试样的抗拉强度分别为  $(858.1 \pm 18.9)$  MPa、 $(851.4 \pm 9.2)$  MPa 和  $(830.5 \pm 24.4)$  MPa, 屈服强度 (YS) 分别为  $(702.9 \pm 25.4)$  MPa、 $(687.6 \pm 15.1)$  MPa 和  $(670.9 \pm 8.7)$  MPa, 可以看出多激光成形时试样的强度都略有下降, 但变化不显著。此外, 多激光成形试样的延伸率[双激光成形试样的  $(25.9 \pm 2.1)\%$  和四激光成形试样的  $(25.4 \pm 0.7)\%$ ] 小于单激光成形试样的延伸率  $[(30.3 \pm 1.2)\%]$ , 前者仅为

后者的 83% 左右。这意味着多激光重熔产生的孔洞以及微裂纹缺陷不仅会减小试样的致密度, 还会降低试样的韧性。在拉伸过程中, 试样中的孔洞和微裂纹是应力集中部位, 拉应力载荷会加速裂纹的扩展, 导致过早断裂, 从而减小了成形试样的延伸率。

#### 4 分析与讨论

单激光和多激光成形 GH3536 合金的熔池和缺陷形成机理如图 10 所示。多激光成形 GH3536 试样的主要缺陷为孔洞缺陷, 由于成形时采用的是优化后的

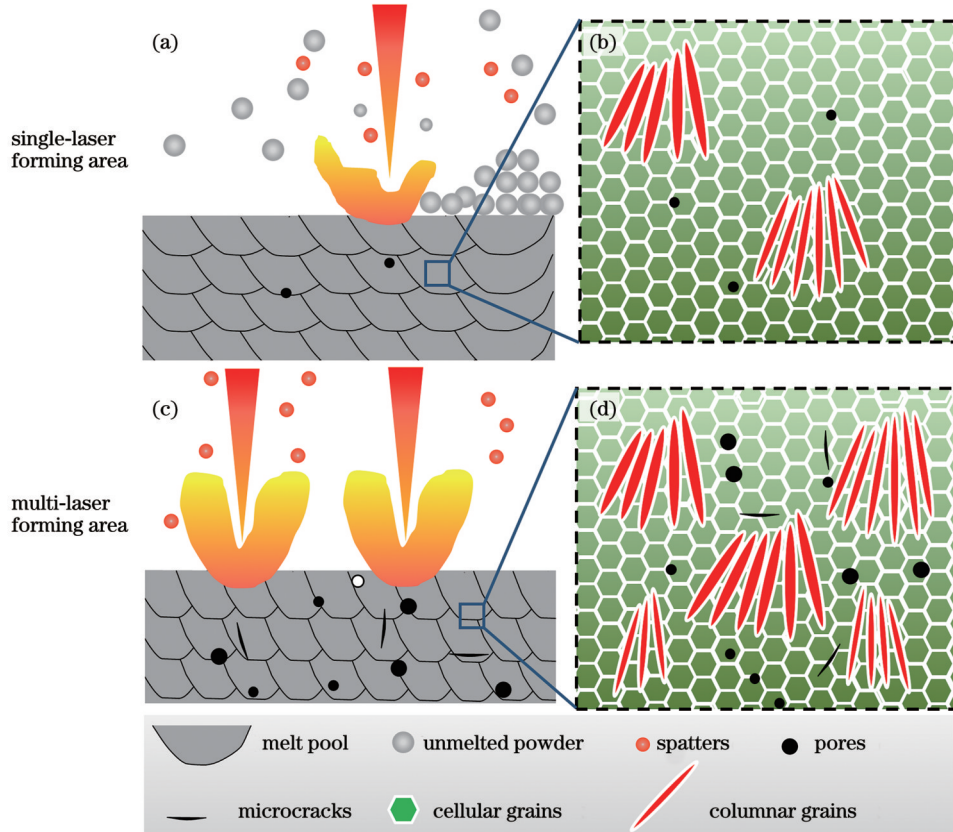


图 10 单激光成形区域以及多激光成形区域的差别。(a)(c)激光作用区域及熔池;(b)(d)晶粒特征及缺陷

Fig. 10 Differences between single-laser forming area and multi-laser forming area. (a)(c) Laser action zones and molten pools; (b)(d) grain characteristics and defects



工艺参数,单激光成形 GH3536 试样几乎全致密。在 L-PBF 成形过程中,激光束扫描试样表面预铺的粉末层,激光诱导粉末熔化,会产生溅射并且卷入气体,从而形成部分细小孔洞缺陷,并且还会有部分未熔粉末附着在成形试样上[图 10(a)]。当多激光束作用时,激光搭接区域相较于单激光成形区域进行了一次额外的激光重熔过程(四激光束作用区域重熔三次),充分熔化了表面的未熔粉末,从而改善表面质量。但能量输入过大,加速了液态金属的凝固和熔池流动,冷却速度更快,并且熔融粉末的溅射更加显著,气体更容易被困在熔池中,从而残存下来形成孔洞缺陷。此外,能量输入越大,熔池的温度梯度也越大,会产生越大的反冲压力以及越强的 Marangoni 对流,造成液相的过熔和过热,导致熔池稳定性下降,从而在熔池中形成孔洞缺陷。这些缺陷会在拉伸过程中作为失效源,造成多激光成形试样塑性的降低。

此外,由于单激光直接作用在粉末层上,而多激光束直接熔化上一层已经凝固的 GH3536 试样,因此搭接区域的晶粒生长是沿着已成形的柱状晶向上生长的,沿竖直方向形成更深更窄的熔池;并且凝固试样的热导率远远大于金属粉末,因此多激光搭接区域的冷却速度以及温度梯度更大,成形试样沿建造方向的择优生长更为显著,部分晶粒再结晶也造成沿水平方向的柱状晶转变为等轴晶[图 10(d)]。同时,多激光搭接区域的能量起伏需要一定的准备时间,加上冷却速度较快,这些因素产生了极大的过冷度,固液两相的自由能差异过大,即相变驱动力过大,从而形成较小的临界晶核半径,导致形核率的提升。此外,多激光搭接区域的激光重熔具有能量密度大、热加工时间短的特征,在光斑移走后冷却速度很快,晶粒生长时间短,因此多激光成形试样的晶粒得到细化。

激光束对试样的重熔也是一个退火过程,伴随着温度的升高,成形试样内部的残余应力得到释放。残余应力的存在会使试样在拉伸时产生变形和开裂等工艺缺陷,从而造成强度和塑性的降低。尽管多激光成形试样的内部缺陷较多,但残余应力的释放使得其强度与单激光成形试样相当,塑性也达到了 25% 以上。

## 5 结 论

采用四激光 ML-PBF 设备,分别在单激光、双激光以及四激光作用区域成形了 GH3536 试样。通过对试样缺陷、显微组织、残余应力以及力学性能的对比分析,明确了多激光粉末床熔融成形时的缺陷形成机理以及微观组织演变,主要结论如下:

1) 随着成形时引入激光束的增多, GH3536 试样的表面质量得到改善,但是致密度降低了,从单激光成形试样的 99.82% 分别降低到双激光成形试样的 92.35% 以及四激光成形试样的 98.97%。ML-PBF 成形时的激光重熔过程实现了 GH3536 合金相邻熔化道

和相邻层之间的良好冶金,提高了熔池之间的搭接率,造成了柱状晶向等轴晶的转变。

2) 在多激光成形过程中,晶粒经历重熔过程发生了部分再结晶,晶粒的择优生长更加显著。单激光成形试样的织构指数为 3.040,而双激光以及四激光成形试样的织构指数提升到了 3.403 和 3.465,在一定程度上加强了 GH3536 试样的各向异性。同时,多激光成形时的再结晶作用导致搭接区小角度晶界向大角度晶界转变,大角度晶界的比例由 65.9% (单激光区) 逐渐降低到 50.1% (双激光区) 以及 46.3% (四激光区)。

3) ML-PBF 成形能降低试样内部的残余应力,单激光、双激光以及四激光成形 GH3536 试样的残余应力值分别为 192.3、106.5、44.1 MPa。ML-PBF 对 GH3536 试样的强度影响不大,单激光、双激光以及四激光成形试样的强度分别为 858.1、851.4、830.5 MPa。而其中产生的孔洞以及微裂纹缺陷对试样的延伸率有较大的影响,单激光、双激光以及四激光成形 GH3536 试样的延伸率分别为 30.3%、25.9% 和 25.4%。

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# Study on Microstructure and Properties of Overlap Region of GH3536 Alloy Processed by Multi-Laser Powder Bed Fusion

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## Abstract

**Objective** Nickel-based superalloys possess many desirable properties, including high strength, desirable oxidation resistance, and superior thermal stability, and are widely utilized as the preferred materials for crucial hot-end components in the aerospace field. As an important structural material in the aerospace industry, the GH3536 nickel-based superalloy is used to manufacture aero-engine combustion chambers and other high-temperature components of aircraft engines with high operating temperatures and complex structures. Early research on GH3536 mainly focused on deformation behavior, heat treatment, and welding. However, with increasing demand for high-performance, lightweight, and heavy-duty aerospace equipment, higher requirements are placed on traditional complex component manufacturing. Laser powder bed fusion (L-PBF), also known as selective laser melting (SLM), has been introduced to fabricate GH3536 complex-structure components. However, bulkhead, siding, and casing structures tend to be large, and conventional L-PBF technology is not capable of building large parts owing to the limited building dimensions and efficiency. Multi-L-PBF (ML-PBF) technology combines the advantages of high precision and high efficiency and is more suitable for building large-size and complex-structured GH3536 components owing to the composition of multiple single-laser beam modules. However, studies on the influence of defects, microstructures, and mechanical properties of GH3536 parts with different laser beams involved in the ML-PBF process are limited. In this study, using quadruple-laser ML-PBF equipment, the effects of different laser beams on the micro/macro properties of L-PBF-processing GH3536 parts are investigated. In addition, the differences in the defect characteristics, microstructure, residual stress, and tensile properties of the single-, dual-, and quadruple-laser-processing samples are examined. This study is expected to provide a better understanding of multi-laser interactions on the samples, and a scientific basis for the application of nickel-based materials in the aerospace fields.

**Methods** Using the large-size, four-laser ML-PBF equipment and the gas-atomized GH3536 nickel-based superalloy powder particles with the particle size of 21.2–58.9  $\mu\text{m}$ , GH3536 samples were prepared using single-, dual-, and quadruple-laser beams with the optimized process parameters. First, the relative densities of the samples were measured using the Archimedes method and micrograph analysis. The optical microscopy and scanning electron microscopy equipped with an electron back-scattered diffraction (EBSD) detector were employed to examine the microstructures of the cubic specimens. The residual stress of the samples was measured using an X-Ray diffraction (XRD) testing machine. In addition, a high-temperature endurance testing machine was used to test the room-temperature tensile properties of the alloys.

**Results and Discussions** Re-melting during the ML-PBF process melts the unmelted powder particles on the upper surface and penetrates the powder layer better, which helps to improve the surface quality (Fig. 3). With the increase of laser beams involved in the ML-PBF process, the relative density gradually decreases from 99.82% to 92.35% and 98.97%, respectively, which is mainly due to the pores and microcracks produced during the re-melting process (Fig. 4). In the ML-PBF process, grains in the re-melting regions grow on the solidified materials, which hinders the growth of columnar crystals. With an increase in the number of laser beams, a large number of columnar crystals gradually transform into cellular crystals (Fig. 5). The texture index of the samples along the horizontal direction increases from 3.040 (single-laser) to 3.403 (dual-laser) and 3.465 (quadruple-laser), whereas the volume fraction of high-angle grain boundaries (HAGBs) gradually decreases from 65.9% to 50.1% and 46.3%, respectively (Figs. 6 and 7). This is primarily attributed to the recrystallization of grains during the ML-PBF process, which leads to the transformation of HAGBs to low angle grain boundaries (LAGBs), causing a more significant preferred growth of grains and obvious anisotropy of materials. Values of the residual stress of single-, dual-, and quadruple-laser processing samples are 192.3, 106.5, and 44.1 MPa, respectively. The tensile strengths of the samples are 858.1 (single-laser), 851.4, and 830.5 MPa, respectively, while the elongation at break is 30.3%, 25.9%, and 25.4%. The main reason for this may be that ML-PBF can induce pore and microcrack defects, which are stress concentration components that accelerate crack propagation under tensile stress, resulting in premature fracture failure, and thus reducing the elongation of the samples.

**Conclusions** GH3536 nickel-based superalloy is prepared via ML-PBF, and the defects, microstructures, and mechanical properties in single-, dual-, and quadruple-laser-processing regions are investigated. The results indicate that the surface quality improves with an increase in laser beams introduced during the ML-PBF process, while the relative density decreases from 99.82% (single-laser) to 92.35% (dual-laser) and 98.97% (quadruple-laser). Simultaneously, after re-melting in the overlap regions during the ML-PBF inducing recrystallization, the preferred growth orientation along (001) is more apparent, the texture index increases from



3.040 to 3.403 and 3.465, and the volume fraction of LAGBs decreases from 65.9% to 50.1% and 46.3%. Under the multiple laser repeat scanning process, the residual stress in the overlap regions also reduces, where residual stress values of the single-, dual-, and quadruple-laser processing regions are 192.3, 106.5, and 44.1 MPa, respectively. All the samples display an equivalent tensile strength of more than 800 MPa, while the pores and microcracks deteriorate the ductility of the overlap regions. The elongation at break decreases from 30.3% (single-laser) to 25.9% (dual-laser) and 25.4% (quadruple-laser). This work is expected to provide an efficient reference and theoretical guidance for large-size nickel-based superalloy components fabricated via ML-PBF.

**Key words** laser technique; multi-laser powder bed fusion; selective laser melting; nickel-based superalloys; microstructure; mechanical properties