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Effect of laser welding parameters on the microstructure and mechanical properties of dissimilar hastelloy X/Haynes 188 joints

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Abstract	<p>Laser welding is currently a versatile and reliable assembly technique. To save weight in aerostructures and propulsion units and to reduce thermal distortions and the heat-affected zones in these assemblies, autogenous laser welding seems promising. This experimental work addresses dissimilar autogenous laser welding of thin alloy sheets using a Yb:YAG source and a particular optical fibre setting. Hastelloy X and Haynes 188 are the two base metals studied for this assembly. An experimental design allows us to determine the weldability domain. The macro- and microstructure and mechanical properties of the weld seams are investigated. EBSD and EDS analysis revealed microstructural and chemical heterogeneities in the high-power welded seam, but XRD results demonstrated a homogeneous depletion of the carbide density in its fusion zone. Despite such microstructural features, the mechanical properties of Haynes 188 were found to be satisfactory and slightly greater than those of the weaker Hastelloy X.</p>
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Effect of Laser Welding Parameters on the Microstructure and Mechanical Properties of Dissimilar Hastelloy X/Haynes 188 Joints



JOËL ALEXIS, JÉRÉMIE GRANEIX, and YANNICK BALCAEN

Laser welding is currently a versatile and reliable assembly technique. To save weight in aerostructures and propulsion units and to reduce thermal distortions and the heat-affected zones in these assemblies, autogenous laser welding seems promising. This experimental work addresses dissimilar autogenous laser welding of thin alloy sheets using a Yb:YAG source and a particular optical fibre setting. Hastelloy X and Haynes 188 are the two base metals studied for this assembly. An experimental design allows us to determine the weldability domain. The macro- and microstructure and mechanical properties of the weld seams are investigated. EBSD and EDS analysis revealed microstructural and chemical heterogeneities in the high-power welded seam, but XRD results demonstrated a homogeneous depletion of the carbide density in its fusion zone. Despite such microstructural features, the mechanical properties of Haynes 188 were found to be satisfactory and slightly greater than those of the weaker Hastelloy X.

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I. INTRODUCTION

IN jet engines, performance and efficiency are directly related to the combustion temperature. This statement has promoted the constant research and development of new alloys and new thermomechanical routes to produce disks, blades, combustion chambers and exhausts.^[1–3] Several superalloys can be associated with optimizing the different parts of the combustion chambers depending on the mechanical and environmental constraints. The nickel-based alloy Hastelloy X is, for example, associated with the Haynes 188 cobalt-based alloy in projected engines. These alloys are commonly assembled by semi-automatic gas tungsten arc welding (GTAW) techniques with filler wires of the same grade as nickel-based superalloys.^[4] These welding techniques produce a high thermal distortion due to high energy input. In contrast, the laser welding process induces smaller deformations and can be automated.^[5] Moreover, new laser sources, such as disk laser sources, look promising. Currently, disk lasers have a higher power, higher efficiency and higher optical quality than older CO₂ laser sources.^[6,7]

The weldability of a superalloy is inversely proportional to the presence of hardening elements. An acceptable level of weldability has been set at maximum levels of 6 pct at. titanium and aluminium.^[8] The presence of these chemical elements, which determine the transformations γ/γ' as well as chemical elements other than sulphur or phosphorus,^[9] can cause hot cracking by liquation or alternatively embrittlement phenomena after ageing treatment. Liquidation-induced cracks or hot cracks appear at the joints of grains enriched in the solute and are subjected to stress.^[10] To improve their weldability, the base metals (BMs) must in particular have a small grain microstructure with no pre-weld heat treatment.^[11] It is preferable to weld annealed superalloy sheets and not after a quenching and tempering treatment (final condition before use). The annealed condition is more malleable and allows for better accommodation of cooling-induced deformations during welding. The nature of the grain boundaries also seems to play a role in the weld quality. Special grain boundaries significantly reduce the susceptibility to cracking.^[11] Notably, there has been very little documentation regarding the homogeneous Yb:YAG laser welding of the superalloys Hastelloy X and Haynes 188.^[12, 13] This is even more evident for the heterogeneous welding of these two alloys. The success of dissimilar welds involves the metallurgical compatibility of the BMs and filler metal, similar coefficients of thermal expansion and post-weld heat treatment (PWHT) requirements. Researchers interested in GTAW welds, even CO₂ laser welding of these two alloys, reported some problems of liquation, segregation

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79 in the fusion zone (FZ) and the absence of a heat-af-
 80 fected zone (HAZ).^[14] The strength of the welds is
 81 slightly lower than that of the BM, and their ductility is
 82 also weaker. Variations in mechanical properties are
 83 noted according to the nature and size of the FZ
 84 substructures.^[11,12] The most resistant structure seems to
 85 be a fine dendritic structure,^[13] which is obtained by
 86 minimizing the welding energy while maximizing the
 87 solidification rate.

88 It is well known that the control of welding param-
 89 eters plays a leading role in the welding mode (conduc-
 90 tion or keyhole) and the microstructure of the FZ. The
 91 purpose of this paper is to highlight the effect of
 92 Yb:YAG laser welding process parameters on the
 93 dissimilar welding of Hastelloy X and Haynes 188
 94 sheets. The weldability domain of the heterogeneous
 95 assembly is presented in the first section as a function of
 96 the power density and interaction time. Optimum
 97 welding conditions are determined, and the obtained
 98 weld seams are analysed. The microstructure of these
 99 seams is presented as well as their mechanical properties.

100 **II. EXPERIMENTAL METHODS**

101 *A. Materials*

102 Hastelloy X nickel-based alloy and Haynes 188
 103 cobalt-based alloy sheets (1.2 mm thick) were used.
 104 The plates underwent a solution annealing treatment
 105 (1150 °C for 1 hour, water quenching for Hastelloy X;
 106 1475 °C, 1 h, water quenching for Haynes 188). The
 107 average grain size was 90 μm for the Hastelloy X sheet
 108 (# 4, ASTM E112 standard) and 70 μm for the Haynes
 109 188 sheet (# 4.5, ASTM E112 standard). The chemical
 110 compositions and mechanical properties of the materials
 111 are given in Tables I and II, respectively.

112 *B. Laser Welding*

113 The welds were performed on a TruLaser Cell 3000
 114 machine, which was equipped with a TruDisk Yb:YAG
 115 laser source of 3.3 kW and a coaxial optical “2 in 1”
 116 fibre (100 μm centre fibre, and 400 μm peripheral
 117 fibre).^[15] This machine has distinctive features including
 118 the ability to dynamically vary the spot diameter at the
 119 focal point as well as its position along the optical axis.
 120 The Yb:YAG laser beam is generated from a disk, is
 121 guided in an optical fibre and is delivered with a near
 122 top-hat power density distribution at the focal point.
 123 The core fibre (CF) generates a conventional near
 124 Gaussian beam profile, and the outer fibre (OF) delivers
 125 an annular top-hat power distribution. The use of either
 126 fibre configuration provides an adjustable focal diameter
 127 of 110 to 370 μm for the CF and 450 to 750 μm for the
 128 OF. The focal diameters between 370 and 450 μm are
 129 therefore inaccessible.

130 To limit the number of samples while trying to
 131 investigate the influence of multiple parameters of the
 132 laser welding process explained above, experimental
 133 designs were developed using CORICO software, devel-
 134 oped by CORYENT Company. This software provides

Table I. Chemical Composition of the Hastelloy X and Haynes 188 Sheets (Wt Pct)

	Ni	Cr	Co	Mo	W	Fe	Mn	Si	C
Hastelloy X	Bal.	22.1 ± 0.45	1.46 ± 0.01	8.74 ± 0.05	0.71 ± 0.01	18.6 ± 0.16	0.60 ± 0.03	0.32 ± 0.007	0.067 ± 0.004
Haynes 188	23.9 ± 0.27	22.1 ± 0.1	Bal.	0.37 ± 0.004	14.1 ± 0.12	2.02 ± 0.02	0.79 ± 0.01	0.30 ± 0.005	0.088 ± 0.001


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Table II. Mechanical Properties of the Hastelloy X and Haynes 188 Sheets

		E (GPa)	YS (MPa)	UTS (MPa)	A (Pct)	HV0.5
Hastelloy X	RD	208 ± 10	446 ± 3	799 ± 2	55 ± 2	230 ± 10
	45°	211 ± 15	446 ± 11	793 ± 11	53 ± 2	
	TD	190 ± 20	456 ± 5	806 ± 5	58 ± 8	
Haynes 188	RD	188 ± 1	524 ± 10	1057 ± 14	76 ± 2	275 ± 11
	45°	200 ± 1	516 ± 4	1048 ± 3	71 ± 2	
	TD	201 ± 1	527 ± 8	1055 ± 7	72 ± 2	

140 a data analysis tool that allows us to design and process
 141 non-orthogonal experimental designs and manage non-
 142 linear responses. CORICO software evaluates data by
 143 the iconographic correlation method.

144 The optimal welding parameters were determined
 145 using CORICO software based on “desirability” criteria
 146 attributed to the geometric criteria described in the
 147 standard and to the mechanical properties of the beads
 148 such as yield strength and percent elongation. For each
 149 response Y, a regression equation is created from the
 150 iconographic correlations by CORICO software.

$$Y_I = b_0 + b_1 X_{I1} + \dots + b_p X_{pi} + u_i$$

152 b_p
 153 b_p are the coefficients of the partial regression model,
 154 the constant u_i is the residue term and the regressors X_{pi}
 155 are factors or logical combination of factors correlated
 156 to the best answer. From the regression equations,
 157 optimisation was performed using also the CORICO
 158 software, implementing a Monte Carlo optimisation
 159 method. The desirability criterions of this optimisation
 160 process are determined on mechanical properties
 161 enhancement (both static and fatigue) expected by
 162 geometrical features such as a moderately wide weld
 163 bead, with very low or no undercut and strictly no
 164 underfill. All this procedure is identical to previous work
 165 on both alloys, as.^[16]

166 The variable parameters of the different experimental
 167 designs are as follows: laser source power P (W),
 168 welding head travel speed V (m min^{-1}), gas flow rate
 169 D (L min^{-1}), geometrical configuration of the CF or OF
 170 beam, and spot diameter at the focal point ϕ_f (μm). The
 171 first three parameters are common to all welding
 172 processes. The last two parameters are generally asso-
 173 ciated with the machine configuration. In view of the
 174 results of the bibliography, other welding parameters
 175 have been fixed, namely, the welding direction corre-
 176 sponds to the rolling direction, the position of the focal
 177 point is at $-1/3$ of the sheet thickness, and the gas
 178 shielding towards the weld is argon (20 L min^{-1}). Thirty
 179 samples were welded according to the experimental
 180 design given for each geometrical configuration. The
 181 different weld beads were then observed from metallo-
 182 graphic sections to characterize their shape and size.
 183 More detailed information is available in a previous
 184 publication.^[12]

C. Characterization

185
 186 The microstructure of the weld seams was determined
 187 from metallographic cross sections obtained with the
 188 following preparation procedure: polishing to SiC 4000
 189 grit paper and then polishing to $1/4 \mu\text{m}$ diamond paste
 190 and etching for 2 s with 96 pct HCl and 4 pct H_2O_2
 191 solution. Optical microscopy observations made it
 192 possible to determine the weldability domain from the
 193 morphologies of the weld seams and their defects. The
 194 microstructure of the welds was investigated by optical
 195 (PMG3 Olympus), scanning electron microscopy (SEM,
 196 JEOL 7000F) and electron backscatter diffraction
 197 (EBSD) analyses (Oxford Nordlys Fast camera). The
 198 precipitation density and the evolution of the lattice
 199 parameter of the solid solution γ were analysed by
 200 electron diffraction with transmission electron micro-
 201 scopy (TEM, JEOL JSEM 2100F) and X-ray diffraction
 202 (XRD) (line ID15 of the ESRF facility with a wave-
 203 length of 0.0142222 nm). The hardness was determined
 204 using a ZWICK-ROELL ZHU2.5 instrumented hard-
 205 ness tester at a 500 g load. The tensile tests were carried
 206 out using a DEBEN microtensile machine equipped
 207 with a 5 kN load cell at a strain rate of 10^{-3} s^{-1} to
 208 determine the mechanical behaviour of the produced
 209 assemblies.

III. RESULTS AND DISCUSSION

A. Domain of Weldability

210
 211
 212 The weldability domains of the Hastelloy X–Haynes
 213 188 heterogeneous welds are shown in Figure 1 as a
 214 function of the power density and interaction time. No
 215 cracks in the molten zone or at the interface between
 216 molten zone and the heat affected zone were observed by
 217 X-ray analysis or by optical observation of metallo-
 218 graphic section. Almost all the welds made with the CF
 219 exhibited sufficient penetration and a geometry in
 220 accordance with the NF L 06 395 2000 standard, but
 221 the result is different for the welds made with the OF.
 222 Indeed, only a few welds have sufficient penetration and
 223 comply with the criteria of the standard. A minimum
 224 power density of 10^6 W cm^{-2} and a fluence higher than
 225 5 kJ cm^{-2} are therefore necessary to achieve compliant
 226 welds with the OF. The majority of compliant hetero-
 227 geneous welds are common to the homogeneous welds
 228 of both alloys (Figure 2).^[12, 13]

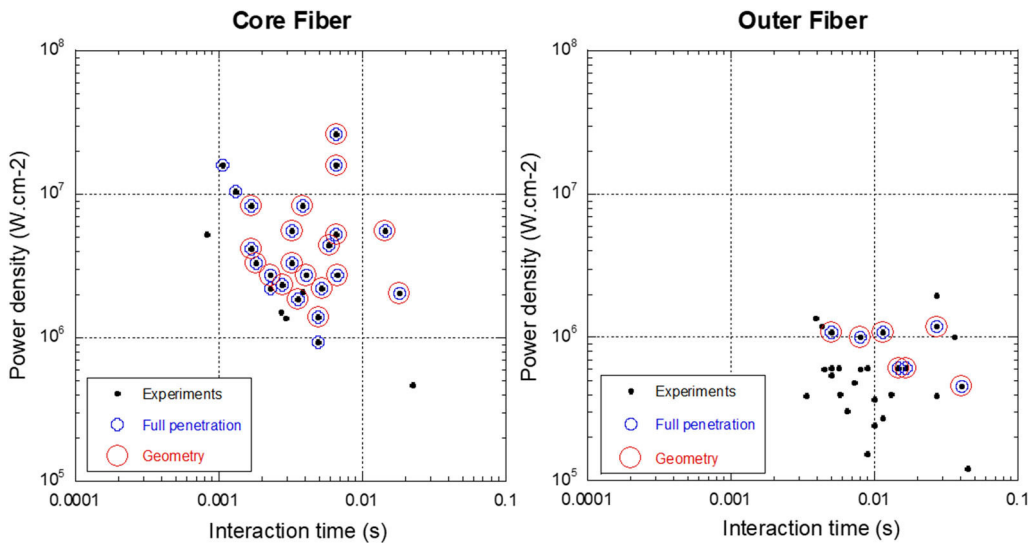


Fig. 1—Weldability domains of the heterogeneous Hastelloy X–Haynes 188 junctions.

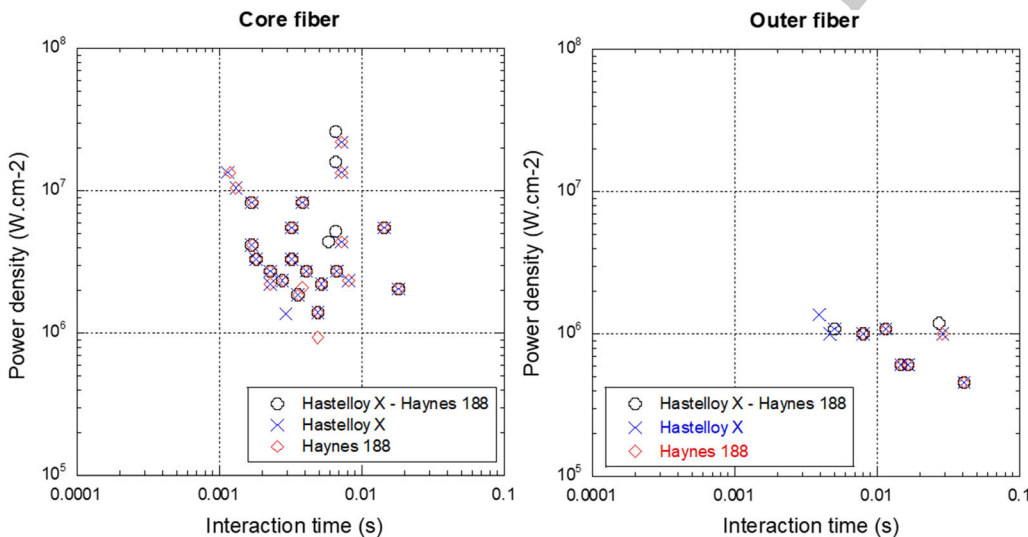


Fig. 2—Weldability domains of the heterogeneous Hastelloy X–Haynes 188 junctions and Hastelloy X or Haynes 188 junctions.

229 The welding conditions optimized to obtain sufficient
 230 penetration and bead geometry that comply with the
 231 abovementioned standard were determined using COR-
 232 ICO software from CORYENT. The optimal processing
 233 conditions are given in Table III.

234 B. Microstructure of the Optimized Welds

235 SEM micrographic cross sections of the obtained weld
 236 seams are presented in Figure 3(a) and (b). The
 237 optimized heterogeneous bead obtained with the core
 238 fibre has well-defined interfaces between the BM and the
 239 FZ. Contrary to what was observed for the homoge-
 240 neous assemblies,^[12] no cracks were found on this
 241 heterogeneous weld bead. This result can be induced by
 242 the low welding speed implemented (1 mm min^{-1}). The
 243 alloys studied, particularly Hastelloy X, are susceptible

244 to liquation-induced hot stress cracking. A low welding
 245 rate leads to a decrease in cooling rate during solidifi-
 246 cation and consequently a decrease in thermal stres-
 247 ses.^[17, 18] The filling of liquefied grain boundaries is also
 248 proposed as factor contributing to the prevention or
 249 reduction of hot cracking especially in laser welding of
 250 Hastelloy X.

251 Some grain boundaries close to the melted zone are
 252 marked, particularly on the Hastelloy X side. Liquation
 253 zones are observable (Figure 3(c)). These different
 254 analyses also remain true for welds made with the OF.
 255 The presence of cellular growth and dendritic zones was
 256 determined from micrographic observations at a higher
 257 magnification. Typical micrographs are shown in
 258 Figure 3(d) to (f). The proportion of dendritic zones is
 259 very important in the Hastelloy X/Haynes 188 core fibre
 260 weld seam. The cellular zone represents only a $150 \mu\text{m}$

261 thick border at the BM-FZ interface. This result is not
 262 the same for the Hastelloy X/Haynes 188 weld obtained
 263 by the OF since the proportions of dendritic and cellular
 264 zones tend to be equal. This last weld is formed of a
 265 broad cellular area close to the BM and a dendritic zone
 266 in its centre.

267 SEM backscattered electron observations revealed
 268 chemical contrasts in the FZ (Figure 4(a), (b)), con-
 269 firmed by energy dispersive spectroscopy (EDS) analy-
 270 ses. The chemical composition of the CF bead FZ
 271 appears to be relatively homogeneous at a macroscopic
 272 scale (Figure 4(d), (f)). EDS profiles confirm this
 273 statement in the centre of the cords. Only slight
 274 variations (approximately 2 to 3 pct) are detected over
 275 100 μm distances. On the other hand, significant
 276 macrosegregations are observed in the Hastelloy X/
 277 Hayne 188-OF cords (Figure 4(c) and (e)). Some areas
 278 of the melted zone present high concentrations of nickel
 279 and molybdenum, a major alloying element of Hastelloy
 280 X, and other areas are cobalt and tungsten rich,

281 typically originating from Haynes 188. These global
 282 analyses were supplemented by local analyses, which
 283 showed a slight increase in the chromium, molybdenum
 284 and tungsten contents in the interdendritic zones.

285 The evolution of the ratio of the thermal gradient to
 286 the solidification rate (G/R) explains the microstructural
 287 changes in the fusion zone. The microstructure of the
 288 weld can progressively change from a planar solidifica-
 289 tion mode at the interface of the weld (maximum G/R),
 290 to a cellular mode and then to a dendritic mode in the
 291 central region of the FZ due to the decrease of G/R
 292 towards the centerline of the FZ. On one hand, the
 293 CF-obtained seam presents an homogeneous distribu-
 294 tion of chemical species, and a dendritic structure. This
 295 could indicate a moderate solidification time and an
 296 enhanced expression of constitutional supercooling. On
 297 the other hand, the weld seam obtained with OF, high
 298 power setting and quite high welding speed, presents a
 299 heterogeneous distribution of solute (Figure 4(e)). The
 300 discontinuous nature of metal-weld solidification pro-
 301 duces the formation of solute bands within the weld.
 302 The growth rate fluctuates cyclically above and below a
 303 mean value which is a function of the welding speed.
 304 Correlated to the presence of a thicker cellular zone,
 305 exposed at Figure 3(b), it could be related to a higher
 306 solidification rate, and steeper thermal gradient, limiting
 307 the abovementioned phenomenon.

308 The microstructures observed for the heterogeneous
 309 bead obtained with the CF are very similar to those of
 310 the homogeneous cords (Figure 5).^[10] The solidification
 311 of the FZ begins with epitaxial growth from partially
 312 melted grains. This zone obeys a cellular growth mode.
 313 Then, the growth of some favourably oriented grains,
 314 presenting (100) to the joint plane, appears. The welding

Table III. Processing Parameter Values for Optimized Welds According to Criteria Based on the Standard NF L 06 395 2000

	Core Fibre	Outer Fibre
Power (W)	765	2330
Welding Speed (m min^{-1})	1	4.62
Spot Size (μm)	203	527
Face Argon Flow (l min^{-1})	35	40
Power Density (kW cm^{-2})	2363	1277
Fluence (kJ cm^{-2})	28.8	8.7

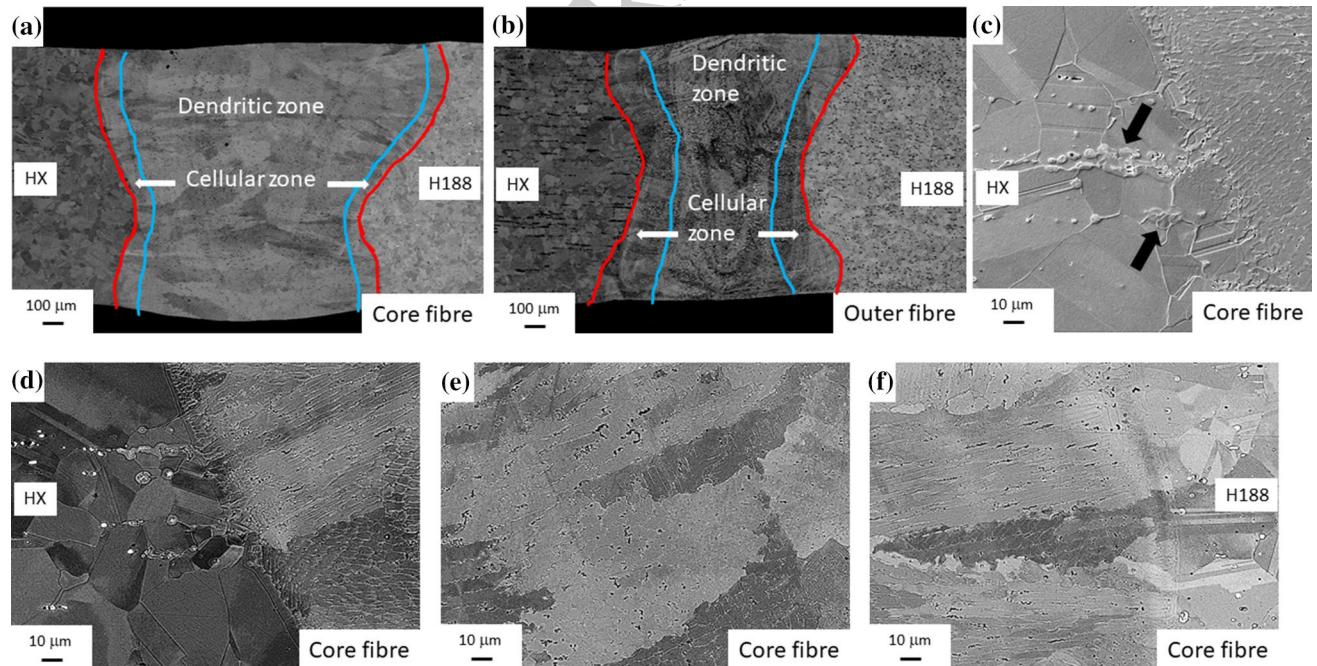


Fig. 3—Cross-sectional observations of the optimized weld by scanning electron microscopy. (a, b) Overall observation and (c, d) observation of the fusion zone–Hastelloy X interface..

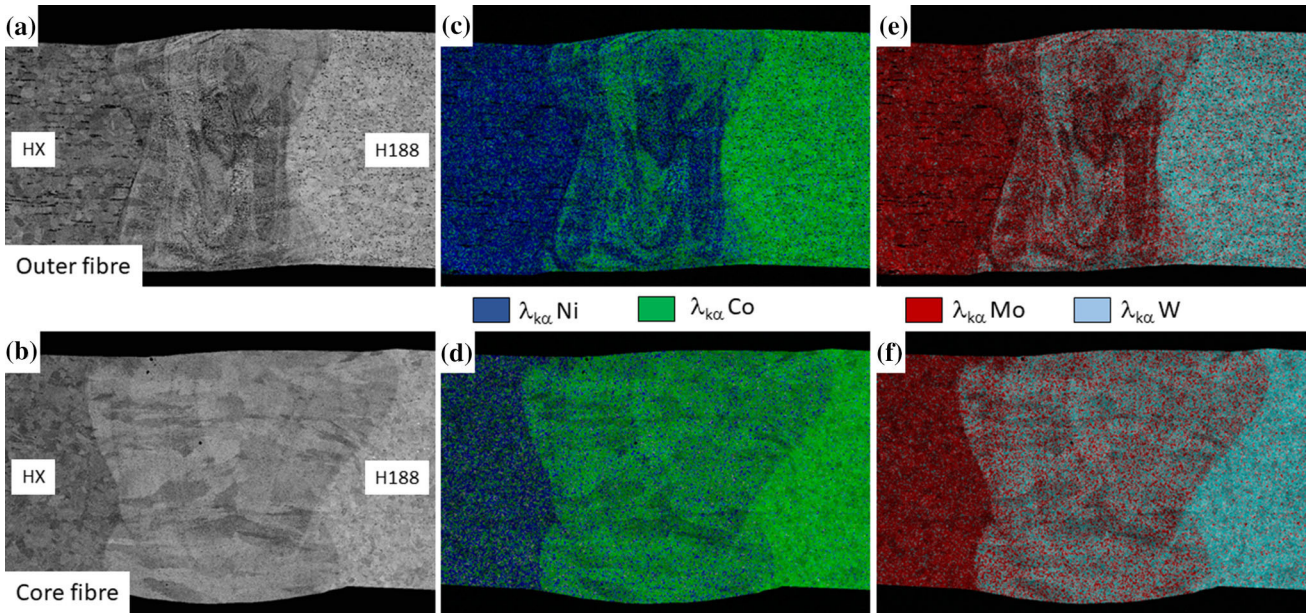


Fig. 4—EDS mappings made in the TD-ND plane of the heterogeneous assemblies of Hastelloy X/Haynes 188. Optimized welds obtained by (a, c, e) outer fibre and (b, d, f) core fibre..

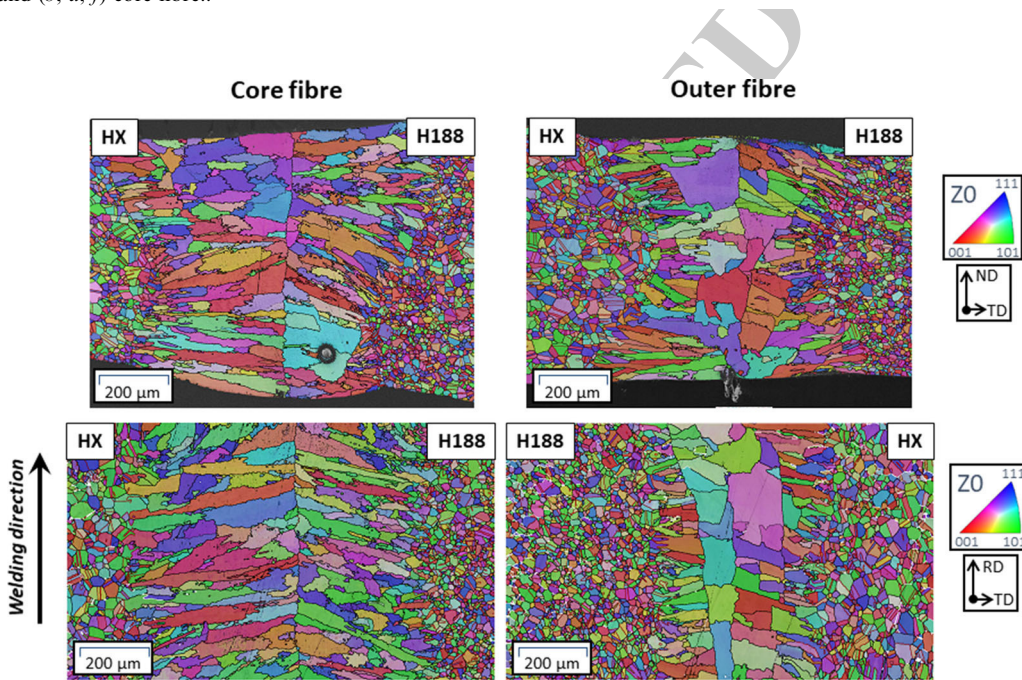


Fig. 5—IPF Z_0 coded EBSD cartographies of the optimized welded seams of the heterogeneous Hastelloy X/Haynes 188 assemblies, on the cross sections (top) and longitudinal sections $400 \mu\text{m}$ above the median plane of the sheet (bottom)..

315 speed seems sufficiently important to extend the melt
 316 pool backwards; the grains then develop in the direction
 317 of the higher temperature gradient. The OF weld
 318 microstructures are significantly different. A columnar
 319 zone consisting of epitaxially grown grains remains.
 320 However, the width of this area is very limited. On the
 321 other hand, large odd-shaped grains are observed in the
 322 centre of the weld. The joint plane is then more diffuse in
 323 this cord. Comparing data displayed on Figures 3(b)
 324 and 4(c) and 5 top right, it seems that odd-shaped grains

325 appearance is correlated to dendritic microstructure in
 326 poorly-mixed solute areas. These microstructural evolu-
 327 tions have not been demonstrated in homogeneous
 328 seams, even if some optimized homogenous assemblies
 329 have been developed with very close processing param-
 330 eters. The possible explanation of this particular
 331 microstructure could be related to the poor mixing of
 332 the chemical species in the weld pool, leading to steep
 333 chemical gradients in it, promoting such heterogeneous
 334 nucleation.

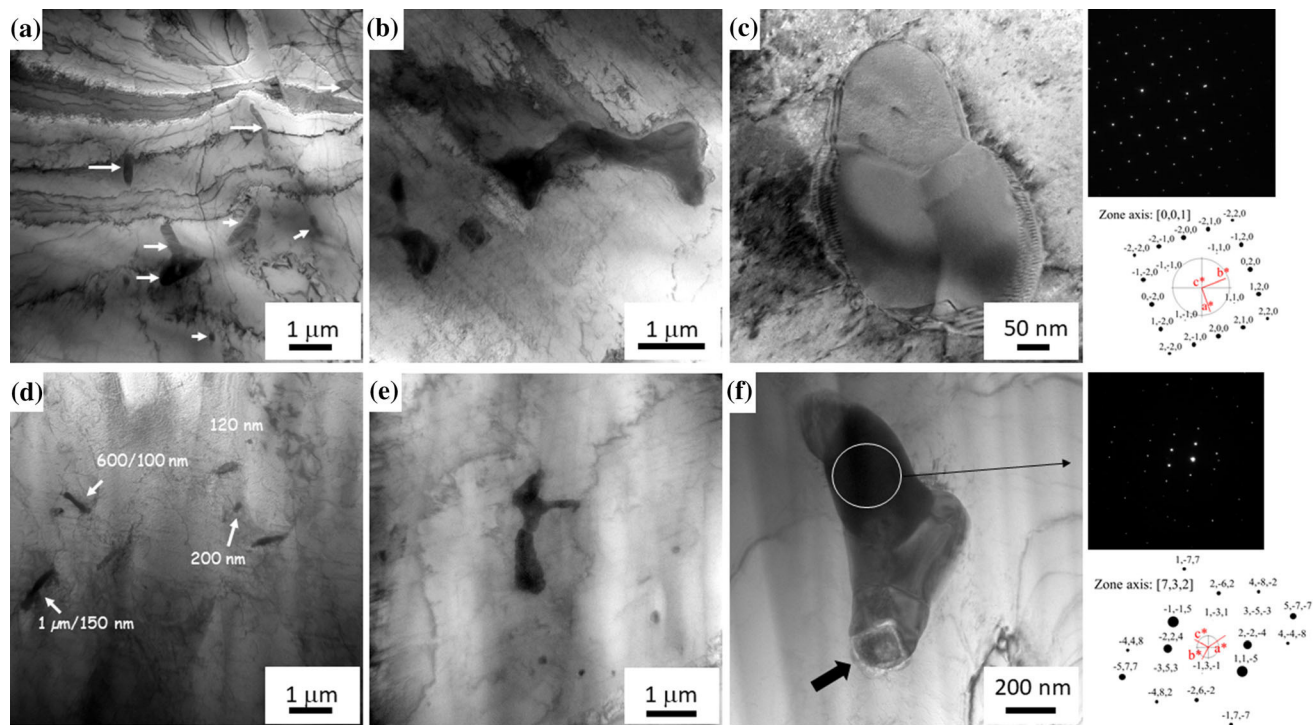


Fig. 6—MET observations of the precipitates present in the weld bead obtained with the (a to c) CF and (d to f) OF. The (c) sigma phase precipitate and (f) M_6C carbide. Indexing of the diffraction patterns with Carine software..

335 The shape and size of the particles in FZs are highly
 336 variable (Figure 6(a) to (f)). In addition, these particles
 337 differ from those observed in the two reference alloys.
 338 Their shape in FZs can be characterized as tortuous and
 339 very rounded. The particles whose sizes were less than
 340 500 nm in the Haynes 188 alloy had an angular shape,
 341 and the larger particles were rounded. For the Hastelloy
 342 X alloy, particles below 500 nm were rounded. The
 343 structure of some particles was determined by diffraction.
 344 The selected area diffraction pattern of the oriented
 345 particles, presented in Figure 6(c), can be indexed
 346 according to the [001] zone axis of the tetragonal σ
 347 phase (P42/mnm space group, lattice parameters $a =$
 348 0.91 nm and $c = 0.48$ nm). Figure 6(f) shows the pattern
 349 obtained for one of the observed carbides. The 0.40 nm
 350 inter-reticular distance corresponds to the (220) planes
 351 of the face-centred cubic (FCC) M_6C carbide. To
 352 confirm the nature of the M_6C carbides, several zone
 353 axes were analysed on some particles (Figure 7).

354 In the melted zone of the OF seams, EDS analysis
 355 highlighted two typical particles: carbides, identified as
 356 M_6C , and lanthanum-rich particles (Table IV). The
 357 M_6C particles generally surround the lanthanum-rich
 358 particles (Figure 6(f)). Regarding the particles in the FZ
 359 of the CF, three typical compositions are highlighted:
 360 σ -phase particles, M_6C carbides and lanthanum-rich
 361 particles. The chemical composition of the M_6C carbides
 362 present in the molten zone whatever the fiber used (CF
 363 or OF) is different from that of the carbides analyzed in
 364 each alloy. They contain both Molybdenum and Tung-
 365 sten. In addition, lanthanum-rich particles are not
 366 detected in the base alloys. The precipitation of these
 367 second phases (M_6C , σ) in the interdendritic zones have

368 already been observed by many authors.^[19, 20] A. 368
 369 Ghasemi et al. show that the M_6C and σ phases appear 369
 370 at the end of solidification of Hastelloy X alloy between 370
 371 1315 and 1280 °C due to the segregation of chromium 371
 372 and molybdenum and not during a solid state transfor- 372
 373 mation during cooling.^[21] 373

374 The evolution of the precipitate densities in the FZ 374
 375 was determined via X-ray analysis by the evolution of 375
 376 the M_6C (422) peak intensity in the FZ compared to the 376
 377 same peak in the BM (Figure 8). The X-ray probe spot 377
 378 was assimilated to a 100 μm diameter disk, scanning the 378
 379 total width of the seam from the BM to BM through the 379
 380 sheet thickness. Superimposing the geometry of the 380
 381 cords on these data, the intensity ratio does not change 381
 382 identically with respect to the two CF and OF cords 382
 383 (Figure 9). With the CF, the population of M_6C 383
 384 precipitates decreases progressively from the BM to 384
 385 the centre of the FZ. Therefore, there appears to be a 385
 386 precipitate density gradient within the FZ. The width of 386
 387 the zone containing a very small proportion of M_6C 387
 388 carbide in the centre of the FZ for the CF condition 388
 389 corresponds approximatively to the width of the laser 389
 390 beam for a power density greater than 10^6 W cm^{-2} . The 390
 391 dissolution seems more important in this area. For OF 391
 392 welding, this progressive evolution disappears, and the 392
 393 peak relative intensity remains constant in the FZ at the 393
 394 20 pct level. These results are comparable to those of the 394
 395 Haynes 188 homogeneous assemblies.^[12] 395

396 The lattice parameter of the solid solution in the FZ is 396
 397 different from that of the nickel-rich solid solution 397
 398 (Hastelloy X) or cobalt-rich solid solution (Haynes 188) 398
 399 (Figure 10). At the centre of the welds, an intermediate 399

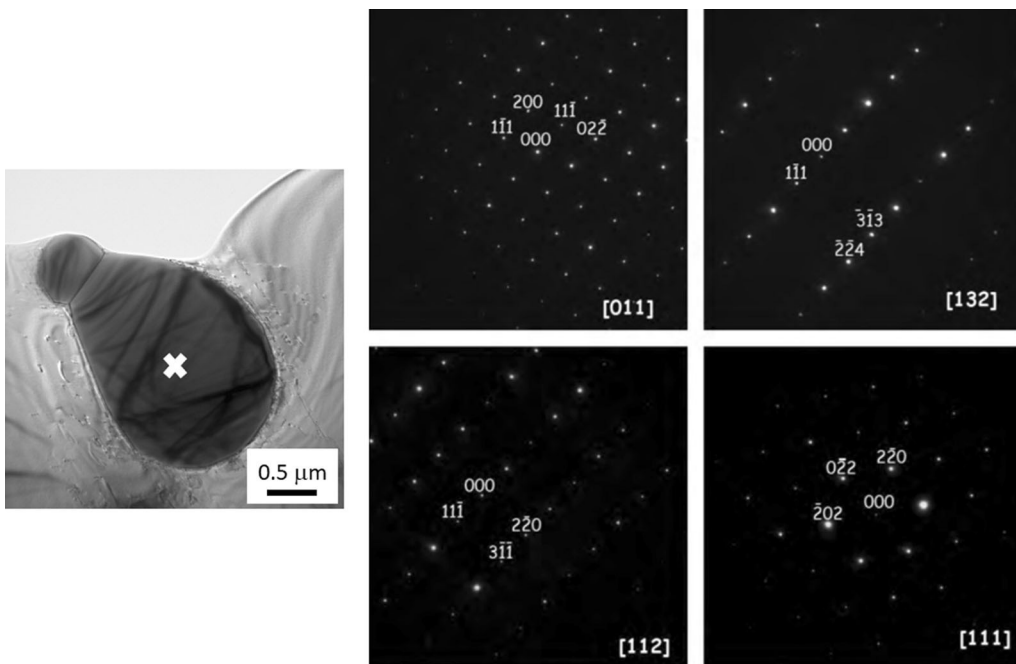


Fig. 7—Electron diffraction patterns of an M_6C carbide for four different zone axis orientations [011] [111] [112] and [132].

Table IV. Chemical Composition of Particles Observed in Alloys and Welds

Atomic Percent	Cr	Ni	Co	Fe	Mo	W	La
M_6C Carbide (Hastelloy X Alloy)	27	26	—	10	35-37	1	—
M_6C Carbide (Haynes Alloy)	27-28	11-15	23-28	1	0.5-1	29-33	—
M_6C Carbide (OF Weld)	35-40	17-23	10-19	6-8	10-18	5	—
M_6C Carbide (CF Weld)	43	16	16	7	18	6	—
σ Phase (CF Weld)	33	29	15	9.5	9	4	—
Lanthanum-Rich Precipitates (OF and CF Welds)	6-18	4-23	3-10	1-7	2-5	0-1	36-83

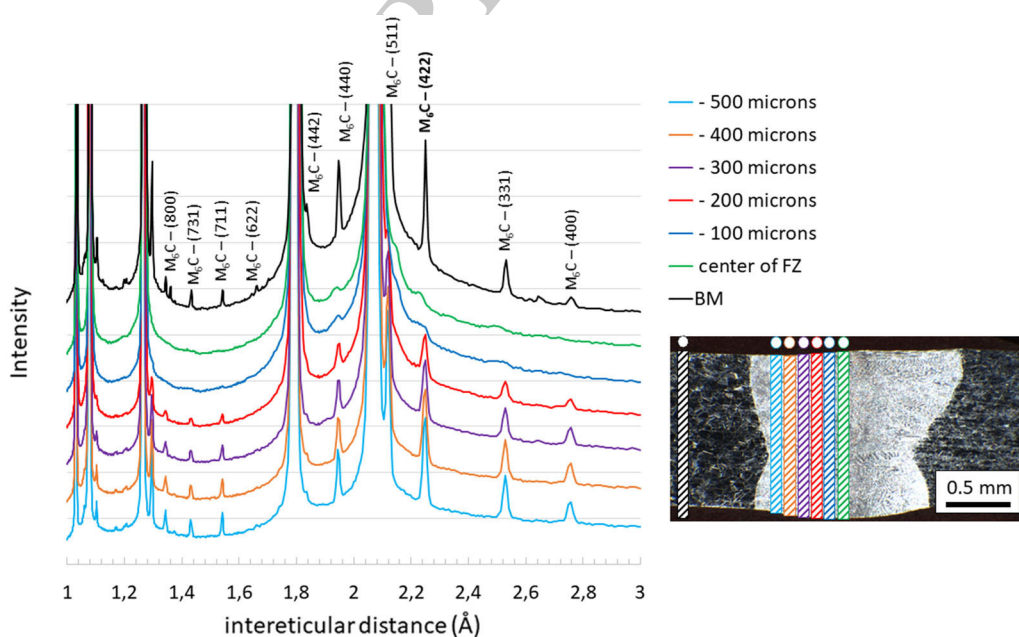


Fig. 8—Superimposed X-rays diffractograms obtained through weld seam.

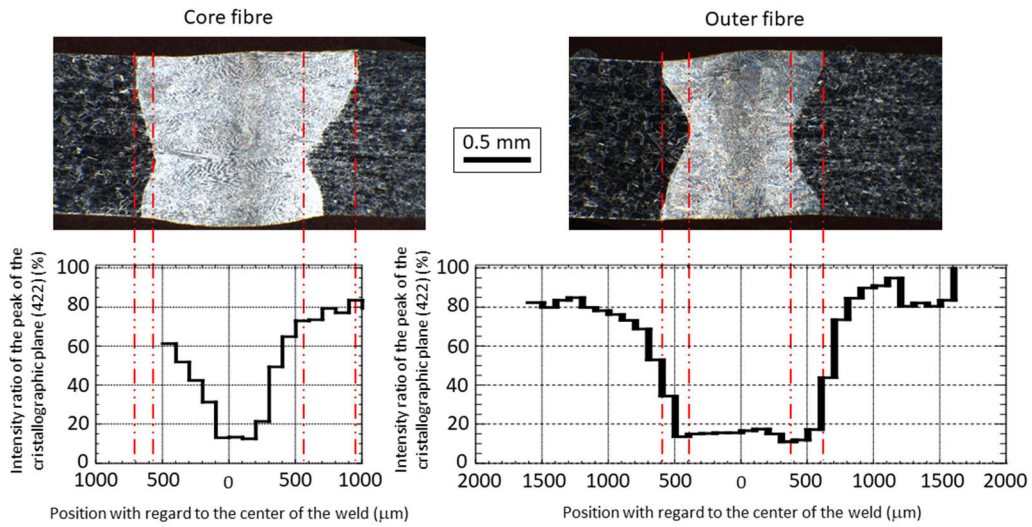


Fig. 9—Location of the X-ray diffraction analysis of the Hastelloy X–Haynes 188 welds obtained in the CF and OF (top). Diffraction intensity ratio by the (422) planes of the M_6C precipitates in the weld with respect to the base metal..

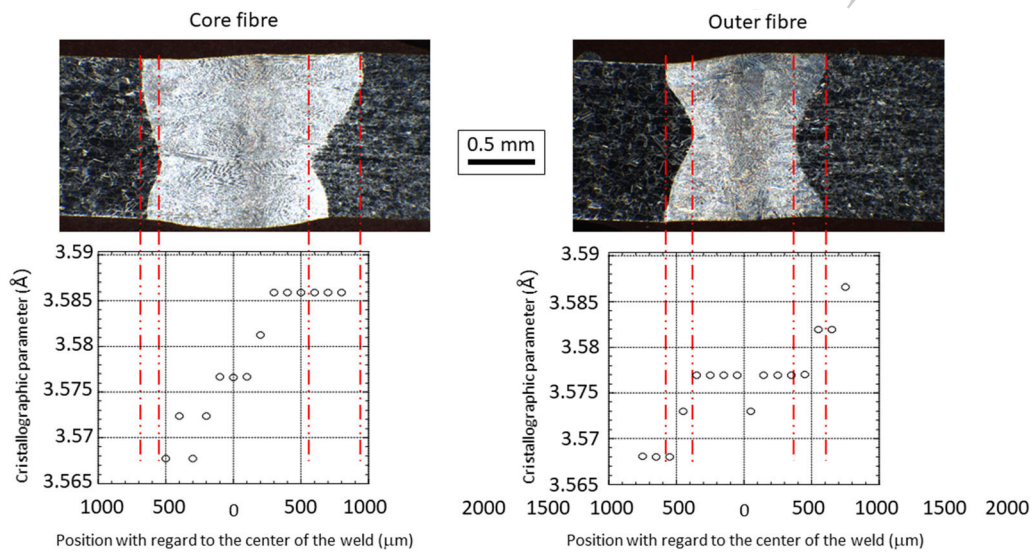


Fig. 10—Evolution of the FCC solid solution lattice parameter..

400 value is taken because of the interdiffusion phenomena
 401 between Ni and Co in the bead.

402 C. Mechanical Properties of the Optimized Welds

403 The mechanical behaviour of the welded specimens is
 404 explored through tensile tests, and the results are
 405 presented in (Figure 11). The tensile behaviour is
 406 intermediate between the two BMs. The mechanical
 407 strength of the welded specimens, regardless of the
 408 selected fibre, is actually close to that of the Hastelloy X
 409 alloy (Table V). A slight reduction in the ductility is
 410 recorded for the CF welds compared to the OF welds.
 411 The fractures of all the specimens are localized in the

Hastelloy X BM, several millimetres from the FZ. This
 finding explains the values of the mechanical properties
 of the heterogeneous weld seams solicited in transverse
 directions. The mechanical strength of these welded
 specimens is in fact conditioned by the maximum
 strength of the weaker Hastelloy X. The slight difference
 in the ductility recorded is explained by a purely
 geometric effect (different width of the welds depending
 on the type of fibre compared to the length of the useful
 zone). The mechanical properties of the welds (FZ and
 HAZ) were determined using tensile specimens extracted
 in the FZ along the welding direction (Figure 9). The
 results obtained for the welds confirm the first results in
 the transverse direction. The mechanical properties of

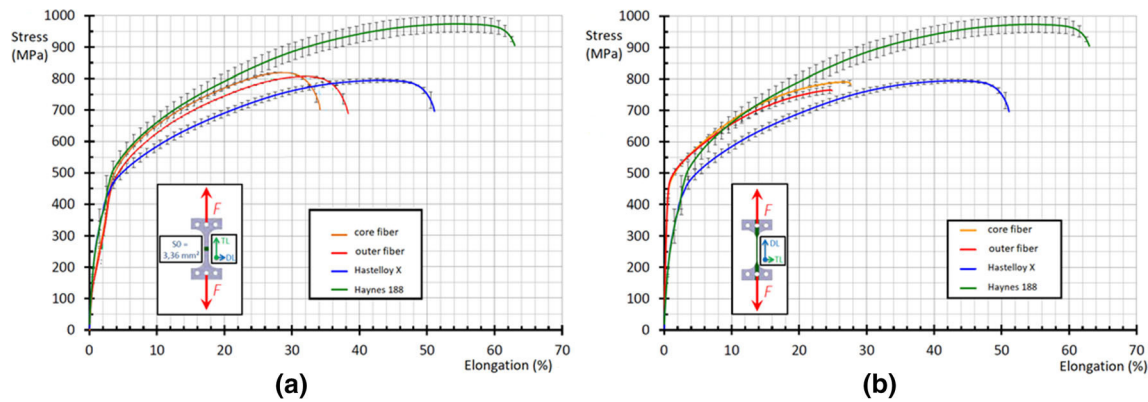


Fig. 11—Tensile tests performed on heterogeneous and homogeneous welds in the transverse direction (a) and welding direction (b)..

Table V. Mechanical Properties of Heterogeneous Assemblies Hastelloy X/Haynes 188 (Tensile Tests in the Transverse Direction)

Test Direction		UTS (MPa)	Elongation (Pct)	Fracture Localization
Transverse Welding Direction	Hastelloy X	795 ± 8	48 ± 0.1	—
	Haynes 188	974 ± 36	57 ± 1.7	—
Transverse Welding Direction	core fibre	819 ± 2	29 ± 0.6	Hastelloy X BM
Transverse Welding Direction	outer fibre	807 ± 2	33 ± 0.3	Hastelloy X BM
Welding Direction	core fibre	792 ± 4	29 ± 1.6	—
Welding Direction	outer fibre	767 ± 13	25 ± 1.6	—

the fused zone solicited in the welding direction for the Hastelloy X/Haynes 188 heterogeneous welds are close to those of the Hastelloy X welds (Table V).

IV. CONCLUSION

Two different optima were determined for each type of fibre, essentially based on geometrical defect reduction. Both optimums produce power densities above 10^6 W.cm² but very different fluence levels, leading to a counter-intuitive wider FZ for the CF weld seam than for the OF weld seam. The obtained weld seams are sound, with weak liquations. The microstructure of the FZ consists of a fine cellular growth zone, which evolves dendritically in the centre of the FZ. The width of the cellular zone is more important in the selected OF welding condition due to the high welding speed and low fluence levels, leading to a high solidification factor G/R . This high solidification rate is also confirmed by the poor solute mixing evidenced by EDX analysis of the OF weld seam. This behaviour could be the origin of the odd-shaped dendritic grains in the centre of the FZ, highlighted in EBSD analysis. Secondary phases, such as carbides, were found to be strongly and homogeneously diluted in this seam, contrary to the CF weld seam, in which the carbide density diminished progressively from the BM to the centre of the FZ. However, the mechanical properties remain satisfactory, and fractures occur in the weaker Hastelloy X.

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CONFLICT OF INTEREST

On behalf of all authors, the corresponding author states that there is no conflict of interest.


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